

Effect of Rolling and Heat Treatment on Microstructure,
Texture Development, and Mechanical Anisotropy of
AA6xxx Aluminum Alloys

by

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FOREWORD

To my father, my first strength.

To my mother, my endless comfort.

To Mahsa, my love, my heart and peace.

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Microstructure, évolution de la texture et anisotropie mécanique des alliages d'aluminium AA6xxx laminés et traités thermiquement

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RÉSUMÉ

Cette étude porte sur l'évolution microstructurale, la texture cristallographique et le comportement de recristallisation des alliages d'aluminium AA6063 et AA6082 soumis au laminage à froid, au laminage pseudo-à-chaud et à un recuit post-laminage. Le travail compare des trajectoires de laminage unidirectionnel et de laminage croisé, à partir d'états initiaux bruts de coulée et bruts d'extrusion, afin d'examiner l'influence du trajet de déformation, du niveau de déformation et de la microstructure initiale sur l'affinement des grains, la texture de déformation, les mécanismes de restauration et l'évolution de la texture de recristallisation. La diffraction des électrons rétrodiffusés (EBSD) a été utilisée pour caractériser la microstructure, la texture cristallographique et le comportement de recristallisation, les régions recristallisées étant identifiées à l'aide de critères de désorientation intragranulaire fondés sur les données EBSD. La microscopie électronique à balayage en électrons rétrodiffusés (BSE-SEM) et la spectroscopie à dispersion d'énergie des rayons X (EDS) ont été utilisées pour caractériser la morphologie, la distribution et la composition élémentaire des particules de seconde phase. Au cours du laminage à froid, la texture évolue progressivement à partir d'orientations initiales faibles ou héritées vers des composantes plus prononcées caractéristiques de la texture de laminage des matériaux à structure cubique à faces centrées (CFC) ; toutefois, la trajectoire dominante d'évolution de la texture dépend du mode de laminage et de l'état initial. Le laminage unidirectionnel favorise un développement plus continu de la texture de déformation, tandis que le laminage croisé modifie l'équilibre et la stabilité des principales composantes de texture à forte déformation. Dans le cas de l'AA6082 à l'état brut d'extrusion, la texture initiale d'extrusion et la microstructure fibreuse continuent d'influencer la réponse au laminage jusqu'à ce qu'une déformation sévère perturbe cette structure d'orientation héritée. Lors du laminage pseudo-à-chaud, la texture et la microstructure évoluent sous l'effet combiné de la déformation et des mécanismes de restauration dynamique, avec des réponses différentes observées pour les alliages AA6063 et AA6082. Le recuit post-laminage met en évidence des comportements distincts de restauration et de recristallisation en fonction de l'historique de déformation antérieur et des conditions de recuit. Ces différences soulignent le fort couplage entre l'historique de déformation et l'évolution microstructurale ultérieure dans les alliages de la série AA6xxx.

Mots-clés : Alliages d'aluminium AA6xxx, procédé de laminage, trajet de déformation, évolution de la texture, mécanisme de restauration

Microstructure, Texture Evolution, and Mechanical Anisotropy of Rolled and Heat Treated AA6xxx Aluminum Alloys

Majid YAZDANI-ESMAEILABAD

ABSTRACT

This study investigates the microstructural evolution, crystallographic texture, and recrystallization behavior of AA6063 and AA6082 aluminum alloys subjected to cold rolling, pseudo-hot rolling, and post-rolling annealing. The work compares unidirectional and cross-rolling routes in as-cast and as-extruded initial conditions to examine how strain path, deformation level, and starting microstructure affect grain refinement, deformation texture, restoration behavior, and recrystallization texture evolution. Electron backscatter diffraction (EBSD) was used to characterize the microstructure, crystallographic texture, and recrystallization behavior, with recrystallized regions identified using EBSD-based intragranular misorientation criteria. BSE-SEM and EDS were used to characterize the morphology, distribution, and elemental characteristics of second-phase particles. During cold rolling, the texture progressively evolved from weak or inherited initial orientations toward more pronounced FCC rolling-texture components; however, the dominant texture path depended on the rolling route and initial condition. Unidirectional rolling promoted a more continuous development of deformation texture, whereas cross rolling changed the balance and stability of the main texture components at high strain. The as-extruded AA6082 condition showed that the initial extrusion texture and fibrous microstructure continued to affect the rolling response until severe deformation disrupted this inherited orientation framework. During pseudo-hot rolling, texture and microstructure evolved under the combined effects of deformation and dynamic restoration, with different responses observed between AA6063 and AA6082. Post-rolling annealing revealed distinct restoration and recrystallization responses depending on the prior deformation history and annealing condition. These differences highlight the strong coupling between deformation history and subsequent microstructural evolution in AA6xxx alloys.

Keywords: AA6xxx aluminum alloys, Rolling process, Strain path, Texture evolution, Restoration mechanism

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LIST OF ABBREVIATIONS AND ACRONYMS

Abbreviation/Acronym	Definition
Materials & Processing	
AA	Aluminum alloy
AA6xxx	6000 series aluminum alloys
AA6063	Aluminum alloy 6063
AA6082	Aluminum alloy 6082
Al	Aluminum
Al–Mg–Si	Aluminum–Magnesium–Silicon
CCR	Cross cold rolling
CHR	Cross pseudo-hot rolling
TMP	Thermomechanical processing
UCR	Unidirectional cold rolling
UHR	Unidirectional pseudo-hot rolling
Crystallography & Microstructure	
CDRX	Continuous dynamic recrystallization
DDRX	Discontinuous dynamic recrystallization
FCC	Face-centered cubic
GDRX	Geometric dynamic recrystallization
GND	Geometrically necessary dislocations
HAGB	High-angle grain boundary
LAGB	Low-angle grain boundary
MDRX	Meta-dynamic recrystallization
NR	Non-recrystallized
PSN	Particle-stimulated nucleation
RX	Recrystallization
Characterization & Analysis Techniques	
DIC	Digital image correlation
EBSD	Electron backscatter diffraction
EDS	Energy dispersive X-ray spectroscopy
FEG-SEM	Field emission gun scanning electron microscope
GOS	Grain orientation spread
GOS/D	Grain orientation spread divided by grain diameter
IPF	Inverse pole figure
KAM	Kernel average misorientation
ODF	Orientation distribution function
PF	Pole figure
SEM	Scanning electron microscopy

Abbreviation/Acronym	Definition
Mechanical Properties & Directions	
ED	Extrusion direction
ND	Normal direction
RD	Rolling direction
TD	Transverse direction
RRSS	Relative resolved shear stress
TF	Taylor factor
UTS	Ultimate tensile strength
Texture Components	
Bs	Brass
Cb	Cube
Cu	Copper
Gs	Goss
RCu	Rotated-Copper
RGs	Rotated-Goss
Institutions	
CMQ	Centre de métallurgie du Québec
CNRC	Conseil national de recherches Canada / National Research Council Canada

LIST OF SYMBOLS AND UNITS OF MEASUREMENT

Symbol	Definition / Unit
Mechanical Properties & Testing	
E	Young's modulus (GPa)
e	Engineering strain (%)
n	Strain-hardening exponent (dimensionless)
r	Lankford coefficient / plastic strain ratio (dimensionless)
$\bar{r}$	Average normal anisotropy (dimensionless)
Δr	Planar anisotropy (dimensionless)
σ	True stress / flow stress (MPa)
σ_y	Yield strength (MPa)
σ_{UTS}	Ultimate tensile strength (MPa)
ε	True strain (dimensionless)
$\dot{\varepsilon}$	Strain rate (s ⁻¹)
Thermomechanical Processing	
T	Temperature (°C or K)
t	Time (s, min, or h)
Z	Zener–Hollomon parameter / temperature-compensated strain rate (s ⁻¹)
Crystallography & Texture	
$\{hkl\}$	Miller indices for crystallographic planes (dimensionless)
$\langle uvw \rangle$	Miller indices for crystallographic directions (dimensionless)
M	Taylor factor (dimensionless)
m	Schmid factor (dimensionless)
V_f	Volume fraction (%)
$\varphi_1, \Phi, \varphi_2$	Bunge Euler angles defining crystallographic orientation (°)
θ	Misorientation angle (°)

Symbol	Definition / Unit
Microstructure & EBSD Analysis	
D	Grain diameter / average grain size (μm)
GOS	Grain orientation spread ($^{\circ}$)
GOS/D	Grain orientation spread normalized by grain diameter ($^{\circ}/\mu\text{m}$)
KAM	Kernel average misorientation ($^{\circ}$)
r_p	Particle or precipitate radius (μm)
Angles and Directions	
0°	Parallel to the rolling direction / extrusion direction; longitudinal direction
45°	Diagonal to the rolling direction
90°	Perpendicular to the rolling direction; transverse direction
$^{\circ}$	Degree; angle measurement unit

INTRODUCTION

AA6xxx aluminum alloys are widely used in lightweight structural applications because they combine relatively low density, corrosion resistance, formability, and heat-treatable mechanical properties (Sheikh, Naik, & Samantaray, 2026). These alloys are based mainly on the Al–Mg–Si system, where Mg–Si-based precipitates provide an important precipitation-strengthening mechanism and allow the final properties to be adjusted through heat treatment (Sheikh et al., 2026). Their mechanical and microstructural response is therefore controlled not only by chemical composition, but also by the thermomechanical processing route applied to the material. In particular, rolling and subsequent heat treatment can modify grain morphology, dislocation structure, crystallographic texture, recrystallization behavior, and directional mechanical response (Driver & Engler, 2004). These effects are important because the final performance of processed aluminum products depends on how deformation and restoration mechanisms interact during processing.

Within the AA6xxx family, AA6063 and AA6082 represent two relevant Al–Mg–Si alloys with different processing responses. AA6063 and AA6082 differ in their alloying levels and resulting mechanical performance; AA6082 generally contains higher Mg and Si levels and can achieve higher strength in heat-treated conditions, making it particularly relevant for structural applications (European Aluminium Association, 2002; Sheikh et al., 2026). Because of their different alloying levels and expected differences in recovery and recrystallization behavior, comparing these two alloys under similar rolling and heat-treatment conditions provides a useful basis for understanding how composition and initial condition affect texture evolution and restoration behavior. In this thesis, AA6063 and AA6082 were therefore selected to document how two AA6xxx alloys respond to different deformation and thermal processing routes.

Rolling was selected as the main deformation process because it provides a controlled framework for studying the effect of deformation level and strain path on microstructure and texture evolution. During rolling, the thickness reduction, deformation temperature, and rolling route influence grain refinement, stored deformation energy, and the development of

crystallographic texture (Driver & Engler, 2004; Nayan et al., 2019). In unidirectional rolling, deformation is imposed along a constant rolling direction, while in cross rolling, the sample is rotated between passes. This change in strain path can disturb the continuity of the deformation texture, modify the stability of specific texture components, and affect the later response during annealing (Chrominski & Lewandowska, 2022; Nayan et al., 2019).

Previous studies have shown that strain path and deformation temperature strongly affect microstructure, texture, and restoration behavior during rolling (Lypchanskyi et al., 2024; Nayan et al., 2019). However, the combined effects of alloy composition, initial microstructural condition, rolling path, and deformation temperature on texture evolution and subsequent recrystallization have received less systematic attention in AA6xxx alloys. This provides the motivation for the comparative approach adopted in this thesis.

To document the behavior of AA6xxx aluminum alloys under different processing conditions, this thesis examines cold rolling, pseudo-hot rolling, post-rolling annealing, and selected mechanical anisotropy responses. Cold rolling was used to study strain accumulation, grain subdivision, and deformation-texture evolution as a function of increasing reduction. Pseudo-hot rolling was used to examine the material response when deformation occurs together with dynamic restoration mechanisms. Post-rolling annealing was then applied to evaluate the transition from the deformed state toward recovered or recrystallized structures. Finally, selected tensile tests were used to relate the processed microstructures and textures to directional mechanical behavior.

Three initial material conditions were considered: as-cast AA6063, as-cast AA6082, and as-extruded AA6082. The as-cast materials provide reference conditions with relatively weak initial texture and are used to compare the response of the two alloys during rolling. The as-extruded AA6082 condition provides a contrasting initial state, characterized by a fibrous microstructure and inherited extrusion texture. This condition was used to document how a pre-existing fibrous structure and strong initial texture influence subsequent rolling, annealing, and texture evolution.

The main objective of this thesis is to document the microstructural evolution, crystallographic texture development, recrystallization behavior, and mechanical anisotropy of AA6063 and AA6082 under different rolling and post-rolling heat-treatment conditions. More specifically, the work examines how rolling route, deformation level, deformation temperature, initial material condition, and annealing temperature affect grain refinement, deformation texture, dynamic restoration, static recrystallization, and recrystallization texture evolution. This approach makes it possible to compare the effect of unidirectional and cross rolling under cold and pseudo-hot conditions and to identify how the prior deformed state influences the annealing response.

Electron backscatter diffraction was used as the main characterization method to analyze grain morphology, grain-boundary structure, crystallographic texture, and recrystallized fraction (Hansen & Juul Jensen, 2011). EBSD-based parameters, including grain orientation spread and grain orientation spread normalized by grain diameter, were used to separate recrystallized and non-recrystallized regions during annealing analysis. This is important because recrystallization does not occur uniformly in deformed aluminum alloys; it depends on local stored energy, internal misorientation, prior texture, and the stability of specific texture components (A. Ayad, Ramoul, Rollett, & Wagner, 2021). Microtensile testing was used as a complementary analysis to document the directional mechanical response of selected processed conditions.

This thesis is organized as follows. Chapter 1 reviews the background related to AA6xxx aluminum alloys, deformation process and texture analysis. Chapter 2 presents the research context, including the research problem, the main objective, and the specific objectives of the thesis. Chapter 3 describes the materials, rolling procedures, post-rolling annealing treatments, EBSD characterization, texture analysis, recrystallization partitioning, and mechanical testing methods. Chapter 4 presents the microstructural and texture evolution during cold rolling and pseudo-hot rolling. Chapter 5 discusses post-rolling annealing, including recovery, recrystallization, and recrystallization texture evolution. The appendix presents complementary mechanical testing results and anisotropy analysis for selected rolling conditions.

CHAPTER 1

LITERATURE REVIEW

1.1 AA6xxx Aluminum Alloys: AA6063 and AA6082

The AA6xxx series refers to heat-treatable wrought aluminum alloys based mainly on the Al–Mg–Si system. In the aluminum alloy designation system, the first digit identifies the main alloy family, and 6xxx alloys are mainly associated with magnesium and silicon additions. Wrought aluminum alloys are intended for hot and/or cold working processes, such as rolling, extrusion, forging, and drawing. They are generally divided into heat-treatable and non-heat-treatable groups. Heat-treatable alloys can be strengthened through suitable thermal treatment. Within this classification, the 6xxx series belongs to the heat-treatable Al–Mg–Si alloy family, as shown in Figure 1-1 (European Aluminium Association, 2002).

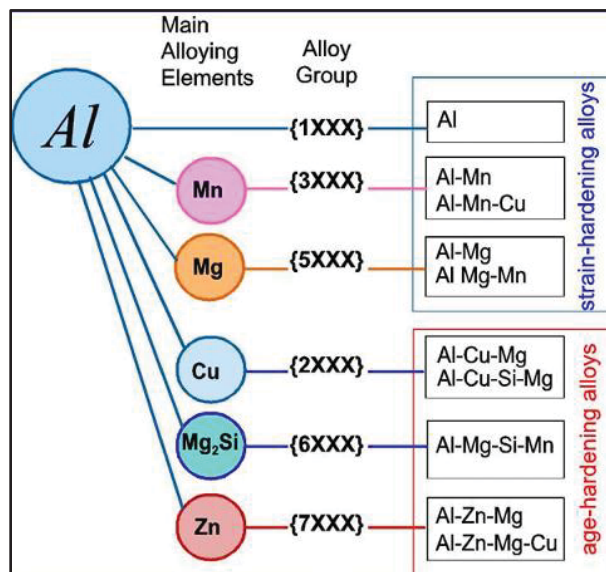


Figure 1-1 Classification of conventional aluminum alloy groups and the position of 6xxx alloys within the Al–Mg–Si heat-treatable alloy family

Taken from European Aluminium Association (2002)

In AA6xxx alloys, magnesium and silicon are the principal alloying elements responsible for precipitation hardening. Mg combines with Si to form Mg_2Si -based precipitates, which provide an important strengthening mechanism during solution treatment and aging. The amount and balance of Mg and Si affect precipitation behavior, microstructural stability, intermetallic phase formation, and the response to thermomechanical processing (Sheikh et al., 2026). This compositional balance also explains why different AA6xxx alloys can show different strength levels and processing responses. As shown in Figure 1-2, AA6063 is located in a lower Mg–Si composition range, while AA6082 contains relatively higher Mg and Si contents. Therefore, AA6063 is generally associated with better formability, surface quality, and extrudability, whereas AA6082 has a stronger precipitation-hardening potential and is more suitable for higher-strength structural applications (Belov, Eskin, & Aksenov, 2005; European Aluminium Association, 2002).

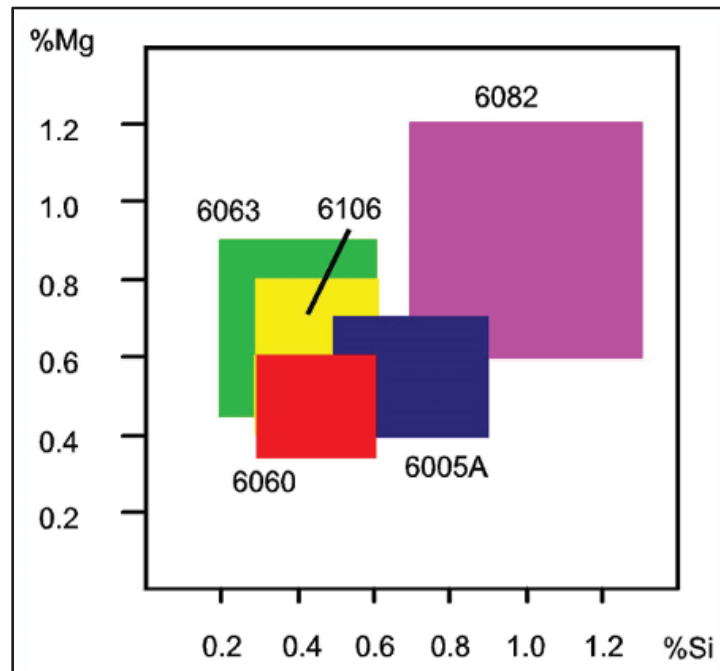


Figure 1-2 Approximate Mg and Si composition ranges of selected AA6xxx alloys, showing the relative positions of AA6063 and AA6082 within the Al–Mg–Si alloy family

Taken from European Aluminium Association (2002)

This precipitation-hardening response gives AA6xxx alloys a useful combination of strength, formability, corrosion resistance, and heat-treatability (Humphreys & Hatherly, 2004; Nasi, Dhoska, Sulejmani, Koka, & Koça, 2026). Minor additions of elements such as Cu, Cr, Mn, Sc, and Zr can further modify precipitation behavior, grain structure, recrystallization resistance, and final mechanical properties (Harma & Sanders, 2025). Dispersoid-forming elements such as Cr, Mn, Sc, and Zr are particularly relevant because they can influence grain-boundary mobility and microstructural stability during thermomechanical processing (Harma & Sanders, 2025).

AA6xxx alloys are widely used in engineering applications because they combine relatively low density, moderate-to-high strength, good formability, corrosion resistance, and heat-treatability (Dragatogiannis et al., 2019; Sheikh et al., 2026). These characteristics make them attractive for lightweight transportation applications, where weight reduction, mechanical performance, and corrosion resistance are important design requirements (Dragatogiannis et al., 2019; Poznak, Freiberg, & Sanders, 2018). The AA6xxx series is also used in structural, architectural, marine, and selected aerospace applications because its properties can be adjusted through alloy composition, thermomechanical processing, and heat treatment (Sheikh et al., 2026; Nasi et al., 2026). Within this alloy family, AA6063 and AA6082 are important because they represent two commonly used Al–Mg–Si alloys with different combinations of strength, formability, surface quality, corrosion resistance, and processing response (Sheikh et al., 2026; Nasi et al., 2026).

AA6063 is an Al–Mg–Si alloy commonly used in applications where formability, surface quality, corrosion resistance, and moderate strength are required. This alloy is widely associated with extruded products because of its good extrudability and ability to produce profiles with smooth surface finish (Nasi et al., 2026). Nasi et al. reported that AA6063 is used in architectural profiles, decorative elements, transport panels, furniture, and lightweight frame-type applications where surface condition and corrosion resistance are important (Nasi et al., 2026). The same study reported a tensile strength of approximately 247 MPa for AA6063, indicating that the alloy provides useful mechanical strength while retaining favorable forming and finishing characteristics (Nasi et al., 2026). These characteristics make

AA6063 relevant for studying the effect of rolling and heat treatment on microstructure, texture, and mechanical behavior in a relatively formable AA6xxx alloy.

AA6082 is another Al–Mg–Si alloy used in applications where higher structural performance is required (Sheikh et al., 2026). This alloy is commonly associated with automotive, structural, and marine components because it combines strength, corrosion resistance, formability, and heat-treatability (Dragatogiannis et al., 2019; Sheikh et al., 2026). Dragatogiannis et al. stated that 6xxx aluminum alloys generally provide medium-to-high strength, high corrosion resistance, and good formability, which supports their use in lightweight transportation applications (Dragatogiannis et al., 2019). Studies on AA6082 also show that its mechanical and corrosion performance are strongly affected by processing and heat-treatment conditions, including aging, surface condition, and thermomechanical history (Sheikh et al., 2026). Lypchanskyi et al. also studied AA6082 in the T6 condition and showed that its mechanical response during elevated-temperature deformation is strongly controlled by the interaction between deformation mechanisms and precipitation behavior (Lypchanskyi et al., 2024). Therefore, AA6082 is relevant for studying how rolling and post-rolling heat treatment influence microstructure, texture, and mechanical behavior in a higher-strength AA6xxx alloy.

1.2 Rolling and Strain-path Control

Rolling is a metal forming process where a metallic workpiece undergoes plastic deformation by passing between two rotating rolls separated by a gap smaller than the initial material thickness as presented in Figure 1-3 (Driver & Engler, 2004). During this interaction, compressive forces reduce the workpiece thickness while the material flows forward, resulting in elongation along the rolling direction (Groover, 2020). This method is widely employed to produce flat aluminum products, including sheets, plates, and foils (Driver & Engler, 2004). The evolution of microstructure and overall properties in crystalline materials is determined by processing parameters such as thickness reduction, roll gap, rolling speed, friction, strain path, strain rate, and temperature (Nayan et al., 2019).

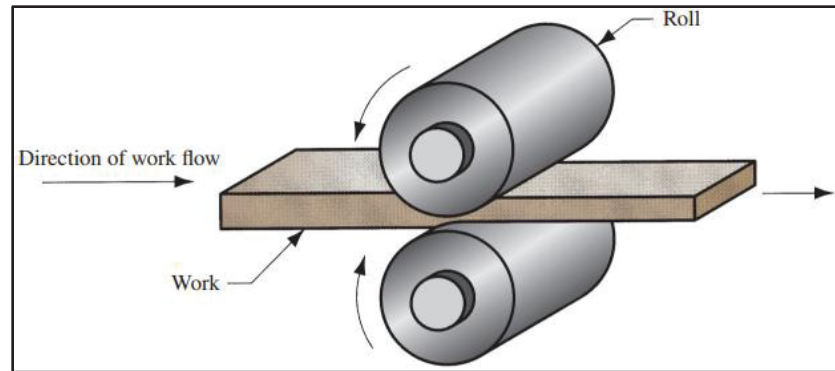


Figure 1-3 The rolling process - flat rolling
Taken from Groover (2020)

Rolling is categorized into cold and hot processes based on the operating temperature relative to the material's recrystallization point (Driver & Engler, 2004). In the present study, both processes are utilized to evaluate the response of AA6xxx series alloys under varying deformation regimes. The deformation imposed during rolling is controlled primarily through the reduction in sheet thickness. For an initial thickness h_0 and a final thickness h_f , the percentage thickness reduction is calculated as:

$$R = \frac{h_0 - h_f}{h_0} \times 100\% \quad (1-1)$$

The corresponding logarithmic strain magnitude in the thickness direction is:

$$\varepsilon_t = \ln \left(\frac{h_f}{h_0} \right) = \ln \left(\frac{1}{1 - R} \right) \quad (1-2)$$

where R is expressed as a decimal in the second expression. Therefore, increasing the thickness reduction increases the plastic strain imposed on the sheet, making thickness reduction one of the principal parameters used to control deformation during rolling (Groover, 2020).

Cold rolling is a deformation process performed at or near room temperature without prior heating of the alloy (Groover, 2020). Since the material does not experience the same thermal softening as in hot rolling, a higher rolling load is generally required to achieve thickness reduction (Driver & Engler, 2004). Cold rolling is commonly used when controlled thickness reduction, good dimensional accuracy, and improved surface finish are required (Humphreys & Hatherly, 2004). The cold reduction level is an important processing parameter because it defines the amount of plastic strain introduced during rolling (Driver & Engler, 2004).

Hot rolling is a thermomechanical process carried out at elevated temperatures where the material's resistance to deformation is significantly reduced (Groover, 2020). This reduction in flow stress allows the alloy to undergo larger thickness reductions per pass with lower rolling forces compared to cold processing. The process is highly sensitive to the interplay between temperature, strain rate, and reduction, which collectively govern the uniformity of deformation through the thickness. In AA6xxx processing, hot rolling is often utilized in the initial stages to refine homogenized material before subsequent cold rolling or aging steps. The elevated temperature enables dynamic microstructural changes, making it an essential step for controlling the starting condition of the alloy (Aryshenskii, Hirsch, Yashin, Sergei, & Kawalla, 2018).

Strain path refers to the rolling route through which deformation is imposed on the sheet. In unidirectional rolling, the sheet is rolled along the same rolling direction during processing, so the rolling direction remains constant and deformation is mainly accumulated along one direction. In cross rolling, the sheet is rotated during processing, commonly by 90° around the normal direction, so the rolling direction changes and RD and TD can be interchanged between passes. This change in rolling direction can destabilize the existing deformation structure and crystallographic orientation developed in the previous pass, because the material is forced to deform under a new strain path. Therefore, unidirectional rolling represents a constant strain path, while cross rolling represents a changed strain path in which deformation is imposed along different in-plane directions during rolling (Nayan et al., 2019).

1.3 Annealing and Restoration Mechanisms

Post-rolling heat treatment and annealing are applied to modify the highly defected microstructure, elongated grains, and deformation texture generated during rolling. During annealing, the restoration of a plastically deformed material is mainly governed by two mechanisms: recovery and recrystallization. Both mechanisms are driven by the reduction of stored deformation energy (Raabe, 2014). When these processes occur during annealing after deformation, they are referred to as static recovery and static recrystallization. When they occur during deformation, they are referred to as dynamic recovery and dynamic recrystallization (Imran & Walther, 2016).

Recovery refers to the early stage of thermally activated restoration in a deformed material. It involves the rearrangement and annihilation of point defects and dislocations, which lowers the stored internal energy without fully replacing the deformed grain structure by newly formed grains (Hansen & Juul Jensen, 2011; Raabe, 2014). Recovery does not usually change the overall grain structure or grain-boundary topology. Instead, it mainly modifies the dislocation arrangement inside the deformed grains. As schematically shown in Figure 1-4, dislocations can annihilate or rearrange into lower-energy configurations through climb, and cross-slip. These processes can lead to the development or refinement of subgrain structures and low-angle boundaries (Humphreys & Hatherly, 2004). Static recovery can occur during annealing at temperatures below the recrystallization temperature and does not require a nucleation stage. Dynamic recovery occurs during deformation, especially in high-stacking-fault-energy metals such as aluminum, where dislocation climb and cross-slip are easier and can promote dislocation reordering and subgrain formation (Humphreys & Hatherly, 2004; Raabe, 2014).

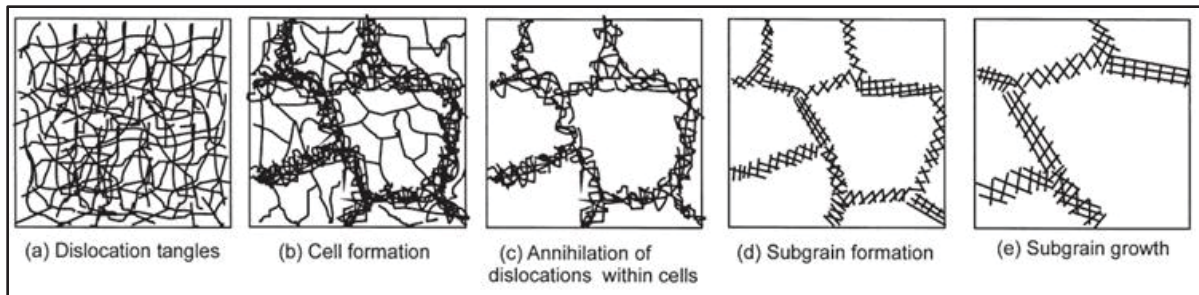


Figure 1-4 Various stages in the recovery of a plastically deformed material

Taken from Humphreys & Hatherly (2004)

In contrast, recrystallization refers to the formation and growth of new, nearly strain-free grains within the deformed matrix. This process occurs through the formation and migration of high-angle grain boundaries (Hansen & Juul Jensen, 2011). Recrystallization therefore replaces the deformed microstructure with new grains that contain much lower defect density (Raabe, 2014). A cold-worked material generally needs to be heated to about 0.3–0.5 of its absolute melting temperature for recrystallization to occur. For engineering purposes, the recrystallization temperature is often defined as the temperature at which 50 vol.% of the cold-worked material recrystallizes after 1 h. However, this temperature is not fixed and depends on material purity, since high-purity materials can recrystallize at lower temperatures than impure materials (Raabe, 2014).

In recrystallization, two main phenomenological forms can be distinguished: discontinuous recrystallization and continuous recrystallization. In discontinuous recrystallization, the process occurs heterogeneously through nucleation and growth. New recrystallized grains form at specific sites and then grow into the deformed matrix by high-angle boundary migration. As schematically shown in Figure 1-5, this mechanism produces a clear replacement of the strain-hardened structure by new recrystallized grains. In continuous recrystallization, there is no clear nucleation-and-growth stage. Instead, the deformed substructure evolves gradually through recovery, subgrain growth, increasing boundary misorientation, and the progressive formation of high-angle boundaries. As shown in Figure 1-6, this mechanism is associated with the gradual transformation of subgrains into recrystallized grains. Therefore,

discontinuous recrystallization is mainly associated with localized nucleation and growth, while continuous recrystallization is associated with gradual microstructural evolution (Humphreys & Hatherly, 2004).

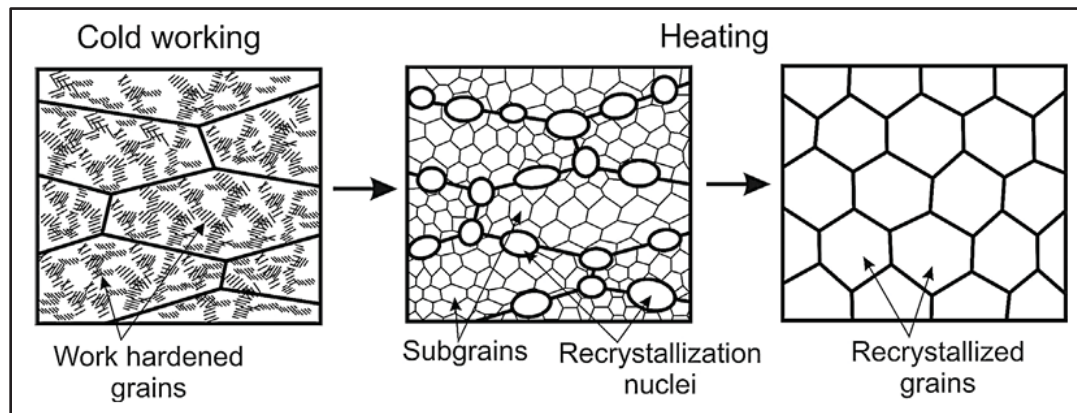


Figure 1-5 Schematic representation of discontinuous static recrystallization during annealing of a strain-hardened material, showing the formation and growth of recrystallization nuclei

Taken from Sakai, Belyakov, Kaibyshev, Miura, & Jonas (2014)

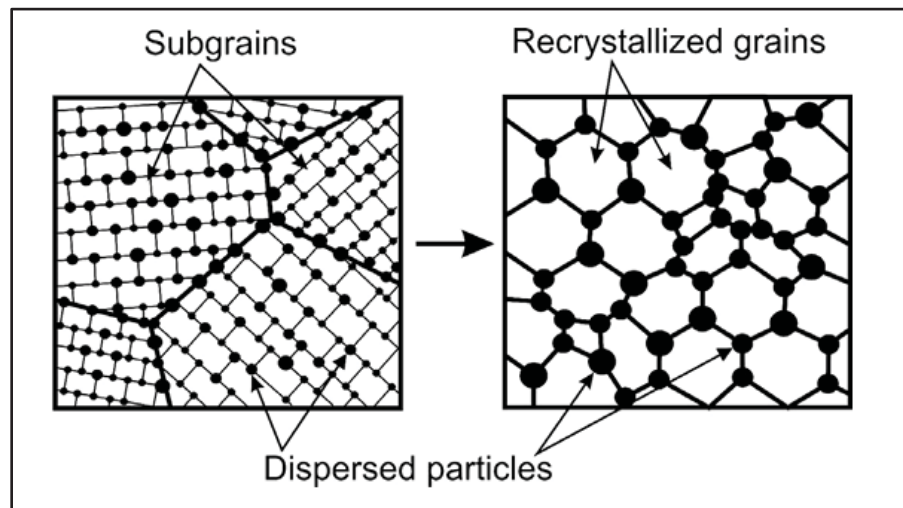


Figure 1-6 Schematic representation of continuous static recrystallization, showing the gradual transformation of subgrains into recrystallized grains

Taken from Sakai et al. (2014)

Recovery and recrystallization can also compete with each other because both consume stored deformation energy. If recrystallization occurs and the deformation substructure is consumed, further recovery becomes limited. Conversely, extensive recovery before recrystallization can reduce the stored energy and decrease the driving force for recrystallization. As a result, prior recovery can change both the kinetics and the nature of recrystallization. The boundary between recovery and recrystallization is not always clear because recovery-related rearrangements can also contribute to the formation of recrystallization nuclei (Humphreys & Hatherly, 2004).

The outcome of post-rolling heat treatment depends strongly on prior processing variables, including the initial deformation route, cold reduction level, and annealing temperature (P. Du et al., 2024; Imran & Walther, 2016). Variations in the annealing path, such as multi-step recovery and recrystallization treatments, can refine grain morphology and weaken mechanical anisotropy (L. Huang et al., 2017; Mao & Yang, 2014). In addition, second-phase particles can affect microstructural evolution during post-rolling heat treatment. For example, large intermetallic particles can promote particle-stimulated nucleation (PSN), leading to a more random recrystallization texture rather than a strong Cube texture (Kumar, Gupta, Dandekar, & Khatirkar, 2020).

1.4 Crystallographic Texture, EBSD Characterization, and Plastic Anisotropy

Texture refers to the preferred crystallographic orientation of grains in a polycrystalline material. In a material with a random orientation distribution, grains do not show a dominant orientation, and the macroscopic response is closer to isotropic. However, thermomechanical processing, such as rolling, extrusion, and annealing, can make certain crystallographic orientations more frequent than others. This non-random distribution of grain orientations is known as crystallographic texture. Texture is important because many properties of polycrystalline materials, especially anisotropy, depend on the orientation distribution of grains (Kestens & Pirgazi, 2016). In rolled aluminum sheets, texture also affects the directional mechanical response, formability, and surface quality of the material (Fukutomi & Okayasu,

2016). The texture state can therefore change during deformation and subsequent annealing, depending on the processing route and the restoration mechanisms that occur after rolling (Traka, 2022).

Texture can be represented using pole figures, inverse pole figures, and orientation distribution functions (ODFs). A pole figure is a stereographic projection that shows the distribution of a selected crystallographic pole or plane normal with respect to the sample reference directions, such as the rolling direction (RD), transverse direction (TD), and normal direction (ND). The idealized pole figures for a drawn wire and a rolled sheet are shown in Figure 1-7. An inverse pole figure (IPF) shows which crystallographic directions are aligned with a selected sample direction, such as RD, TD, or ND. In EBSD maps, IPF coloring is often used to show the crystallographic orientation of each measured point relative to a chosen sample direction. An example of an inverse pole figure for cold-drawn aluminum wire is shown in Figure 1-8. However, pole figures and inverse pole figures do not fully describe the complete orientation distribution of all grains. For this reason, the orientation distribution function is used to describe the frequency of crystallographic orientations in three-dimensional orientation space, commonly using Euler angles. This representation also allows important texture fibers in FCC metals, such as the α -, β -, γ -, and τ -fibers, to be identified, as shown in Figure 1-9 (Humphreys & Hatherly, 2004).

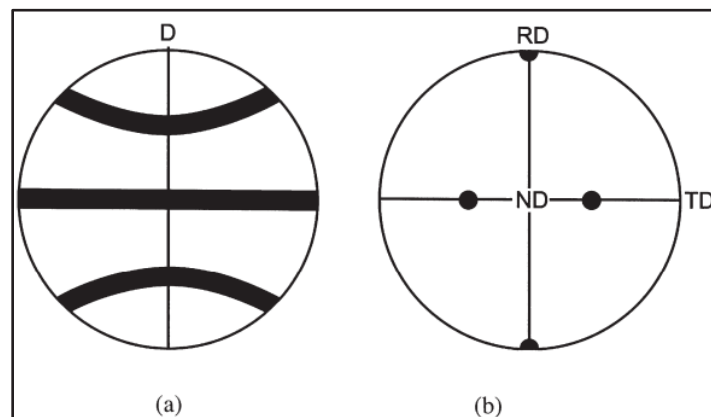


Figure 1-7 Idealized 100 pole figures for: (a) drawn wire, showing $\langle 110 \rangle$ fibre texture, (b) rolled sheet, showing $\{110\}\langle 001 \rangle$ rolling texture

Taken from Humphreys & Hatherly (2004)

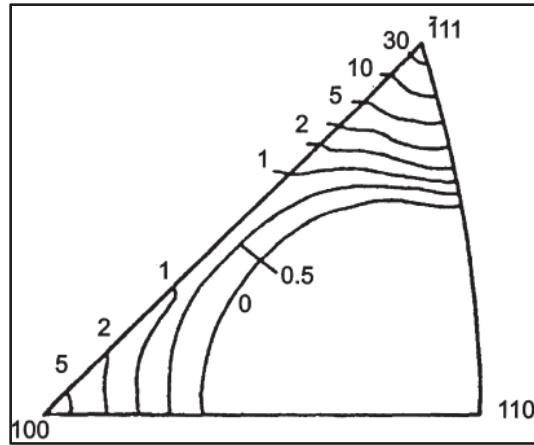


Figure 1-8 Inverse pole figure for cold-drawn aluminum wire, showing the distribution of crystallographic directions aligned with the drawing direction

Taken from Humphreys & Hatherly (2004)

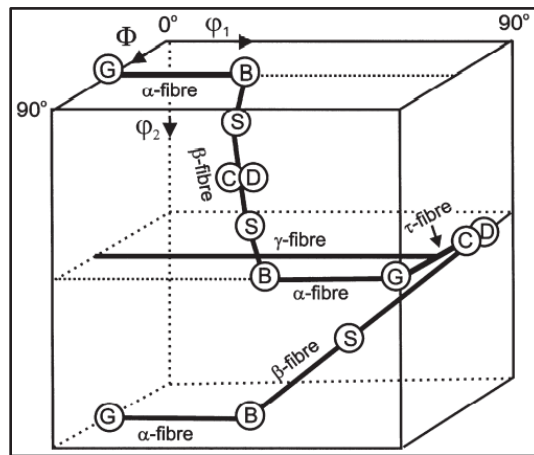


Figure 1-9 Representation of important texture fibers in FCC materials in Euler space

Taken from Humphreys & Hatherly (2004)

Electron backscatter diffraction (EBSD) is commonly used to characterize crystallographic orientation at the local scale. It can provide orientation measurements from small selected regions or from larger mapped areas of a sample surface. These orientation data can then be displayed pixel by pixel to produce orientation maps. In recrystallization and texture studies, EBSD is useful because it allows microstructure and texture to be analyzed together. It can be used to obtain individual grain orientations, construct orientation maps, support texture

analysis from many measured points, and examine boundary misorientation, grain and subgrain size, and recrystallized fraction. Therefore, EBSD is well suited for studying deformed and annealed microstructures (Humphreys & Hatherly, 2004).

1.4.1 Deformation Texture

Deformation texture refers to the preferred crystallographic orientation distribution that develops during plastic deformation, such as rolling (Kestens & Pirgazi, 2016). During deformation, slip is activated on specific crystallographic planes and directions, and the active slip systems depend on the initial grain orientation and the imposed loading condition. As deformation proceeds, grain rotation affects the formation of dislocation boundaries, orientation gradients, and local misorientations. Therefore, texture and deformation microstructure should not be considered separately, because both evolve together during rolling and later influence recovery and recrystallization during annealing (Hansen & Juul Jensen, 2011).

In rolled FCC metals, deformation texture is commonly described using orientation fibers, especially the α -fiber and β -fiber. For aluminum alloys, the β -fiber is particularly important because it contains the main rolling-texture components. This fiber extends from Brass $\{110\}\langle 112 \rangle$, through S $\{123\}\langle 634 \rangle$, to Copper $\{112\}\langle 111 \rangle$ (Hirsch & Lücke, 1988). These components are widely used to describe the deformation texture of rolled aluminum alloys and to follow the evolution of preferred orientations during rolling (Nyung Lee, 2005).

In FCC metals such as aluminum, slip at room temperature mainly occurs on $\{111\}\langle 110 \rangle$ systems. High-stacking-fault-energy FCC metals, including aluminum, commonly develop a pure-metal or copper-type rolling texture during cold rolling. This texture is mainly composed of Copper, Brass, and S components. In FCC aluminum sheets, these components are usually located along the β -fiber, which extends from Brass $\{110\}\langle 112 \rangle$, through S $\{123\}\langle 634 \rangle$, to Copper $\{112\}\langle 111 \rangle$ (Kestens & Pirgazi, 2016; Yuan, Wang, Zhang, Liu, & Gao, 2011).

The strength of deformation texture generally increases with rolling reduction because grains progressively rotate toward stable orientations during plastic deformation (Humphreys & Hatherly, 2004). However, deformation texture is not controlled only by the reduction level. Rolling geometry, friction, deformation temperature, initial grain size, shear banding, second-phase particles, and strain path can also affect the development of texture (Nyung Lee, 2005). Deformation temperature is especially important because many industrial rolling operations are performed under warm or hot conditions, where dynamic recovery and changes in deformation behavior can modify the final orientation distribution (Humphreys & Hatherly, 2004).

Since plastic anisotropy is strongly controlled by texture, understanding rolled sheets requires attention to the orientation distribution of grains. (J. Chen, Poole, & Parson, 2018; Thomesen, Hopperstad, & Børvik, 2021)

1.4.2 Recrystallization Texture

Recrystallization texture refers to the orientation distribution that develops during annealing when new, nearly strain-free grains nucleate and grow within the deformed matrix. It differs from deformation texture because it forms during thermal treatment rather than directly during plastic deformation (Traka, 2022). During primary static recrystallization, elongated and highly deformed grains are replaced by new grains with much lower dislocation density (Figure 1-10) (Lee, Jo, & Jeon, 2023). This replacement can significantly change the crystallographic orientation distribution of the material (Traka, 2022).

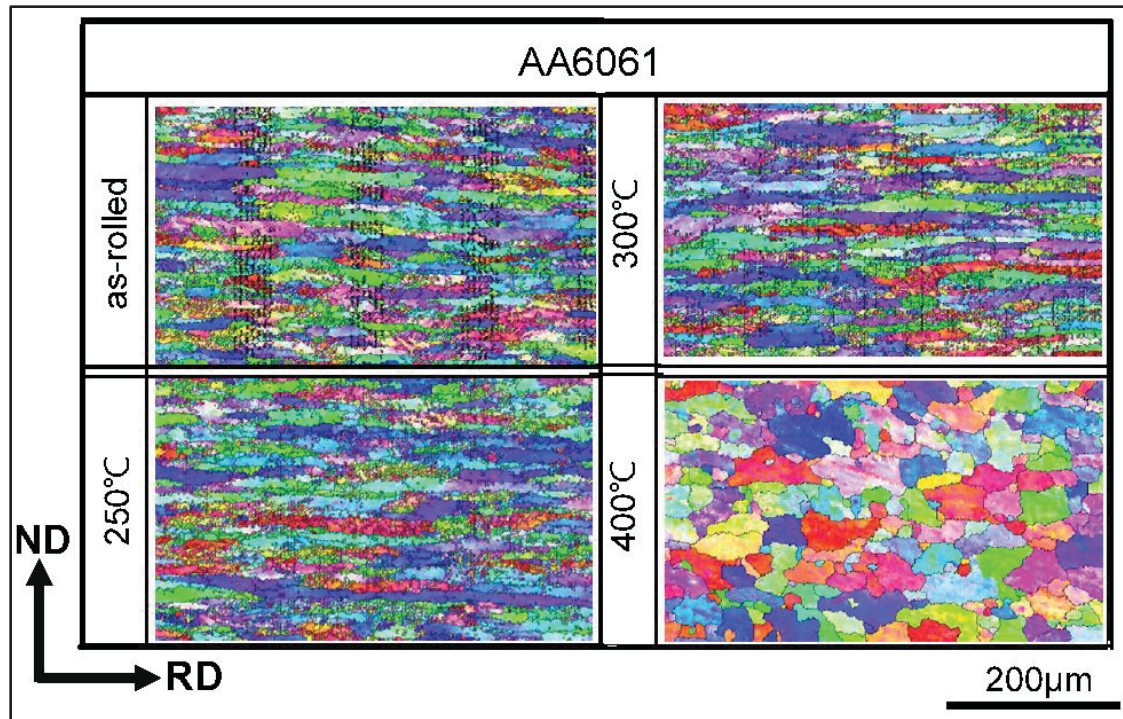


Figure 1-10 EBSD orientation maps showing the transition from elongated rolled grains to equiaxed recrystallized grains in AA6061 with increasing annealing temperature

Taken from Lee, Jo, & Jeon (2023)

The development of recrystallization texture is strongly affected by the prior deformed state. Since recrystallization is driven by stored deformation energy associated with dislocations and other defects, the final recrystallization texture depends on the prior deformation texture, local stored-energy distribution, and deformation substructure (Hansen & Juul Jensen, 2011). It is also controlled by the orientations of the nuclei that form, the availability of nucleation sites, orientation gradients, and the mobility of grain boundaries during annealing (Kestens & Pirgazi, 2016).

The formation of recrystallization texture is commonly explained by oriented nucleation and selective growth. Oriented nucleation refers to a recrystallization mechanism in which nuclei do not form with random orientations. Instead, some crystallographic orientations nucleate more easily or more frequently because they already exist in favorable regions of the deformed microstructure. In this mechanism, the final recrystallization texture is controlled mainly by

the orientations that become recrystallization nuclei during annealing (Hansen & Juul Jensen, 2011; Kestens & Pirgazi, 2016). For aluminum alloys, Cube-oriented regions are often discussed in relation to oriented nucleation. Cube-oriented substructures can recover strongly, develop high-angle boundaries, and reach a critical size more easily, allowing Cube-oriented nuclei to contribute to a strong Cube recrystallization texture (Mao & Yang, 2014).

Oriented growth refers to a recrystallization mechanism in which nuclei with different orientations may form, but some grow preferentially because they have a favorable crystallographic relationship with the surrounding deformed matrix. In FCC metals such as aluminum, this growth advantage is often associated with a misorientation of about 40° around a $\langle 111 \rangle$ axis between the recrystallized grain and the deformed matrix. Boundaries with this relationship can migrate more easily, allowing favorable orientations to grow faster and dominate the final recrystallization texture (Hansen & Juul Jensen, 2011; Mao & Yang, 2014).

Recrystallization components in rolled aluminum alloys can be described using several groups of orientations. Cube $\{001\}\langle 100 \rangle$ and R $\{124\}\langle 211 \rangle$ components are often associated with nucleation in relatively low-stored-energy deformation bands. Goss $\{011\}\langle 100 \rangle$, Q $\{013\}\langle 231 \rangle$, and P $\{011\}\langle 122 \rangle$ orientations may be associated with grains originating from deformation bands or shear regions. Recrystallization texture may also form through rolling-texture transformations involving $40^\circ\langle 111 \rangle$ rotations, while retained rolling-texture orientations can indicate insufficient driving force or pinning effects that restrict grain-boundary migration (Engler & Lücke, 1992).

Particle-stimulated nucleation (PSN) is another important mechanism affecting recrystallization texture in aluminum alloys. PSN occurs around coarse second-phase particles, typically larger than $1\text{ }\mu\text{m}$, that remain non-deformable during plastic deformation (Engler, Vatne, & Nes, 1996). During rolling, these particles create highly strained particle-deformation zones in the surrounding matrix, where stored energy and local orientation gradients are high (K. Adam & Field, 2023; Sidor, Decroos, Petrov, & Kestens, 2015). During subsequent annealing, subgrains within these zones can grow rapidly and act as preferred nucleation sites for recrystallized grains (Engler et al., 1996). Because the orientations in particle-deformation zones are usually complex and less ordered, PSN can weaken strong recrystallization textures,

such as Cube, and promote a more random overall texture in the final sheet (Kumar et al., 2020).

1.4.3 Plastic Anisotropy and r-value

Anisotropy means that a material property changes with direction. In rolled sheets, anisotropy can develop because rolling elongates grains and produces preferred crystallographic orientations. Texture and anisotropy are related, but they are not the same. Texture describes the orientation distribution of grains, whereas anisotropy describes the direction-dependent mechanical response that can result from this orientation distribution. Therefore, a strongly textured sheet may show different mechanical behavior along the rolling direction, transverse direction, and diagonal direction (Kestens & Pirgazi, 2016).

In the sheet forming, anisotropy is commonly described using normal anisotropy and planar anisotropy. Normal anisotropy represents the resistance of a sheet to thinning during forming, while planar anisotropy represents directional differences within the sheet plane. Good drawability is generally associated with high normal anisotropy and low planar anisotropy. Therefore, anisotropy is important for evaluating the formability of rolled sheets and depends on the combined effects of crystallographic texture, grain morphology, and processing history (Traka, 2022).

The r-value, or Lankford coefficient, is commonly used to describe plastic-flow anisotropy. It represents the resistance of a sheet to thinning during tensile deformation and is determined from the relationship between transverse strain and thickness strain. Changes in r-value with tensile direction indicate that plastic flow differs depending on the loading direction (Thomesen et al., 2021). The Δr value represents planar anisotropy and is calculated from the directional Lankford values measured at 0° , 45° , and 90° to the rolling direction. A lower Δr indicates lower in-plane anisotropy and better deep-drawing behavior (Sidor, 2021).

1.5 Cold Rolling and Deformation-texture Evolution

1.5.1 General Features of Cold Rolling

Cold rolling changes both the grain structure and crystallographic texture of aluminum sheets. During cold rolling, grains become elongated along the rolling direction, dislocation density increases, and stored deformation energy accumulates in the microstructure (Traka, 2022). This effect is illustrated in Figure 1-11, where EBSD orientation maps of 1100 aluminum show the transition from a coarse grain structure in the annealed condition to a highly elongated deformation structure after cold rolling (Williams, Chen, Ramesh, & Dandekar, 2013). These changes are accompanied by grain rotation toward deformation-texture components, especially as rolling reduction increases (Kestens & Pirgazi, 2016). Therefore, the texture formed during cold rolling is closely linked to the deformation structure produced during rolling, including orientation gradients, dislocation boundaries, and local stored-energy differences (Hansen & Juul Jensen, 2011).

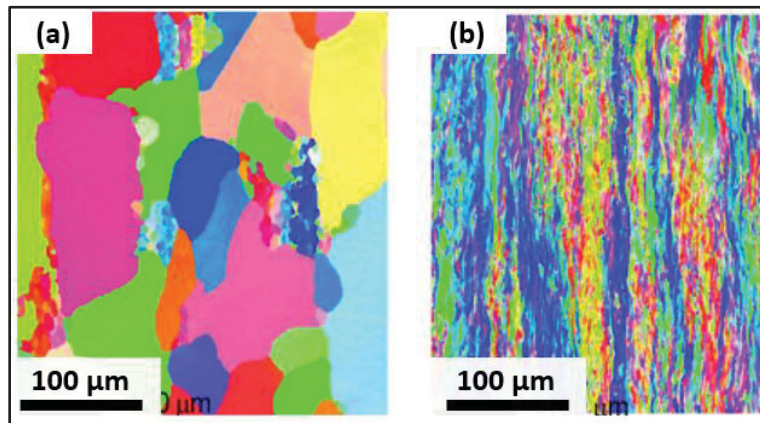


Figure 1-11 EBSD orientation maps of 1100 aluminum showing (a) the annealed condition and (b) the cold-rolled condition

Taken from Williams et al. (2013)

Cold-rolling texture should be understood as an evolving orientation distribution rather than as a fixed set of components (Hirsch & Lücke, 1988). Hirsch and Lücke (1988) showed that

rolling texture develops as grain orientations rotate toward stable or metastable orientations during deformation. Their simulations indicated that the final rolling texture depends on active slip systems, deformation constraints, strain level, shear relaxation, and initial texture. At low strain, orientations tend to concentrate along main orientation fibers. With increasing strain, they evolve toward typical FCC rolling-texture components. In high-stacking-fault-energy metals such as aluminum, the rolling texture is mainly copper-type and is commonly described by the β -fiber. In lower-stacking-fault-energy materials, the texture tends to shift more toward brass-type components (Hirsch & Lücke, 1988).

The development of cold-rolling texture is also influenced by slip activity and crystallographic stability (Dillamore & Roberts, 1964). Dillamore and Roberts (1964) showed that FCC rolling texture evolves through slip-induced crystal rotation and depends strongly on cross-slip activity. They emphasized that cross-slip affects whether the rolling texture develops as an alloy-type texture or as a pure-metal, also called copper-type, texture. When cross-slip is limited, the texture tends to remain near the alloy-type orientation, mainly $\{110\}\langle 112 \rangle$. When cross-slip is extensive, orientations rotate toward the pure-metal or copper-type texture and may approach $\{112\}\langle 111 \rangle$ at very high reductions. Therefore, rolling texture is not fixed at one component; it changes with strain, stacking-fault energy, material type, and cross-slip activity (Dillamore & Roberts, 1964).

1.5.2 Effect of Rolling Reduction and Alloy Composition

The intensity of deformation texture generally increases with rolling reduction (Sidor et al., 2015). Sidor et al. (2015) showed that cold rolling transformed an initially near-random as-cast texture into a strong rolling texture in an Al–Mg–Mn–Si alloy. The rolling texture contained typical plane-strain-compression components, including α -fiber and β -fiber orientations. As rolling reduction increased from 74.2% to 99.1%, the deformation texture became stronger, and the texture index increased almost linearly with strain (Sidor et al., 2015).

Liu and Morris (2002) also showed that cold rolling gradually transformed initial recrystallization textures into a β -fiber rolling texture in continuous-cast AA5xxx aluminum

alloys, including AA5005, AA5052, AA5754, AA5182, and high-Fe AA5754. The hot bands were first annealed to develop recrystallization textures and then cold rolled to reductions up to 95%. The β -fiber volume fraction increased with true rolling strain in all alloys, showing that cold rolling progressively rotates initial orientations toward stable deformation-texture components. The authors quantified this evolution by integrating ODF intensities within 15.5° of the β -fiber centerline and showed that the increase in β -fiber volume fraction followed an Avrami-type relationship. They also reported that alloy composition affected the kinetics of β -fiber formation: higher Mg increased the kinetic constant and reduced the strain exponent, while higher Mn had the opposite effect (W. C. Liu & Morris, 2002a).

Similar trends were reported in other aluminum alloys. Papadopoulou et al. (2020) showed that increasing cold reduction from 52.5% to 92% in AA5182 strengthened the rolling texture and changed the relative balance of the main components (Sofia Papadopoulou, Kontopoulou, Gavalas, & Papaefthymiou, 2020). Chen et al. (2024) also showed that increasing cold-rolling reduction from 41% to 82% strengthened the β -fiber texture in an Al–Mg–Si–Zn alloy sheet and changed the dislocation structure from dislocation walls and microbands to lamellar bands (K.-X. Chen et al., 2024). Kumar et al. (2020) reported a similar response in AA3003, where hot rolling followed by cold rolling produced elongated and banded grains along the rolling direction. Cold rolling also strengthened typical FCC deformation-texture components, including Brass, Copper, S, and Goss, while the Cube component decreased (Kumar et al., 2020).

1.5.3 Effect of Initial Texture, Strain Path, and Rolling Route

The initial texture can strongly affect the final rolling texture because many observed rolling textures are transient rather than fully stable states (Hirsch & Lücke, 1988). Hirsch and Lücke (1988) showed that a strong S component after cold rolling may reflect the deformation path, starting texture, and orientation movement along the β -fiber, rather than only a final stable texture state (Hirsch & Lücke, 1988). Similarly, Mao and Yang (2014) showed that the initial texture before rolling affected the rolling texture after 95% cold reduction. An initial Cube

texture promoted a stronger S-type rolling texture, while an initial rotated Cube texture promoted a stronger Copper-type rolling texture. This shows that deformation texture after heavy rolling can depend strongly on the starting orientation state (Mao & Yang, 2014).

Rolling route and strain path are also important processing variables because they determine how deformation is imposed on the sheet during rolling (Gupta et al., 2019). In unidirectional rolling, the sheet is repeatedly rolled along the same direction, so deformation accumulates mainly along one fixed rolling direction. In cross-rolling, the sheet is rotated between passes, commonly by 90° around the normal direction. This rotation changes the relationship between the new rolling direction and the deformation structure produced in the previous pass (Nayan et al., 2019). As a result, the material does not simply continue deforming under the same condition. Instead, the previously formed substructure, grain orientation, and local hardening state are reloaded under a new strain path. This can affect grain rotation, dislocation arrangement, shear-band formation, and the final crystallographic texture (Chrominski & Lewandowska, 2022; Huh, Engler, & Raabe, 1995).

The effect of strain path is especially important in rolled aluminum alloys because texture develops through progressive grain rotation during plastic deformation. Under a constant rolling direction, grains tend to rotate toward stable rolling-texture orientations as reduction increases. In FCC aluminum alloys, this usually promotes the development of β -fiber components, including Copper, S, and Brass orientations (Yuan et al., 2011). When the rolling direction is changed, the existing texture is reoriented with respect to the new sample reference frame. This can temporarily destabilize previously developed rolling-texture components and change the rate at which the material returns toward stable deformation orientations (Ma, Wang, Zhang, & Liu, 2014). Therefore, the final texture depends not only on the amount of reduction, but also on the sequence and direction of deformation applied during rolling (Chrominski & Lewandowska, 2022).

The imposed deformation route can also affect cold-rolling texture (Miszczyk, Paul, Driver, & Poplewska, 2017). Miszczyk et al. (2017) showed that plane strain compression produced a typical β -fiber rolling texture in AA1050, while equal channel angular extrusion produced a shear-type texture close to $\{124\}<561>$. This confirms that deformation texture depends not

only on strain level, but also on the imposed strain path. The corresponding differences in grain morphology and local crystallographic orientation are illustrated in Figure 1-12 (Miszczyk et al., 2017). Similarly, the rolling route can strongly affect texture development. In unidirectional rolling, typical rolling-texture components generally strengthen with increasing reduction, whereas cross-rolling can destabilize the existing texture by changing the rolling direction between passes. It was reported that for AA6082, repeated 90° changes in rolling direction disturbed the usual β -fiber development and produced a texture closer to α -fiber-related orientations with $\langle 110 \rangle // \text{ND}$ (Chrominski & Lewandowska, 2022).

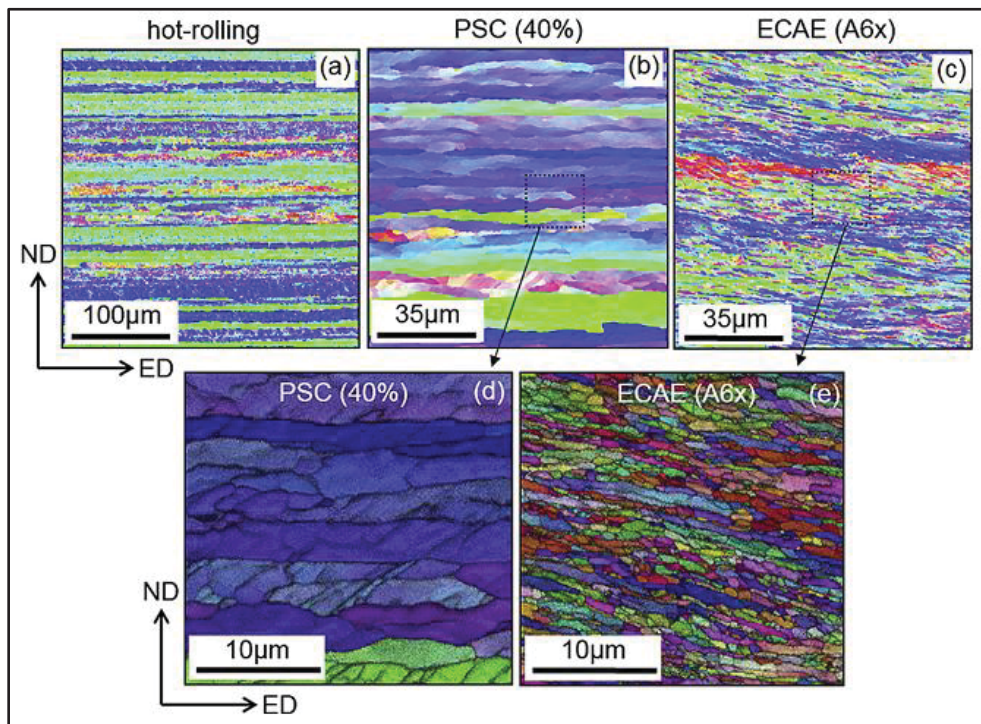


Figure 1-12 Microstructures and EBSD orientation maps of AA1050 in the initial hot-rolled condition, after 40% plane-strain compression (PSC), and after six passes of equal-channel angular extrusion (ECAE) through route A, illustrating the influence of the imposed deformation route and strain path on grain morphology and crystallographic orientation

Taken from Miszczyk et al. (2017)

Several studies have shown that changing the rolling route affects deformation texture before annealing. Ma et al. (2014) reported that unidirectional rolling of AA3105 strengthened the β -fiber texture with increasing reduction, including Copper, S, and Brass components. During

cross-rolling, the original β -fiber orientations became unstable in the new sample reference frame and gradually rotated back toward the β -fiber as reduction increased. At high cross-rolling reductions, strong Brass and S components developed, while the Copper component remained weak. This shows that changing the rolling direction can modify the relative stability of deformation-texture components and alter the balance between Copper-, S-, and Brass-type orientations (Ma et al., 2014).

The influence of rolling route is also related to the initial grain structure and initial texture. Yuan et al. (2011) studied AA5754 and high-Fe AA5754 aluminum alloys and showed that the rolling direction relative to elongated recrystallized grains affected the rate of deformation-texture formation. Unidirectional rolling caused a faster decrease in Cube and rotated Cube components and a slightly faster development of β -fiber texture than cross-rolling. This result indicates that the starting microstructure should not be treated as a passive condition. Instead, the initial grain shape and initial orientation distribution can influence how grains rotate during rolling and how quickly stable rolling-texture components develop (Yuan et al., 2011).

For AA6082, Chrominski and Lewandowska (2022) provided a more directly relevant example by studying the combined effect of initial fiber orientation and rolling path on microstructure and texture evolution. Their material initially contained strong $\langle 001 \rangle$ and $\langle 111 \rangle$ fiber textures parallel to the extrusion direction. When rolling was applied parallel to the initial fiber axis, the fibrous morphology was mostly preserved, and the material rotated toward typical rolling-texture orientations, including Copper and Goss components. When rolling was applied perpendicular to the initial fiber axis, the original fiber structure was more strongly disrupted, and grains rotated mainly toward the Copper component. This comparison shows that, even under unidirectional rolling, the response of the material depends strongly on the orientation of the initial microstructure relative to the imposed rolling direction (Chrominski & Lewandowska, 2022).

In the cross-rolled AA6082 condition, the deformation path became more complex because the rolling direction changed by 90° after each pass. This repeated change disturbed the usual development of the β -fiber texture. Instead of forming only typical rolling-texture components, the material developed a texture closer to α -fiber-related orientations, with $\langle 110 \rangle$ //ND. The

initial fibers were also broken more severely under cross-rolling than under unidirectional rolling. Therefore, the initial texture did not simply disappear during rolling. It affected the deformation path, the breakage of the fibrous structure, the formation of substructure, and the final deformation texture of AA6082 (Chrominski & Lewandowska, 2022).

Cross-rolling can also affect the deformation structure by promoting less uniform slip and shear-band formation. Huh et al. (1995) showed that changing the rolling direction modified both deformation structure and texture in low-carbon steel sheets. In straight-rolled samples, deformation occurred mainly through microband formation and elongation, with only limited shear banding. In cross-rolled samples, shear bands became more frequent after about 60–70% reduction. This was attributed to the rotation of the pre-existing microband structure after the rolling direction was changed. The rotated hardened structure made further slip less uniform and promoted shear-band formation (Huh et al., 1995).

The rolling route can also influence the texture that develops after annealing or heat treatment because recrystallization starts from the deformed state produced during rolling. Nayan et al. (2019) showed that unidirectional cold rolling of AA2195 produced a typical Copper-type rolling texture with Copper, S, and Brass components, while two-step cross-rolling destabilized the Copper and S components and promoted a stronger Brass texture. After heat treatment, the unidirectionally rolled sample developed a strong Cube recrystallization texture, whereas the cross-rolled sample showed a weaker annealing texture without a strong Cube component. This indicates that strain path affects not only deformation texture, but also the later recrystallization texture through the orientation distribution and stored-energy state created during rolling (Nayan et al., 2019).

1.5.4 Local Heterogeneity, and Grain Interaction

Through-thickness heterogeneity is another important feature of cold-rolled sheets. Large-strain deformation can create gradients in grain size, grain shape, and crystallographic texture through the sheet thickness (J. Chen et al., 2018; Thomesen et al., 2021). Engler et al. (2001) showed that cold rolling of commercial-purity AA1000A produced a typical rolling texture in

the sheet center, while warm rolling generated shear-texture components near the surface because of friction-induced shear. This shows that the center and surface regions of rolled sheets may not follow the same texture evolution path. The center is mainly controlled by plane-strain compression, while the surface can be affected by roll-sheet friction and local shear deformation (Engler, Kim, & Huh, 2001).

Chen et al. (2018) showed clear through-thickness microstructural and textural heterogeneity in extruded aluminium strips. In the unrecrystallized condition, the near-surface region contained less elongated grains and a shear-related texture, while the interior exhibited elongated grains and a plane-strain deformation texture. After recrystallization, a peripheral coarse-grain layer with a rotated Cube texture, approximately $\{001\}\langle 110 \rangle$, formed near the surface, whereas the interior contained more equiaxed grains and a Cube- and Goss-dominated texture. These surface-to-centre variations are compared in Figure 1-13.

Local grain interactions can further modify the deformation structure during rolling. Cantergiani et al. (2022) used crystal plasticity simulations to study the stability and fragmentation of Cube-oriented grains during rolling. They showed that Cube fragmentation depended strongly on the surrounding crystallographic environment. During cold rolling, Brass and S neighbors caused the strongest Cube fragmentation, in-grain orientation gradients, and subdomain formation. Copper produced only limited Cube misorientation, while Goss had almost no effect. This means that Cube evolution during rolling is not controlled only by the Cube orientation itself. It also depends on the orientations of the surrounding grains and their local interaction during deformation. Therefore, deformation texture should not be treated only as a list of texture components; it also affects local grain interaction, Cube splitting, and orientation-gradient formation (Cantergiani, Falkinger, & Roters, 2022).

The deformation structure formed during cold rolling affects the later annealing response. Hansen and Juul Jensen (2011) emphasized that recrystallization occurs within a heterogeneous deformed microstructure, where nuclei form and grow in regions with different local orientation gradients and stored-energy levels. Therefore, the cold-rolled texture and deformation substructure provide the starting condition for recovery, recrystallization, and recrystallization texture during post-rolling heat treatment (Hansen & Juul Jensen, 2011). This

confirms that cold rolling changes grain morphology and texture together, and that the deformed structure created during rolling provides the starting condition for later annealing and recrystallization.

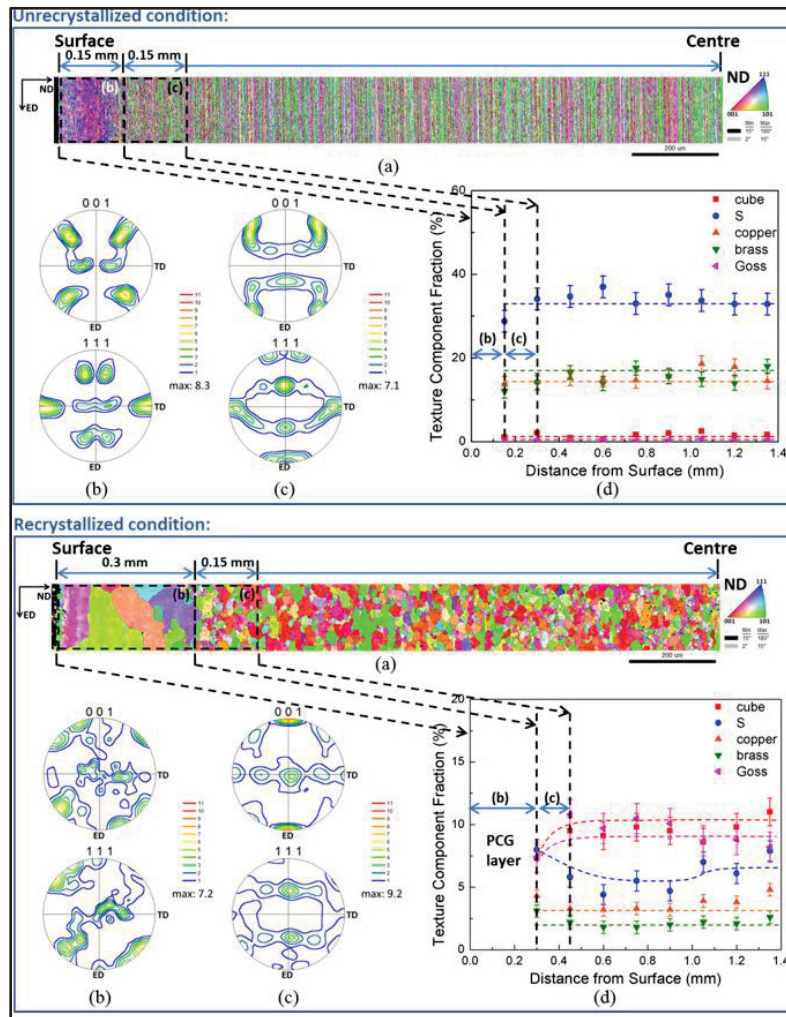


Figure 1-13 Through-thickness microstructure and texture variations in unrecrystallized and recrystallized extruded aluminium strips

Taken from J. Chen et al. (2018)

1.6 Hot Rolling and Dynamic Restoration

Hot deformation refers to plastic deformation at elevated temperature, where deformation occurs together with thermally activated restoration mechanisms such as dynamic recovery and dynamic recrystallization (Lypchanskyi et al., 2024). Unlike cold deformation, where stored energy mainly accumulates through dislocation multiplication, hot deformation allows part of this stored energy to be reduced during deformation through dynamic softening mechanisms (H. Lin, 2021). Therefore, the final hot-deformed microstructure is controlled by the balance between work hardening and dynamic softening. This balance depends strongly on deformation temperature, strain rate, strain level, and the initial microstructural condition (H. Lin, 2021; Lypchanskyi et al., 2024).

During hot rolling, plastic deformation occurs together with thermally activated microstructural changes such as recovery and recrystallization. These processes affect both grain structure and crystallographic texture. Because work hardening is reduced at hot-rolling temperatures, large thickness reductions can be achieved more easily than in cold rolling. However, deformation, recovery, and recrystallization may occur simultaneously, making the final microstructure strongly dependent on rolling temperature, strain rate, and reduction level (Kumar et al., 2020).

Hot rolling can also produce non-uniform microstructure and texture through the sheet thickness because deformation and thermal conditions are not uniform during processing (Aryshenskii et al., 2018). Aryshenskii et al. (2018) showed that stress state, strain intensity, and strain-rate distribution vary within the deformation zone during thermomechanical treatment of 8011 aluminum alloy. By comparing laboratory rolling, industrial rolling, and plane-strain testing, they found that plane-strain and industrial rolling produced inhomogeneous structure and texture. This indicates that the hot-rolling response is controlled not only by nominal rolling reduction, but also by local deformation conditions during processing (Aryshenskii et al., 2018).

The texture developed during hot rolling can differ from that formed during cold rolling. In rolled aluminum alloys, β -fiber components such as Cu, S, and Bs are commonly used to describe rolling texture. Cold rolling may produce comparable intensities of Cu, S, and Bs orientations, whereas Bs texture can become more dominant during hot rolling because of changes in slip activity and stored-energy release (Li, Zhao, Liu, & Li, 2018a). Therefore, controlling hot-rolling temperature and deformation conditions is important for managing both the grain structure and texture evolution. Second-phase particles add further complexity because their size, shape, and distribution can affect recrystallization behavior and final formability (Kumar et al., 2020).

In AA6082, Lypchanskyi et al. (2024) showed that the hot-deformation response changes significantly with temperature. They performed tensile tests at 200, 300, and 400 °C using strain rates from 0.01 to 10 s⁻¹. At 200 °C, deformation was mainly controlled by work hardening and intragranular slip. At 300 °C, the response shifted toward a transition between slip-dominated deformation and dynamic recrystallization. At 400 °C, dynamic softening became dominant, and the microstructure showed clear evidence of continuous dynamic recrystallization. This indicates that increasing temperature promotes restoration during deformation and changes the dominant deformation mechanism from work-hardening-controlled slip toward recovery- and recrystallization-assisted softening (Lypchanskyi et al., 2024).

Strain rate also affects the hot-deformation mechanism because it controls the time available for thermally activated processes. At lower strain rates, deformation proceeds more slowly, so dislocations have more time to rearrange, annihilate, and form subgrain structures. This promotes dynamic recovery and can support the gradual transformation of low-angle boundaries into higher-angle boundaries. At higher strain rates, deformation occurs more rapidly, so slip-related features and work-hardening effects remain more visible, while recovery and recrystallization are less developed (Lypchanskyi et al., 2024).

Therefore, at lower temperatures or higher strain rates, deformation is more strongly controlled by slip and work hardening. In contrast, at higher temperatures or lower strain rates, dynamic restoration becomes more active and can modify the dislocation structure, boundary

misorientation, grain morphology, and texture development. Thus, the same deformation temperature can produce different microstructures depending on the applied strain rate (Lypchanskyi et al., 2024).

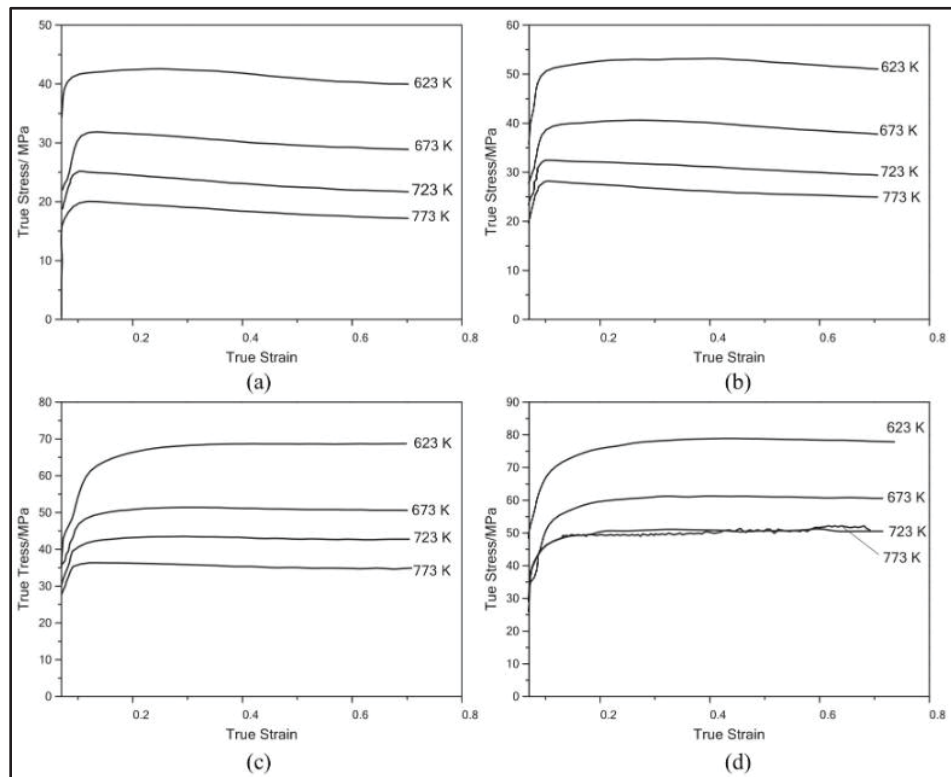


Figure 1-14 True stress–strain curves of hot-compressed AA6082, showing reduced flow stress at higher temperatures and lower strain rates

Taken from Lin (2021)

Lin (2021) also studied dynamic recrystallization in as-cast AA6082 during hot compression. The specimens were compressed by 70% at 350–500 °C and strain rates of 0.01–5 s⁻¹, followed by immediate quenching. The results showed that flow stress decreased with increasing temperature and decreasing strain rate, indicating stronger dynamic softening under lower Zener–Hollomon parameter conditions (Figure 1-14). Although the stress–strain curves did not show a clear peak typical of discontinuous dynamic recrystallization, the microstructural observations showed that continuous dynamic recrystallization was the dominant mechanism in hot-deformed AA6082. During deformation, dislocations rearranged into cell walls and

subgrain boundaries. With increasing strain, low-angle boundaries became progressively more misoriented and gradually transformed into high-angle boundaries, leading to the formation of new recrystallized grains (H. Lin, 2021).

The initial microstructure can also influence the hot-deformation response. Liao et al. (2019) studied the hot deformation behavior and recrystallization mechanisms of AA7055 aluminum alloy in two starting conditions: homogenized and pre-rolled T3. Both materials were compressed to 60% height reduction at 370–450 °C and strain rates of 0.01–10 s⁻¹. Their results showed that the initial condition strongly affected flow stress and recrystallization during hot deformation. The pre-rolled T3 material showed higher flow stress than the homogenized material, but it also produced a higher recrystallized fraction after deformation. This was attributed to its smaller initial grains, finer particles, and more pre-existing subgrains, which increased the driving force for recrystallization (B. Liao et al., 2019).

Liao et al. (2019) also reported that DDRX and CDRX occurred in both starting conditions, while GDRX appeared only in the homogenized material and MDRX only in the pre-rolled T3 material. Processing maps were used to identify suitable hot-working windows. The optimum domain was 405–450 °C at 0.01–0.08 s⁻¹ for the homogenized material and 430–450 °C at 0.01–0.1 s⁻¹ for the pre-rolled T3 material. In unstable domains, localized deformation and shear bands were more likely, while the optimum domains promoted more stable deformation and favored CDRX. Although this study was conducted on AA7055 rather than AA6082, it supports the broader point that hot deformation is controlled not only by temperature and strain rate, but also by the starting microstructure (B. Liao et al., 2019).

1.7 Post-rolling Annealing and Recrystallization Texture Evolution

1.7.1 General Role of Post-rolling Annealing

Post-rolling heat treatment and annealing are used to modify the microstructure produced during rolling. After rolling, the material can contain elongated grains, deformation texture, stored deformation energy, and a high density of defects. During annealing, these deformation-

induced features can be reduced or reorganized through recovery and recrystallization (P. Du et al., 2024; Imran & Walther, 2016). Therefore, post-rolling annealing does not only soften rolled aluminum sheets. It also affects recrystallized grain morphology, recrystallization texture, and anisotropy.

Several empirical trends describe recrystallization behavior after deformation. First, a critical minimum level of prior deformation is required before recrystallization can initiate. As the degree of prior deformation increases, the recrystallization temperature generally decreases because of the higher stored energy available to drive the transformation. Finer initial grains can also reduce the recrystallization temperature, whereas larger grains usually require more deformation to recrystallize under similar annealing conditions. The final recrystallized grain size is usually smaller when high prior deformation is combined with lower annealing temperature. In contrast, deformation at higher temperatures can leave less stored energy in the material and may increase the subsequent recrystallization temperature (Imran & Walther, 2016). Therefore, the annealing response depends on both the deformation history and the thermal path.

Du et al. (2024) showed that recrystallization behavior in rolled aluminum sheets depends strongly on the prior rolling route, rolling reduction, and annealing temperature. They studied continuous-cast AA3003 Al–Mn alloy by cold rolling hot-band sheets to 40–90% reduction using unidirectional rolling and pseudo cross-rolling, followed by annealing at 399 °C and 427 °C. At 399 °C, the recrystallized grains became coarser and more elongated with increasing reduction, whereas at 427 °C, the recrystallized grain size decreased as reduction increased. Cross-rolling also promoted recrystallization more effectively than straight rolling under similar annealing conditions (P. Du et al., 2024). Therefore, post-rolling annealing does not only soften rolled aluminum sheets; it also changes the recrystallized grain morphology depending on the prior rolling condition and annealing temperature.

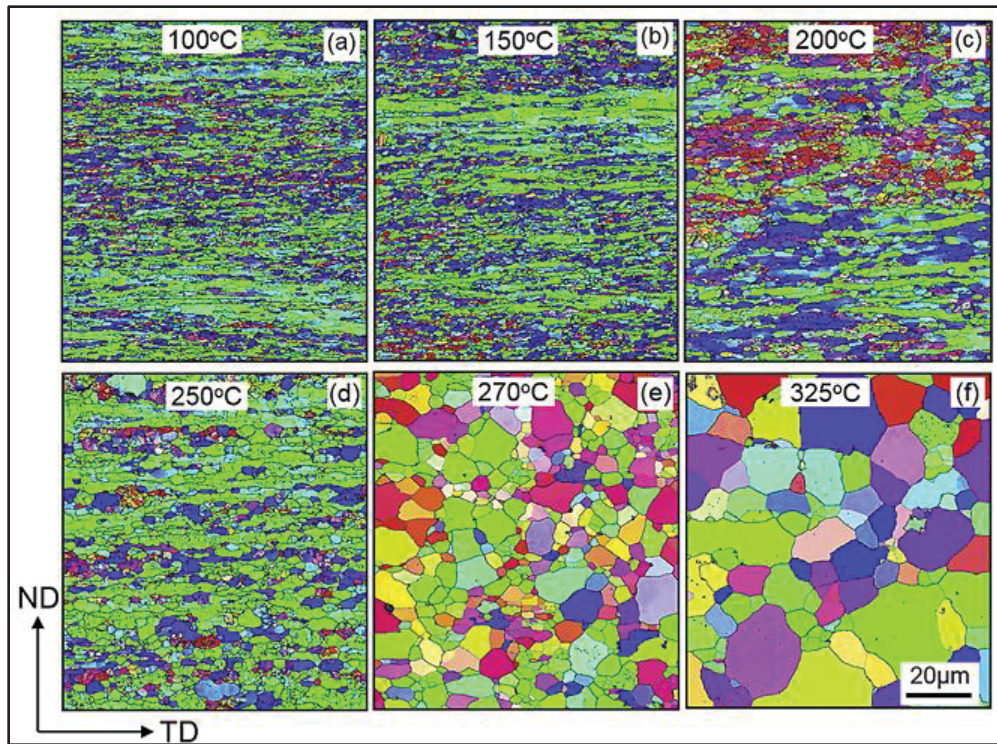


Figure 1-15 EBSD orientation maps showing the microstructural evolution of ECAE-deformed AA1050 during annealing, from an elongated deformed structure to equiaxed recrystallized grains

Taken from Miszczyk et al. (2017)

The deformation route also affects recrystallization kinetics during annealing. Miszczyk et al. (2017) compared AA1050 samples deformed by plane strain compression and equal channel angular extrusion. The plane strain compression sample was deformed to 40% reduction, corresponding to a logarithmic strain of 0.51. The equal channel angular extrusion sample was processed through six passes using route A, where the billet orientation was kept unchanged between consecutive passes. The deformed samples were then annealed for 1 h between 100 °C and 450 °C. The PSC samples reached about 50% recrystallization near 300 °C and were almost fully recrystallized near 350 °C. In contrast, the ECAE samples recrystallized at lower temperatures, reaching about 50% recrystallization at 250 °C and complete recrystallization at 300 °C. This difference was attributed to the larger imposed strain in the ECAE samples (Miszczyk et al., 2017). The progressive transformation of the deformed microstructure into

equiaxed recrystallized grains with increasing annealing temperature is illustrated in Figure 1-15.

1.7.2 Annealing Path, Recovery Condition, and Recrystallization Response

The annealing path can control the final recrystallized grain structure and texture. Huang et al. (2017) compared single-step annealing and two-step annealing in cold-rolled twin-roll-cast 3003 Al–Mn alloy sheet. The sheet was cold rolled from 6.4 mm to about 4.5 mm, corresponding to 28.6% reduction, while the total deformation from casting and cold rolling was considered to be about 50%. They compared single-step annealing at 530 °C for 15 h with two-step annealing, consisting of pre-recovery at 450 °C for 5 h followed by recrystallization annealing at 530 °C for 15 h (L. Huang et al., 2017).

The two annealing paths produced different grain structures and textures. The single-step treatment produced coarse elongated grains with strong Cube texture, while the two-step treatment produced finer quasi-equiaxed grains with weaker ND-rotated Cube texture. The two-step treatment reduced the average grain size from about 152.4 μm to 35.2 μm and decreased the grain aspect ratio from 5.23 to 2.03. The partially recrystallized two-step sample retained a strong Brass deformation texture, but the final texture was strongly modified by the annealing path. In the single-step sample, Cube-oriented subgrains likely had a growth advantage, producing a strong Cube texture. In the two-step sample, many randomly oriented nuclei formed inside precipitate-free zones, leading to a much weaker final texture. This improvement was attributed to precipitate-free zones formed during pre-recovery, which reduced particle pinning and allowed more recrystallization nuclei to form during the second annealing step. This increased nucleation, refined the grain structure, and weakened the recrystallization texture (L. Huang et al., 2017).

The annealing path can also affect mechanical anisotropy through its effect on recrystallization texture. Huang et al. (2017) reported that the single-step annealed sheet showed clear directional differences in r -values, while the two-step annealed sheet showed more uniform r -values and a much lower planar anisotropy value. This indicates that the final recrystallization

texture and anisotropy are controlled not only by prior deformation, but also by the recovery and annealing path before recrystallization (L. Huang et al., 2017).

Mao and Yang (2014) showed that the annealing path can strongly affect the final recrystallization texture in 95% cold-rolled aluminum sheets with different purities and initial textures. Recovery pretreatment at 240 °C for 600 s reduced stored energy and slowed recrystallization. In contrast, direct annealing at 300 °C caused faster recrystallization and produced a more randomized texture (Mao & Yang, 2014).

Poorganji et al. (2010) studied the effect of cold reduction and non-isothermal annealing on recrystallization behavior and texture evolution in AA6451 aluminum alloy. The samples were annealed non-isothermally to 380 °C, held for 20 min, and furnace cooled. Their results showed that sufficient cold reduction was needed to obtain a fully recrystallized fine-grained structure with weak texture. The 44% cold-rolled sample remained partially recrystallized and retained a strong Cube texture after annealing, while the 64% and 80% cold-rolled samples became fully recrystallized and developed weaker textures. This behavior was related to the higher stored energy produced by larger cold reductions, which promoted full recrystallization during annealing (Poorganji, Sepehrband, Jin, & Esmacili, 2010).

Poorganji et al. (2010) also showed that restoration occurred in two main stages. At lower annealing temperatures, extended recovery occurred in some favorably oriented grains, where Cube-oriented bands could form and grow. At higher temperatures, discontinuous recrystallization produced new randomly oriented grains, leading to a weaker final texture. This behavior was affected by the interaction between precipitates, recovery, recrystallization, and grain orientation. Fine precipitates delayed recovery by restricting dislocation rearrangement, while precipitate coarsening at higher temperatures reduced pinning and allowed recrystallized grains to nucleate and grow (Poorganji et al., 2010).

1.7.3 Deformation Texture, and Recrystallization Texture

After rolling and annealing, recrystallization can weaken the deformation texture and promote new orientation components depending on the prior deformation state, stored-energy distribution, and annealing condition (Sofia Papadopoulos et al., 2020). Papadopoulos et al. (2020) showed this behavior in AA5182 aluminum alloy. After annealing at 350 °C for 1 h, the overall texture intensity decreased and the random texture fraction increased. Recrystallization-related components such as Q, R, and Goss also became more pronounced, although some deformation-texture components remained after annealing (Sofia Papadopoulos et al., 2020).

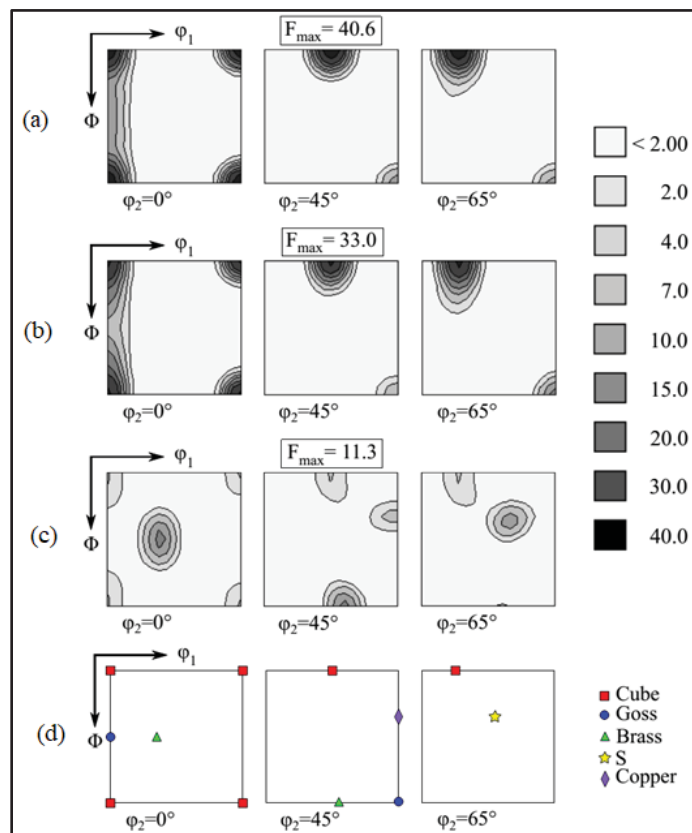


Figure 1-16 ODFs for the alloys (a) AA6061, (b) AA6063 and (c) AA6110, and (d) showing the location of the individual texture components

Taken from Thomesen et al. (2021)

The recrystallization response also depends on the alloy condition and the microstructure present before annealing (Thomesen et al., 2021). Thomesen et al. (2021) compared texture evolution in AA6061, AA6063, and AA6110 aluminum alloys. AA6061 and AA6063 exhibited typical recrystallization textures, characterized mainly by a strong Cube component and a minor Goss component. In contrast, AA6110 showed Cube-related orientations along the β -fiber, which was associated with a non-recrystallized extruded microstructure. The variation in recrystallization-related texture orientations, including their location and intensity in AA6061, AA6063, and AA6110, is shown in Figure 1-16, as reported by Thomesen et al. (2021). This comparison indicates that the final texture depends not only on the alloy family, but also on whether the material has fully recrystallized or still retains features of the prior deformation structure (Thomesen et al., 2021).

The recrystallization texture formed during annealing is strongly connected to the deformation texture produced before annealing. The deformed state contains variations in dislocation density, boundary spacing, misorientation, stored energy, and crystallographic orientation, and these local differences affect where recrystallized grains nucleate and how they grow. Since many nuclei have orientations already present in the deformed state, the final recrystallization texture is closely related to the prior deformation texture and stored-energy distribution (Hansen & Juul Jensen, 2011). Recrystallization texture should therefore be interpreted together with the deformation structure that existed before annealing.

Recrystallization also changes the balance between deformation-texture and recrystallization-texture components. In FCC metals, rolling textures can transform into specific recrystallization textures during annealing, but the final texture remains sensitive to alloy composition and processing history. Even when the initial deformation texture is similar, small changes in chemistry or thermomechanical processing can modify the final recrystallization texture (Imran & Walther, 2016). Obermayer et al. (2025) showed this transition in cold-rolled AA1050 sheets, where Brass, S, and Copper components decreased during static recrystallization, while Cube and inverse Goss components increased (Obermayer, Krempaszky, & Werner, 2025). This shows that annealing changes the texture state, but the transformation path still depends on the prior rolled structure.

The connection between deformation texture and recrystallization texture is commonly discussed through oriented nucleation and oriented growth. In oriented nucleation, some orientations form recrystallization nuclei more frequently during the early stage of recrystallization. In oriented growth, nuclei may form with different orientations, but grains with favorable orientation relationships to the surrounding deformed matrix grow more rapidly and become dominant in the final texture. In aluminum alloys, oriented growth is often associated with favorable orientation relationships between recrystallized grains and deformation-texture components, especially relationships involving rotations around the $\langle 111 \rangle$ axis (Engler et al., 1996).

The relative importance of oriented nucleation and oriented growth depends on the deformation state and annealing condition. Mao and Yang (2014) showed that oriented nucleation becomes more visible when the effective driving force is moderate and recovered subgrains or cell blocks can act as preferred nuclei. When the driving force is very high, recrystallization may occur rapidly, a wider range of orientations can nucleate, and the final texture can become more random. In contrast, oriented growth requires enough time and space for grains with favorable orientation relationships to grow preferentially (Mao & Yang, 2014).

Sebald and Gottstein (2002) also showed that nucleation conditions can affect subsequent texture evolution during grain growth. Their statistical model considered random nucleation, nucleation at shear bands, and nucleation from pre-existing nuclei. When the misorientation distribution between nuclei and the deformed matrix was narrow, stored-energy distribution became the dominant factor controlling textural evolution during growth. When the misorientation distribution was wider, grain-boundary mobility became more important. This shows that recrystallization texture is affected not only by which nuclei appear, but also by the local orientation and stored-energy environment into which they grow (Sebald & Gottstein, 2002).

1.7.4 Cube Texture, and Rolling-route Effects

Cube texture in recrystallized aluminum sheets is often discussed in relation to the surrounding deformation matrix. Engler et al. (1996) showed that Cube-oriented grains can have a growth advantage when they grow into favorable deformed orientations, while Cube-band nucleation can also contribute to Cube recrystallization texture (Engler et al., 1996). Lee (2005) suggested that Copper-oriented regions may also contribute to Cube recrystallization because they can contain higher stored energy and tend to disappear rapidly when Cube increases during annealing (Nyung Lee, 2005). This indicates that Cube recrystallization is not controlled only by the presence of Cube nuclei; it also depends on the stored-energy distribution and orientation relationship between the recrystallized grains and the surrounding deformed matrix.

Miszczyk et al. (2017) showed this relationship clearly in AA1050 aluminum alloy. In plane-strain-compressed samples, a strong Cube recrystallization texture developed during annealing, especially at 300 °C and above. Cube grains formed mainly inside or near S-oriented deformed regions, and the reduction of the S component was associated with Cube growth. In contrast, equal-channel-angular-extruded samples retained weak Cube orientations over a wide temperature range, but these orientations did not develop into a strong Cube texture during annealing (Miszczyk et al., 2017). This shows that pre-existing Cube fragments alone are not sufficient to produce a strong Cube recrystallization texture; the surrounding deformation matrix is also important.

Cantergiani et al. (2022) showed that Cube fragmentation during rolling can affect later recrystallization texture. Since retained Cube grains can grow during recrystallization and strengthen Cube texture, fragmentation of Cube-oriented regions during rolling may reduce the growth advantage of Cube during annealing (Cantergiani et al., 2022).

Mao and Yang (2014) further showed that recrystallization texture depends on initial texture, deformation texture, recovery condition, and effective driving force. In high-purity aluminum with recovery pretreatment, a strong Cube recrystallization texture formed, especially when

the initial material contained Cube texture. In commercial-purity aluminum with strong Copper rolling texture, an approximately 18° ND-rotated Cube texture formed, which was related to a favorable 40° $\langle 111 \rangle$ orientation relationship with the Copper deformation texture (Mao & Yang, 2014).

The deformation route can also change the recrystallization texture by changing the deformation texture before annealing. Ma et al. (2014) studied AA3105 sheets cold rolled by unidirectional rolling and cross-rolling, followed by annealing at 482°C for 3 h. In the annealed unidirectionally rolled sheets, the P component became dominant and increased with rolling reduction, while the annealed cross-rolled sheets followed a different texture path from $\{012\}\langle 021 \rangle$ through $\{163\}\langle 235 \rangle$ toward P and Goss as reduction increased. The authors concluded that P texture formation required the presence of the Copper component in the deformed matrix because of a favorable orientation relationship between P and Copper (Mao & Yang, 2014).

Nayan et al. (2019) reported a similar effect of rolling route in AA2195. Unidirectional rolling produced a Copper-type deformation texture that later transformed into a strong Cube recrystallization texture after heat treatment, while two-step cross-rolling promoted a stronger Brass texture and led to a weaker annealing texture without a strong Cube component. This shows that the strain path affects recrystallization texture through the deformation structure and orientation distribution created during rolling (Nayan et al., 2019).

Du et al. (2024) also showed that the recrystallization texture after annealing depended on the deformation texture formed before annealing. In straight-rolled AA3003 sheets, M $\{113\}\langle 110 \rangle$ and P $\{011\}\langle 455 \rangle$ recrystallization textures were dominant, while pseudo cross-rolled sheets could develop r-Cube texture at intermediate reductions. These differences were linked to the deformation texture formed before annealing and to concurrent precipitation during annealing (P. Du et al., 2024).

Engler et al. (1996) showed that recrystallization texture in AA1145 depended on both deformation level and annealing temperature. As shown in Figure 1-17, the Cube orientation decreased with increasing deformation, while the R orientation increased. Cube was favored

at lower deformation levels because of rapid nucleation and favorable growth conditions, whereas R became more pronounced when Cube was less dominant. Particle-related nucleation also contributed to texture formation, but Cube- and R-oriented grains had stronger competitive nucleation and growth advantages. The influence of deformation level and annealing temperature on the dominant nucleation and growth mechanisms is summarized in Table 1-1 (Engler et al., 1996).

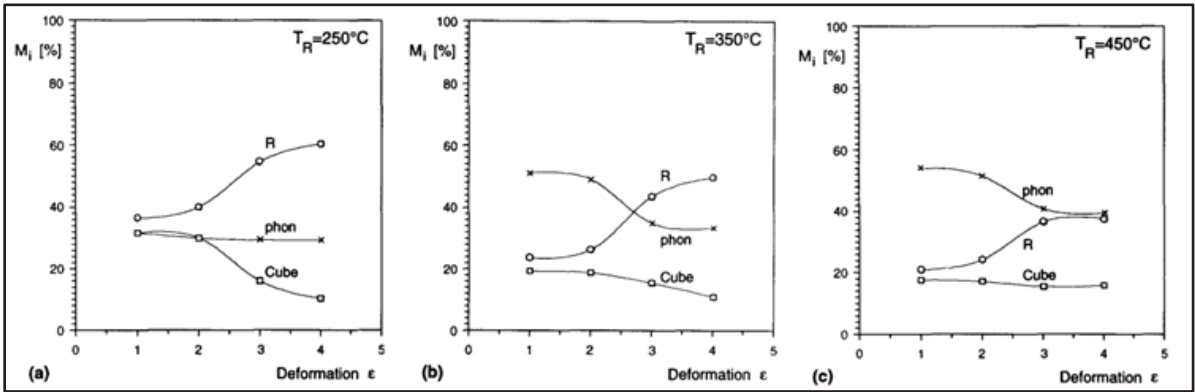


Figure 1-17 Volume fractions of the main recrystallization texture components ‘Cube’, ‘R’ and ‘phon’ in dependence on the deformation level. (a) $T_R=250^\circ\text{C}$; (b) $T_R=350^\circ\text{C}$; (c) $T_R=450^\circ\text{C}$

Taken from Olaf Engler et al. (1996)

Table 1-1 Influence of annealing temperature, T_R , and deformation, ϵ , on the formation of typical recrystallization texture orientation and the underlying mechanisms

Taken from Olaf Engler et al. (1996)

Orientation	Mechanism	T_R	ϵ
Cube	Preferred nucleation (Cube-bands)	↓	↓
	Growth selection	↑	↑
R	Preferred nucleation (Grain boundaries)	↓	↓
Random (phon)	PSN	↑	-

1.7.5 Particle-stimulated Nucleation, and Mg_2Si Effects

Particle-containing aluminum alloys add another level of interaction between deformation texture, local strain heterogeneity, and recrystallization texture. Sidor et al. (2015) studied an Al–Mg–Mn–Si alloy after different cold-rolling reductions and subsequent annealing. Their results showed that recrystallization texture changed with rolling reduction: Cube orientations became scattered along the η -fiber, the Q component increased with strain, and P and Goss changed only slightly. The study showed that recrystallization texture was controlled by competition between nucleation in the bulk deformed matrix and nucleation near particles. Their model, which included orientation selection during nucleation and micro-growth selection, reproduced the main features of the experimental recrystallization textures (Sidor et al., 2015).

Engler and Hirsch (2002) explained that different recrystallization texture components in AA6xxx sheets can originate from different nucleation mechanisms. Cube texture can develop from Cube bands already present in the deformed microstructure, while R orientation can develop from grain-boundary nucleation and has a favorable growth relationship with S components in the rolling texture. P and weak or random textures can be associated with particle-related nucleation around large particles (Engler & Hirsch, 2002).

The kinetics of recrystallization are strongly affected by solutes and particles. High-purity materials can recrystallize at lower temperatures, while impurities and alloying elements can retard recrystallization by reducing grain-boundary mobility. Solute atoms may segregate to moving boundaries and produce solute drag, which slows boundary migration and increases the apparent activation energy for recrystallization. Second-phase particles can also restrict boundary motion through particle pinning or Zener pinning (Raabe, 2014). Therefore, recrystallization during annealing depends not only on stored deformation energy, but also on the interaction between solutes, particles, and moving grain boundaries.

The effect of particles depends strongly on their size, distribution, and stability during annealing. Coarse particles can promote localized recrystallization by creating deformation

heterogeneities during rolling, while fine particles can restrict grain-boundary migration through Zener drag (K. Adam & Field, 2023; L. H. Liao, Jin, Gallerneault, & Esmacili, 2014). Adam and Field (2023) showed this behavior in hot-rolled AA7050 aluminum alloy. They found that large particles produced local deformation heterogeneities during rolling, creating regions with high stored energy and randomized local texture. During annealing, recrystallized grains formed near these coarse particles, showing that particle distribution can strongly affect the spatial pattern of recrystallization (K. Adam & Field, 2023).

Sidor et al. (2015) also showed that large non-deformable constituent particles can produce particle-affected deformation zones during cold rolling. These zones create local strain fields that differ from the macroscopic rolling strain and become preferred sites for recrystallized grain nucleation during annealing. As a result, recrystallization texture is influenced not only by the bulk rolling texture, but also by local deformation around particles and nucleation near particles (Sidor et al., 2015).

In AA6xxx alloys, the Mg_2Si precipitation state is especially important because it can change both boundary mobility and the dominant recrystallization mechanism (Engler & Hirsch, 2002). Engler and Hirsch (2002) showed that fine Mg_2Si particles can impede recrystallization through Zener drag and selectively suppress particle-related nucleation, which may favor a stronger Cube texture. In contrast, large particles can promote particle-related recrystallization and weaken the recrystallization texture. Therefore, the final texture in AA6xxx sheets depends strongly on the whole processing chain, including homogenization, hot rolling, hot-band precipitation, cold rolling, and final solution annealing (Engler & Hirsch, 2002).

Liao et al. (2014) studied recrystallization in a co-cast AA3xxx–AA6xxx aluminum alloy laminate and showed that recrystallization behavior differed between the AA3xxx clad layer and the AA6xxx core layer. Two cold-reduction levels, 60% and 80%, were examined. The rolled sheets were then annealed at 380, 420, and 480 °C for 1 h, and also at 540 °C for 30 s to represent a solutionizing-type treatment for the AA6xxx core. The clad layer developed finer and more equiaxed grains, while the core layer contained larger and more elongated grains. This difference was linked to the particle state in each layer. Large constituent particles promoted recrystallization nucleation, especially in the clad and interface regions, while fine

particles restricted boundary migration through Zener drag. In the AA6xxx core layer, original grain boundaries were identified as the main recrystallization nucleation sites (L. H. Liao et al., 2014).

Annealing temperature can also change the effect of particles by modifying their stability. Liao et al. (2014) showed that, at temperatures below solutionizing, fine particles restricted recrystallization and grain growth through Zener drag. During short annealing at 540 °C for 30 s, fine Mg₂Si/Si-type particles dissolved in the AA6xxx core, reducing Zener drag and increasing the net driving pressure for recrystallization. Higher cold reduction also increased stored energy and helped produce more equiaxed grains (L. H. Liao et al., 2014).

Second-phase particles during heat treatment can also influence recrystallization by changing the particle population during annealing (Q. Du, Poole, & Wells, 2012). Du et al. (2012) developed a mathematical model for predicting precipitation kinetics during heat treatment of multicomponent aluminum alloys. Their model predicted particle size distribution, number density, composition, and volume fraction during heat treatment (Q. Du et al., 2012). Later, Du et al. (2018) used a CALPHAD-coupled Kampmann–Wagner numerical framework to predict microstructure evolution during non-isothermal heat treatment of AA6061 and AA6082 alloys. Their results showed that multimodal Mg–Si particle size distributions can form during cooling because of multiple nucleation events and local microchemistry variations, especially in AA6082 (Qiang Du, Jia, Tang, & Holmedal, 2018).

Chen et al. (2024) showed that the recrystallization texture in an Al–Mg–Si–Zn alloy sheet after T4P treatment depended on the deformation texture created during cold rolling. At lower reductions, Cube-band nucleation was dominant, while at higher reductions, particle-related nucleation became more significant. In the 82% cold-rolled sheet, the higher amount of Copper and S deformation texture promoted CubeND and P orientations during recrystallization, which was related to a favorable $\langle 111 \rangle$ relationship between CubeND and the Copper/S-oriented deformed matrix (K.-X. Chen et al., 2024). This shows that increasing deformation can change both the stored-energy state and the dominant recrystallization mechanism.

Kumar et al. (2020) studied microstructure and texture evolution in AA3003 aluminum alloy during casting, hot rolling, cold rolling, and annealing. Their EBSD maps show that the as-cast alloy contained large grains and Al–Fe–Si particles along grain boundaries, as shown in Figure 1-18. After hot rolling at 500 °C and subsequent cold rolling, the grains became elongated and banded along the rolling direction, while second-phase particles were fragmented into smaller particles. After annealing at 450 °C for 16 h, a more uniform microstructure with an average grain size of 22 μm was obtained, although the particle-size distribution remained non-uniform (Kumar et al., 2020).

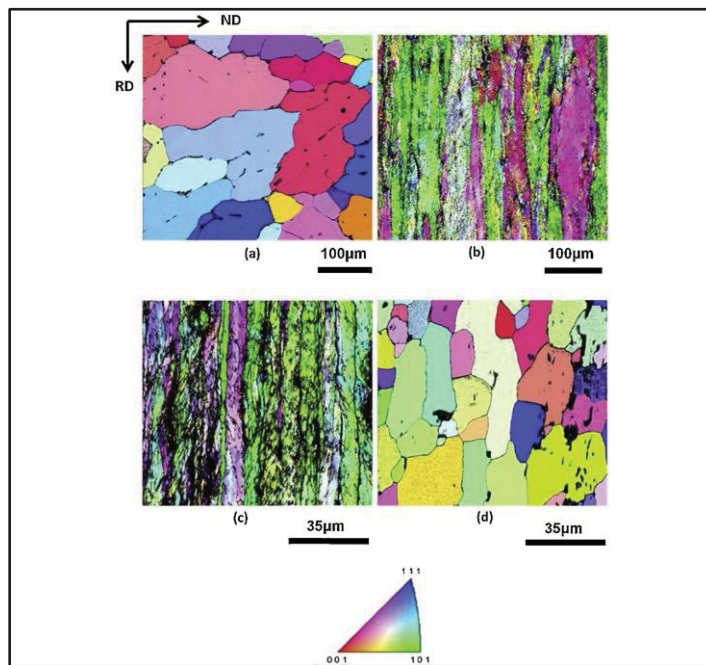


Figure 1-18 Electron backscattered diffraction (EBSD) inverse pole figure (IPF) maps of (a) as-cast, (b) hot rolled, (c) cold rolled, and (d) annealed samples

Taken from Kumar et al. (2020)

Kumar et al. (2020) also showed that texture changed during processing. The as-cast alloy had a weak Cube texture, while cold rolling reduced the Cube component and promoted strong Brass, Copper, and S components. After annealing, the Cube component did not increase significantly, and the annealing texture was mainly attributed to particle-stimulated nucleation. This was related to recrystallization around large particles, which promoted a more random

recrystallization texture instead of a strong Cube texture. Therefore, particles can change the expected transition from deformation texture to recrystallization texture (Kumar et al., 2020).

Particles can therefore strongly affect recrystallization during post-rolling annealing. Solutes can slow boundary migration through solute drag, while fine particles can restrict grain-boundary motion through Zener drag (Raabe, 2014). Coarse particles can promote localized recrystallization by creating deformation zones during rolling, while particle dissolution at high temperature can reduce pinning and increase the net driving pressure for recrystallization (L. H. Liao et al., 2014). In AA6xxx alloys, Mg_2Si precipitation is particularly important because it can affect both recrystallization kinetics and recrystallization texture (Engler & Hirsch, 2002). Therefore, the final annealed microstructure depends on the combined effect of deformation history, annealing condition, particle distribution, and precipitation state.

1.7.6 Through-thickness Effects and Overall Interpretation

Through-thickness texture variation can also affect recrystallization texture after annealing (Engler et al., 2001). Engler et al. (2001) showed that, after recrystallization annealing in commercial-purity AA1000A aluminum, the sheet center mainly developed Cube-related texture, while the surface layers retained or developed shear-related recrystallization textures. The $\{111\}$ //ND fiber became stronger in the warm-rolled surface layers, while the $\{001\}<110>$ component decreased (Engler et al., 2001). A similar effect of local deformation condition was reported by Zhang et al. (2019) in extruded aluminum profiles. They observed that the profiles mainly contained $<111>$ and $<100>$ fiber textures, with lower texture intensity near the surface than at the center. They related this surface texture weakening to more complete dynamic recrystallization near the surface, which reduced the original deformation texture (C. Zhang, Wang, Zhang, Zhao, & Chen, 2019).

Therefore, recrystallization texture after rolling and annealing depends on the prior deformation texture, degree of recrystallization, particle state, and local deformation condition through the sheet thickness. Annealing can increase random texture and promote recrystallization-related components such as Cube, Goss, Q, and R, but some deformation-

texture components may remain if recrystallization is incomplete (Sofia Papadopoulou et al., 2020; Thomesen et al., 2021). These factors are important for interpreting the microstructural evolution of rolled AA6xxx alloys because recovery and recrystallization modify the deformation structure produced during rolling and influence the final texture and anisotropy.

1.8 EBSD-based Identification of Deformation and Recrystallization Textures

Plastic deformation produces a heterogeneous substructure rather than a uniform distribution of defects. During rolling, dislocations organize into cell boundaries, cell blocks, incidental dislocation boundaries, and geometrically necessary boundaries. With increasing strain, these structures become finer, more closely spaced, and more misoriented, and in heavily rolled aluminum they may develop into lamellar structures nearly parallel to the rolling plane (Hansen & Juul Jensen, 2011). This heterogeneous deformation structure is important because recrystallization does not begin uniformly throughout the material. Instead, regions with higher stored energy, stronger orientation gradients, and larger local misorientation can recrystallize earlier than regions with lower internal misorientation (Traka, 2022).

EBSD is useful for analyzing this transition because it links microstructure with crystallographic orientation at the grain and subgrain scale. Traka (2022) showed that regions with higher local orientation gradients, higher KAM values, and higher stored energy recrystallized earlier during annealing, while regions with lower internal misorientation resisted recrystallization (Traka, 2022). Yamamoto et al. (2011) also showed in an Al–Mg–Si alloy that recovery and recrystallization do not occur as a simple single-step process. After cold rolling and annealing, some subgrains developed into coarse recovered regions, while others grew into recrystallized grains (Yamamoto, Tsukamoto, & Okai, 2011). This indicates that separating deformed, recovered, and recrystallized regions requires parameters that can describe internal orientation spread and local deformation heterogeneity.

Several EBSD-based parameters can be used to describe the deformation state. KAM reflects short-range local misorientation between neighboring pixels and is commonly used to identify regions with high local orientation gradients. GND density is related to lattice curvature and

provides an estimate of geometrically necessary dislocations (Field, Merriman, Allain-Bonasso, & Wagner, 2012).

However, both parameters are sensitive to EBSD acquisition conditions. Field et al. (2012) showed that GND estimates depend strongly on step size, orientation resolution, and whether the analysis is based on 2D or 3D data. KAM is also step-size dependent because it does not include an explicit physical length scale unless it is normalized by step size. Therefore, KAM and GND are useful for visualizing local deformation heterogeneity and stored-energy-related regions, but they should be used cautiously as direct criteria for separating recrystallized and deformed grains (Field et al., 2012).

GOS provides a grain-level measure of internal orientation spread. It is calculated as the mean misorientation between all pixels in a grain and the mean orientation of that grain. High GOS usually indicates a deformed grain with internal orientation gradients, while low GOS is usually associated with a recrystallized or weakly deformed grain (Abdelhak Ayad, Allain-Bonasso, Rouag, & Wagner, 2012). Ayad et al. (2012) showed that GOS increases with plastic deformation and decreases during annealing, reflecting the transition from a deformed to a recrystallized state (Abdelhak Ayad et al., 2012). Similarly, Lai et al. (2025) used GOS to follow recovery and recrystallization in cold-rolled AA3005, where GOS decreased rapidly during annealing as the microstructure recovered and recrystallized (Lai, Li, Yang, Tan, & Yan, 2025).

However, GOS alone can be affected by grain size and grain detection parameters. Larger grains may contain orientation gradients over longer distances, while small deformed or substructured grains may show low internal orientation spread and be incorrectly classified as recrystallized (A. Ayad et al., 2021). For this reason, GOS/D has been proposed as a complementary parameter because it normalizes GOS by grain diameter and reduces the influence of grain size (Abdelhak Ayad et al., 2012). Allain-Bonasso et al. (2012) showed that GOS/D is especially sensitive to plastic deformation heterogeneity. During tensile deformation of IF steel, GOS increased by about 490%, while GOS/D increased by about 850%, showing that GOS/D can capture accumulated intragranular deformation more clearly than GOS alone (Allain-Bonasso, Wagner, Berbenni, & Field, 2012).

Ayad et al. (2021) directly addressed the separation of recrystallized and non-recrystallized grains in partially recrystallized microstructures. They showed that GOS alone can misclassify small deformed or substructured grains as recrystallized. To reduce this problem, they proposed a combined criterion based on grain size larger than 3 pixels, $\text{GOS} < 2.5^\circ$, and $\text{GOS}/D < 1^\circ/\mu\text{m}$. This criterion gave a very low recrystallized fraction in the deformed state and a very high recrystallized fraction in the fully recrystallized state, indicating reliable classification across different recrystallization stages (A. Ayad et al., 2021).

This combined method is especially relevant for partitioning deformation texture from recrystallization texture. GOS describes grain-scale internal orientation spread, while GOS/D adds a size-normalized measure of intragranular misorientation. Using both parameters reduces the risk of misclassifying small deformed grains and provides a more reliable way to separate recrystallized grains from deformed or recovered grains in partially recrystallized rolled microstructures.

1.9 Processing-dependent Anisotropy and Formability of Rolled AA6xxx Alloys

1.9.1 Texture, Microstructure, and Anisotropy Indicators

Crystallographic texture is one of the main factors controlling mechanical anisotropy in rolled and extruded aluminum alloys. Since the deformation behavior of a single crystal depends on crystallographic orientation, a polycrystalline sheet with preferred grain orientations can show different mechanical responses along different loading directions (Kestens & Pirgazi, 2016). In precipitation-hardened aluminum alloys, the final response is also affected by precipitate state, grain morphology, and processing history (Nayan et al., 2019). Therefore, anisotropy in aluminum sheets and profiles should be interpreted through the combined effects of texture, microstructure, and thermomechanical processing.

Although crystallographic texture is a major source of anisotropy in aluminum alloys, it is not the only factor controlling directional mechanical response. Hansen and Juul Jensen (2011) explained that anisotropy in rolled sheets can arise from both crystallographic texture and

microstructural directionality. Crystallographic texture affects flow stress because grains with different orientations deform differently. In addition, aligned deformation-induced boundaries can create structural anisotropy in the rolled material (Hansen & Juul Jensen, 2011).

Mechanical properties in aluminum alloys are also affected by several microstructural features, including recrystallized grain size, recrystallized fraction, particle or precipitate content, and crystallographic texture. This point is especially important in precipitation-hardened aluminum alloys, where strength anisotropy can result from the combined effects of crystallographic texture and precipitate strengthening (K. Adam & Field, 2023).

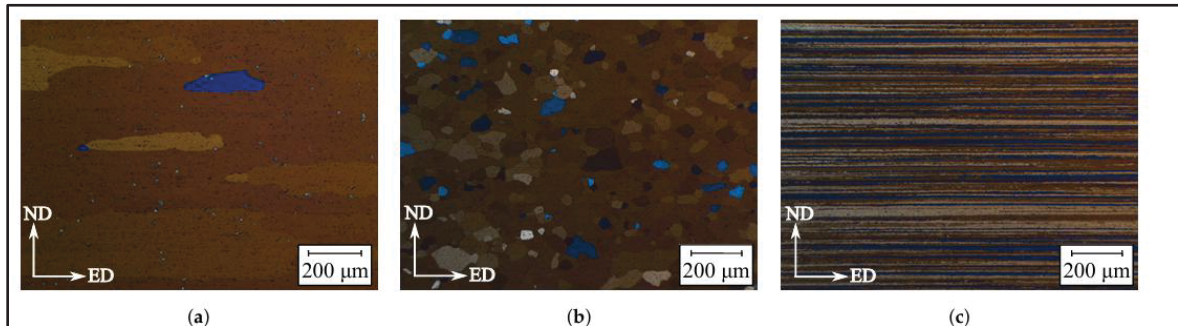


Figure 1-19 Optical micrographs showing the grain structure of the extruded AA6xxx alloys: (a) AA6061, (b) AA6063, and (c) AA6110. AA6061 and AA6063 exhibit recrystallized grain structures, while AA6110 shows a non-recrystallized structure with highly elongated grains along the extrusion direction

Taken from Thomesen et al. (2021)

Previous studies have shown that recrystallized and fibrous grain structures can produce different levels of plastic anisotropy in deformed aluminum alloys (Khadyko, Marioara, Dumoulin, Børvik, & Hopperstad, 2017; Thomesen et al., 2021). Thomesen et al. (2021) examined this behavior in extruded AA6061, AA6063, and AA6110 using smooth and notched tensile specimens tested in different directions relative to the extrusion direction. The micrographs shown in Figure 1-19 indicate that AA6061 and AA6063 had recrystallized grain structures, while AA6110 retained a non-recrystallized, strongly elongated grain structure. The texture analysis showed that AA6061 and AA6063 had strong Cube-dominated recrystallization textures with a Goss component, whereas AA6110 had a lower-intensity

deformation-type texture with orientations distributed along the β -fiber (Thomesen et al., 2021).

The stress–strain curves shown in Figure 1-20 indicate that the mechanical response changed with tensile direction, confirming the presence of plastic anisotropy. AA6061 and AA6063, which had recrystallization textures, showed strong anisotropic work hardening due to texture differences and texture evolution during tensile deformation. In contrast, AA6110, with deformation texture and pancake-shaped grains, showed a different anisotropic response and lower anisotropic hardening. Since all alloys were tested in the same temper condition, ageing was not considered the main cause of these differences. Thomesen et al. (2021) concluded that crystallographic texture was the main factor controlling strength anisotropy, although grain morphology and particle distribution also affected ductility and failure anisotropy (Thomesen et al., 2021).

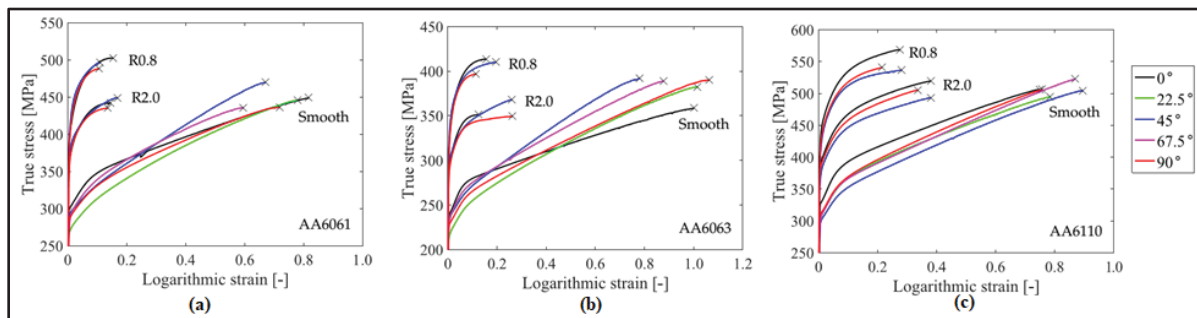


Figure 1-20 Representative stress-strain curves for all tested combinations of tensile direction and specimen geometry of alloy (a) AA6061, (b) AA6063 and (c) AA6110

Taken from Thomesen et al. (2021)

1.9.2 Through-thickness Texture Gradients, Formability, and Earing

In rolled 6xxx aluminum sheets, through-thickness texture gradients can also affect plastic anisotropy. Ghosh et al. (2015) studied EN AW-6016 and EN AW-6061 sheets in the T4 condition and measured texture at the surface, subsurface, and mid-thickness. Their results showed that mid-thickness texture alone was not always sufficient for accurate r-value prediction. Although both alloys had Cube-dominated recrystallization textures, the Goss

component varied through the thickness and was about three times higher at mid-thickness than at the surface. This gradient affected the r -value profile, especially in the 90° direction, where predictions based only on mid-thickness texture overestimated the experimental response (Ghosh, Miroux, & Kestens, 2015).

Chen et al. (2018) also showed that through-thickness texture variation can influence the macroscopic plastic response. They compared unrecrystallized and fully recrystallized conditions using tensile tests at 0° , 45° , and 90° relative to the processing direction. The unrecrystallized condition contained elongated grains and β -type texture, while the fully recrystallized condition contained equiaxed grains with Cube and Goss textures. The unrecrystallized material showed lower yield stress and work hardening at 45° , while the recrystallized material showed a more uniform response. The r -values at 45° and 90° were strongly affected by through-thickness texture gradients (J. Chen et al., 2018).

Texture control is also important for formability and deep drawability. Engler et al. (2001) studied cold rolling, warm rolling, and final recrystallization annealing in commercial-purity AA1000A aluminum sheet. Their aim was to develop a strong $\langle 111 \rangle \parallel$ ND fiber texture because this texture can improve Lankford r -values and deep drawability. They showed that the route with higher warm rolling produced a stronger $\langle 111 \rangle \parallel$ ND fiber, especially near the surface layers, and improved the predicted r -value response. This indicates that deep drawability depends not only on the presence of favorable texture components, but also on how these components develop through the sheet thickness after rolling and annealing (Engler et al., 2001).

The formability and anisotropy of aluminum sheets are closely related to the balance between deformation texture, recrystallization texture, and random texture. In aluminum alloys, good formability depends on the balance between Cube recrystallization texture and deformation texture components such as Brass, Copper, and S (Kumar et al., 2020). Pole figure and inverse pole figure intensity can be used as indicators of texture strength. Higher PF/IPF intensity means stronger preferred orientation, which usually corresponds to lower isotropic behavior and stronger directional differences in mechanical response (Sofia Papadopoulou et al., 2020).

Therefore, improving formability requires control of the texture balance rather than simply increasing or suppressing one texture component.

Papadopoulou et al. (2020) showed this relationship in AA5182 sheets. Cold rolling increased strength and mechanical anisotropy by elongating grains, strengthening rolling texture, and increasing pole figure intensity. The r -value decreased in the rolling direction but increased in the 45° and 90° directions, showing a more direction-dependent plastic response after rolling. This behavior was related to the development of typical rolling-texture components such as S, Brass, Copper, and Taylor. After annealing, anisotropy decreased because the random texture fraction increased, texture intensity decreased, and r -values became closer to those of the original hot-rolled sheet (Sofia Papadopoulou et al., 2020).

Texture balance also affects earing during deep drawing. Papadopoulou et al. (2020) related Copper, S, and Brass components to 45° ears, while Cube texture was associated with 90° ears (Sofia Papadopoulou et al., 2020). Kumar et al. (2020) also reported that rolling-texture components such as Copper, S, and Brass are associated with 45° ears, while Cube texture is associated with $0^\circ/90^\circ$ ears (Kumar et al., 2020). These results show that earing behavior depends on the relative balance between rolling texture, recrystallization texture, and random texture (Sofia Papadopoulou et al., 2020).

Texture evolution during recrystallization can also change directional elastic behavior. Obermayer et al. (2025) studied cold-rolled AA1050 sheets annealed at 270°C and 290°C . The cold-rolled samples showed higher Young's modulus values in RD and TD and lower values in the diagonal direction. After recrystallization, directional stiffness became more uniform because recrystallization changed the crystallographic texture. They proposed that directional Young's modulus, measured by impulse excitation technique, can be used to estimate static recrystallization kinetics, with EBSD and yield-strength measurements used for validation (Obermayer et al., 2025).

1.9.3 Effects of Rolling Route, Hot-band Texture, and Rolling Temperature

Processing route can change anisotropy by changing the deformation texture and recrystallization response before final heat treatment. Wang et al. (2018) studied an Al–Mg–Si–Cu alloy processed by hot rolling, intermediate annealing, and cold rolling. Increasing hot-rolling reduction from 30% to 50% increased dislocation density, subgrain boundaries, LAGBs, and stored energy. At 30% hot rolling, strong shear textures such as Inverse Brass and Rotated Cube were formed. At 50% reduction, the texture shifted toward E $\{111\}\langle 110 \rangle$ and F $\{111\}\langle 112 \rangle$ components, while the recrystallization-texture fraction also increased. This change affected anisotropy: the 30% hot-rolled sheet showed weak anisotropy and improved elongation, while the 50% hot-rolled sheet showed stronger anisotropy, especially in elongation. This shows that hot-rolling reduction and subsequent processing can change directional mechanical properties through their effect on texture and microstructure (Y. Wang et al., 2018).

The starting hot-band texture can also have an inherited effect on later formability. Sidor (2021) studied Al–Mg–Si sheets with different hot-band textures that were subsequently cold rolled to 86% reduction and annealed at 550 °C for 30 s. The hot-band texture affected the cold-rolled deformation texture, the final recrystallization texture, and the resulting plastic anisotropy. After cold rolling, all samples developed textures mainly along the α -, β -, and η -fibres, but the intensity and distribution of these fibres depended on the initial hot-band texture. Weak hot-band textures produced a more homogeneous β -fibre after cold rolling, whereas a hot band that already contained β -fibre orientations intensified these components during deformation. After annealing, the recrystallization texture was mainly distributed along the α -, θ -, and η -fibres, with components such as Cube, Goss, P, Q, E, and rotated Cube. Brass and other deformation-texture components decreased strongly because low-stored-energy nuclei, such as Cube, Goss, and P, consumed regions with high dislocation density. Cube bands promoted Cube recrystallization texture, while particle-stimulated nucleation and shear-band nucleation contributed to the formation of P, Q, E, Goss, and rotated Cube components. The distributions of recrystallization-texture components along the α -, θ -, and η -fibres for samples A–D are shown in Figure 1-21. Sample A was asymmetrically hot rolled, while samples B–D

were symmetrically hot rolled and exhibited weak, Cube-dominated, and mixed Cube- β -fibre hot-band textures, respectively. The figure therefore illustrates the inherited effect of different hot-band textures after identical cold rolling and annealing (Sidor, 2021).

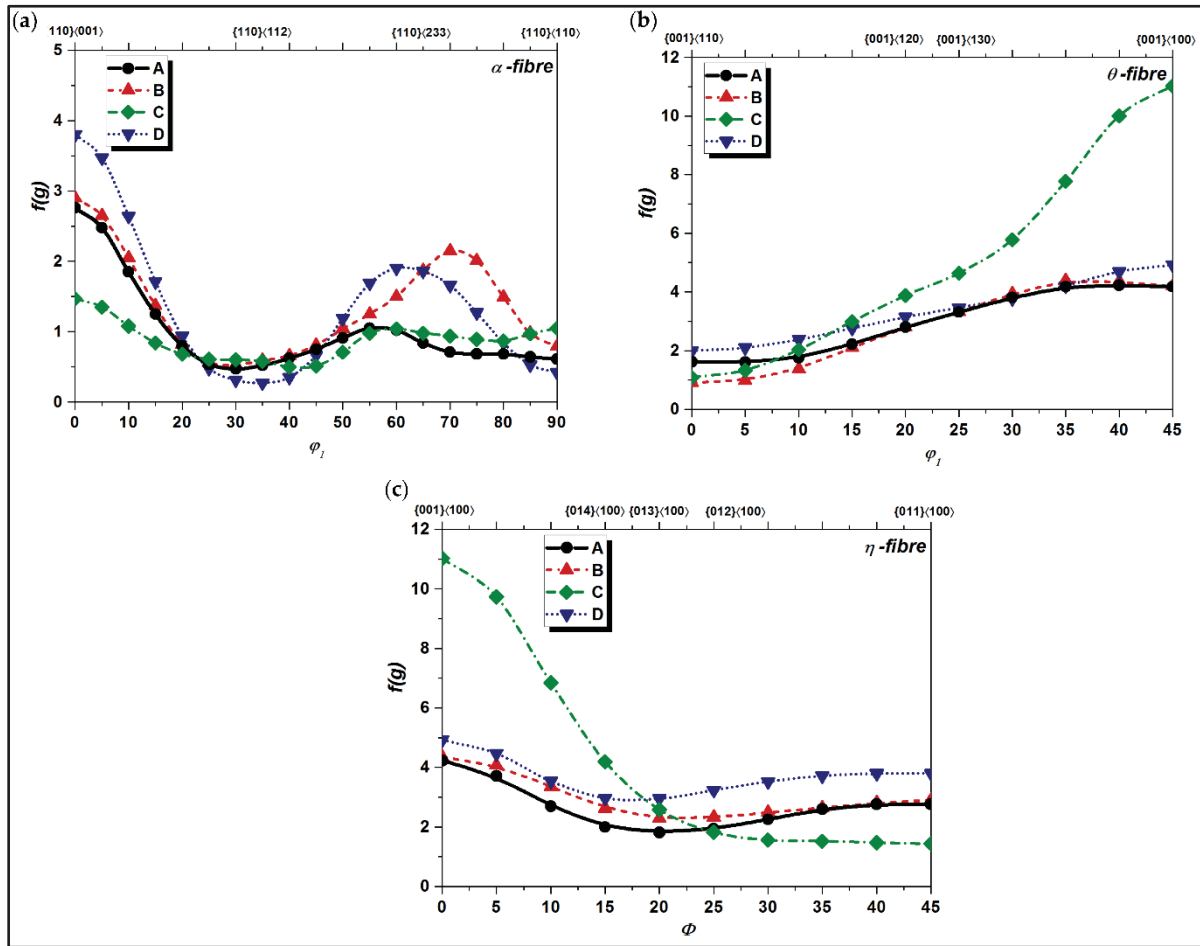


Figure 1-21 Orientation distributions along the (a) α -, (b) θ -, and (c) η -fibres after annealing at 550 °C for 30 s. Samples A–D had different initial hot-band textures

Taken from Sidor (2021)

The final recrystallization texture strongly affected the r-value profile. A strong Cube hot-band texture produced a strong Cube recrystallization texture, a clear V-shaped r-value curve, a low average r-value of about 0.50, and high planar anisotropy with $\Delta r = 0.43$. In contrast, weak hot-band textures produced weaker recrystallization textures, higher average r-values of about

0.67–0.72, and lower planar anisotropy. This shows that hot-band texture can influence later formability by affecting the deformation texture, recrystallization texture, and final plastic anisotropy after annealing (Sidor, 2021).

Warm rolling has also been used to improve formability by promoting favorable texture components. Inoue and Takasugi (2007) studied texture control in Al–Mg and Al–Mg–Si sheets and examined the relationship between average r -value and limiting drawing ratio. The limiting drawing ratio represents the ratio between the largest blank diameter that can be successfully drawn and the punch diameter. They showed that the average r -value is positively related to limiting drawing ratio, meaning that higher r -values improve deep drawability (Inoue & Takasugi, 2007).

The authors explained that the r -value is closely related to crystallographic orientation and can be increased through texture control. They also noted that r -values measured at different angles to the rolling direction reflect planar anisotropy. Conventional cold rolling produced typical β -fiber rolling textures, such as S and Brass, whereas warm rolling generated shear texture near the sheet surface because of roll–sheet friction. This shear texture included $\langle 110 \rangle // \text{RD}$ fiber components and $\{111\}$ -related orientations, which are favorable for higher r -values (Inoue & Takasugi, 2007).

After warm rolling and annealing, the Al–Mg alloy developed a mixed recrystallization texture, with retained shear-texture components near the surface and Cube plus R orientations near the sheet center. However, in AA6016, warm rolling did not improve the overall average r -value compared with cold rolling because the favorable surface texture was too shallow and did not extend sufficiently through the sheet thickness (Inoue & Takasugi, 2007).

Cold-rolling reduction can also control recrystallization texture and formability after heat treatment. Chen et al. (2024) studied Al–Mg–Si–Zn sheets cold rolled by 41%, 55%, 69%, and 82%, followed by T4P treatment, which included solution treatment at 560 °C for 5 min, water quenching, and pre-aging. Increasing cold reduction increased stored deformation energy and affected the recrystallized grain structure after heat treatment. The 82% cold-rolled sheet

showed the best formability, with finer recrystallized grains, a higher average r -value, and a more favorable texture distribution after T4P treatment (K.-X. Chen et al., 2024).

For rolled AA6082-T6, Ye et al. (2023) showed that anisotropy changes with deformation temperature. They tested specimens cut at 0° , 45° , and 90° relative to the rolling direction at 200, 300, and 400 °C. Increasing temperature strongly reduced flow stress and increased ductility. During elevated-temperature deformation, the stress–strain response was controlled by the competition between work hardening and dynamic softening. Work hardening was related to dislocation multiplication, while dynamic softening was associated with dynamic recovery, dynamic recrystallization, grain coarsening, and particle growth. At 200 °C, the matrix still contained high-density dislocations and polygonal dislocation cells, indicating recovery-dominated behavior. At 400 °C, dislocation density decreased, subgrains merged and grew, recrystallized grains appeared, and coarse precipitates formed inside grains and along grain boundaries (Ye et al., 2023).

The directional response was also temperature dependent. At 200 °C, the 0° and 90° samples showed higher stresses than the 45° sample. This difference was related to elongated grains, prior rolling texture, Brass texture, and differences in Schmid factor. The 45° sample had a higher Schmid factor, which made slip easier and reduced strength, while the 0° sample had the lowest Schmid factor and the highest resistance to deformation. The Brass texture also strengthened the rolling direction and weakened the 45° direction. As temperature increased, these directional differences became weaker. At 400 °C, the stress levels in all directions became close to about 40 MPa. This was attributed to easier slip activation at high temperature and partial transformation of the original rolled structure into a recrystallized structure. Dynamic recovery, dynamic recrystallization, subgrain growth, and precipitate coarsening reduced the effects of elongated grains and rolling texture. Therefore, anisotropy in rolled AA6082-T6 was strongest at lower hot-deformation temperature and became much weaker at higher temperature (Ye et al., 2023).

1.9.4 Heat treatment, Precipitation, and Final Anisotropic Response

The influence of heat treatment on anisotropy can be complex in heat-treatable aluminum alloys. Khadyko et al. (2017) studied plastic anisotropy in extruded AA6063 after T1, T6, T7, and O heat treatments. The extruded profile had a strong Cube texture with a minor Goss component, and tensile specimens were cut at different angles to the extrusion direction. Digital image correlation was used to measure in-plane strains, flow stress, normal strain, and strain ratios during tensile testing. The T6 and T7 conditions produced the highest yield strength because of needle-like precipitates oriented along $\langle 100 \rangle$. Heat treatment changed some anisotropy-related parameters, especially the plastic strain ratio in the 45° direction and the flow-stress ratio between the 0° and 22.5° directions, but these changes were smaller than the effect of crystallographic texture (Figure 1-22) (Khadyko et al., 2017).

Nayan et al. (2019) studied unidirectionally rolled and two-step cross-rolled Al–Cu–Li sheets after T6 treatment, including solution treatment, water quenching, and ageing. In the rolled condition, the directional variation in yield strength followed the Taylor factor trend, indicating that texture was the main source of anisotropy. However, after recrystallization, solution treatment, and ageing, the Taylor factor alone could not explain the yield-strength anisotropy, especially in the unidirectionally rolled sample (Nayan et al., 2019).

The authors showed that precipitation strengthening also contributed to the directional response. Using plastic and elastic inclusion models, they found that the unidirectionally rolled and aged sample had higher precipitation-strengthening factors along RD and TD than at 45° to RD. Therefore, precipitates strengthened the RD and TD directions more than the 45° direction. In contrast, these factors changed little with direction in the cross-rolled and aged sample. As a result, precipitation contributed more uniformly, and the strength response became nearly isotropic. This shows that rolling route can affect anisotropy not only through texture, but also through the directional contribution of precipitation strengthening after heat treatment (Nayan et al., 2019).

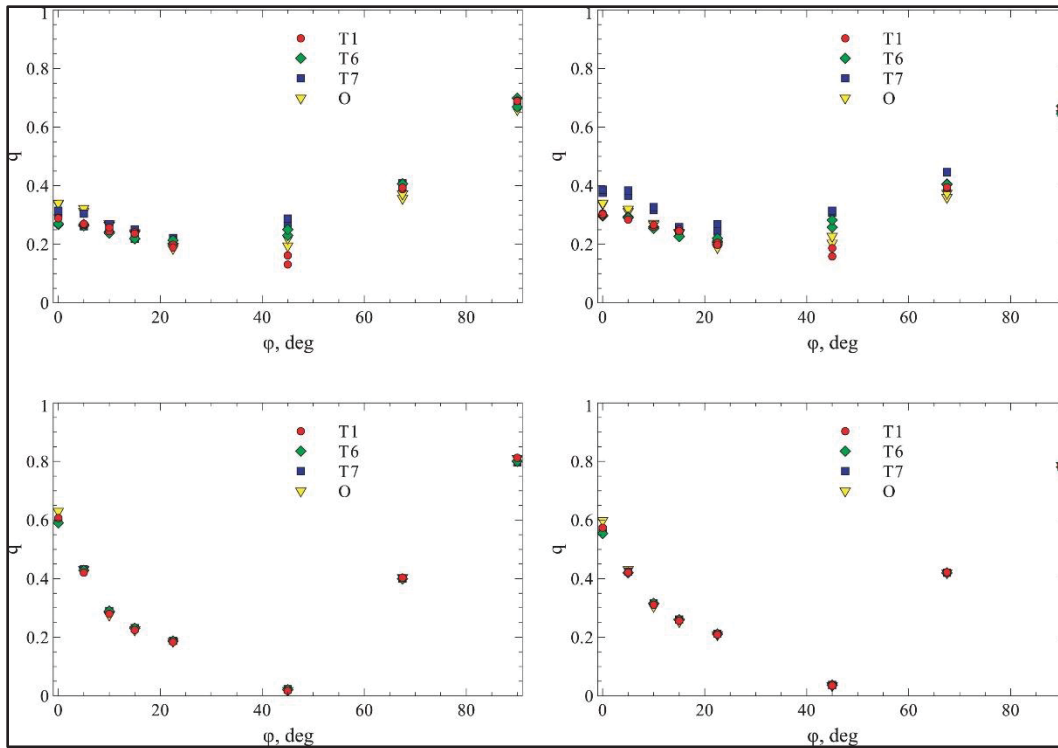


Figure 1-22 Effect of heat-treatment condition on the directional plastic strain ratios of extruded AA6063

Taken from Khadyko et al. (2017)

Rolling route can also affect mechanical anisotropy through its influence on texture and precipitation after heat treatment. Nayan et al. (2019) reported that the cross-rolled AA2195 sample showed about 7% in-plane yield-strength anisotropy in the rolled condition, which decreased to about 1.5% after heat treatment. This suggests that anisotropy was mainly controlled by texture in the rolled state, while both texture and precipitation contributed after aging (Nayan et al., 2019).

For AA6082, Debih and Ouakdi (2018) reported clear anisotropy after hot rolling. They studied a hot-rolled AA6082 plate that was heated to 500 °C for 4 h, rolled with 33% thickness reduction, and air quenched. Tensile specimens were taken in the longitudinal rolling, long transverse, and short transverse directions. Their results showed clear in-volume anisotropy after rolling. The longitudinal rolling and long transverse directions had similar tensile behavior, indicating near-planar isotropy in the rolling plane. However, the short transverse

direction behaved differently, with higher yield strength, tensile strength, stress-hardening capacity, and elastic modulus, but lower elongation. This anisotropy was related to rolling-induced microstructural directionality, including elongated band-like grains in the rolling plane and finer elliptical grains in the short transverse plane (Debih & Ouakdi, 2018).

Intermediate annealing can improve the balance between strength and isotropy in AA6082. Liu et al. (2022) studied AA6082 sheets intermediately annealed at 400 °C, 475 °C, or 550 °C before final cold rolling, solution treatment, and aging. Increasing the intermediate annealing temperature weakened the rolling texture after final cold rolling, especially the β -fiber and Brass-related components. The maximum ODF intensity decreased from 10.0 in IA400 to 5.5 in IA550, and the Brass intensity decreased from 8.0 to 3.6. This reduced in-plane yield-strength anisotropy from about 4.4% to 2.5%. After T6 treatment, the IA550 sheet showed the best overall response, with a yield strength of about 326 MPa, a higher average r-value, and lower planar anisotropy. This improvement was related to coarser Mg_2Si particles formed during intermediate annealing, which promoted particle-related nucleation during solution treatment and produced a finer, more equiaxed microstructure with higher fractions of CubeND and random textures (S. Liu, Xu, Jin, Li, & Peng, 2022a).

Texture control is also important for surface quality. A sharp recrystallization texture, especially strong Cube texture with rotations toward Goss, can promote ridging or paint-brush lines in AA6xxx sheets (Engler & Hirsch, 2002). Cantergiani et al. (2022) also suggested that controlling Cube stability during rolling may help reduce texture-related anisotropy and surface defects such as ridging or roping (Cantergiani et al., 2022).

Overall, texture-related anisotropy and formability are controlled by the full processing route, including hot-band texture, hot rolling, cold rolling, warm rolling, intermediate annealing, recrystallization, and final heat treatment. Favorable r-values and reduced earing require an appropriate balance between rolling texture, recrystallization texture, and random texture, as well as suitable through-thickness texture distribution (Inoue & Takasugi, 2007; Sofia Papadopoulou et al., 2020; Sidor, 2021). For AA6xxx alloys, processing also affects Mg_2Si precipitation and particle-related recrystallization, which can change the final strength, anisotropy, and formability (K.-X. Chen et al., 2024; S. Liu et al., 2022a).

CHAPTER 2

CONTEXT OF RESEARCH

2.1 Introduction

The previous chapter reviewed the main concepts related to AA6xxx aluminum alloys, with emphasis on microstructure, texture, recrystallization, and mechanical anisotropy. It also showed that the final behavior of these alloys depends on the interaction between alloy composition, initial condition, deformation path, and heat treatment.

Based on this background, the present chapter defines the research context of the thesis. It first presents the research problem and then introduces the objectives of the study.

2.2 Research Problem

AA6xxx series aluminum alloys are widely used in structural and transportation applications because of their strength-to-weight ratio, corrosion resistance, and heat-treatable formability. Their final mechanical response after thermomechanical processing is not controlled by one factor alone. It depends on alloy composition, initial microstructure, deformation path, rolling condition, and subsequent heat treatment. These factors influence texture evolution, recovery and recrystallization behavior, and mechanical anisotropy.

Previous studies have investigated deformation texture, recrystallization texture, and anisotropy in aluminium alloys. However, most studies focus on a single alloy, starting condition, or processing route. To the best of our knowledge, no exhaustive study has systematically examined the same AA6xxx alloy under a sufficiently broad range of initial

conditions, rolling parameters, deformation paths, and post-rolling heat treatments. Consequently, the available results are difficult to compare directly, and it remains challenging to draw clear conclusions regarding the origin of the different microstructural, textural, and mechanical responses of these alloys.

This limitation is particularly important because AA6xxx alloys can show different responses depending on their chemistry and starting condition. Differences in Mg–Si content and constituent particles can affect stored-energy accumulation and recrystallization behavior. In addition, the initial grain structure and inherited texture can influence texture stability and the development of anisotropic mechanical behavior. Therefore, a systematic comparison under comparable processing and characterization conditions is needed to clarify how these factors contribute to the final response.

Extrusion adds another difficulty because it involves complex material flow and strong strain gradients across the profile section. For this reason, extrusion is not treated as a deformation process in this thesis. Instead, the as-received extruded AA6082 material is considered as an initial condition. Rolling is selected as the main deformation process because it provides a more controlled route for comparing strain path, thickness reduction, rolling condition, and post-rolling heat treatment.

The central problem addressed in this thesis is therefore the limited systematic documentation of how selected AA6xxx alloys respond to rolling and post-rolling heat treatment when alloy chemistry and initial microstructural state are varied. This gap limits the understanding of how processing history controls microstructure, crystallographic texture, recrystallization, and anisotropic mechanical behavior.

2.3 Research Objectives

The main objective of this thesis is to investigate **in a systematic way** how deformation path, alloy composition, and initial microstructural state control texture evolution, recrystallization behavior, and anisotropic mechanical properties in selected AA6xxx aluminum alloys.

Rolling-based thermomechanical routes were selected because they provide controlled deformation paths. These routes allow the effects of unidirectional and cross deformation to be examined more clearly than in more complex deformation processes such as extrusion. In this thesis, the as-received extruded AA6082 condition is considered only as an initial state not the deformation route.

To achieve this main objective, the following specific objectives are defined:

- I. To determine the effect of cold-rolling path on deformation texture and anisotropic behavior

The effect of cold rolling in different orientations, including unidirectional and cross rolling, is investigated in AA6xxx alloys with different compositions and initial conditions. This objective focuses on how deformation path controls deformation texture, microstructural heterogeneity, and anisotropic mechanical response.

- II. To clarify the mechanisms responsible for texture evolution under pseudo-hot rolling conditions

Unidirectional and cross pseudo-hot rolling routes are investigated to examine texture evolution during high-temperature thermomechanical processing. This objective focuses on the combined effects of deformation, temperature, recovery, recrystallization, and precipitation-related changes on microstructure, texture, and mechanical properties.

- III. To identify the mechanisms controlling texture evolution during static recrystallization after rolling

Post-rolling heat treatments are used to examine recovery, static recrystallization, and recrystallization texture after cold rolling. This objective focuses on how alloy composition, initial condition, and prior deformation path affect recrystallization mechanisms, texture transformation, and final properties.

CHAPTER 3

METHODOLOGY

3.1 Introduction

This study explores the effects of various thermomechanical processes applied to 6xxx series aluminum alloys, focusing on recovery and recrystallization mechanisms and their subsequent impact on crystallographic textures and anisotropic mechanical properties. The experimental approach investigates cold rolling and pseudo-hot rolling as primary deformation paths, followed by post-deformation annealing treatments. At each stage, the material is thoroughly characterized using a comprehensive suite of metallurgical and mechanical tests.

Ultimately, this project contributes by generating foundational experimental data to be integrated into the deformation texture model currently under development by the LOPFA research group. By systematically identifying the influential mechanisms during recrystallization and texture evolution, the results of this thesis will provide the LOPFA team with a valuable, empirically validated tool to update their algorithms and predict recrystallization textures under various severe deformation processes. The specific experimental steps and methodologies are outlined in the following sections.

3.2 Materials and Initial Conditions

The materials used in this investigation were two commercial Al-Mg-Si alloys, AA6063 and AA6082. These alloys were selected because they exhibit different recrystallization behaviors. AA6063 is generally more prone to recovery and recrystallization during thermomechanical processing, whereas AA6082 shows a greater resistance to recrystallization.

To examine the recrystallization behavior under different initial microstructural states, three starting conditions were considered: as-cast AA6063, as-cast AA6082, and as-extruded AA6082. In this study, the as-cast condition was treated as the main reference state because it is more representative of industrial hot rolling practice. For this reason, a smaller number of experiments was carried out on the as-extruded material, mainly for comparison with the rolled as-cast condition.

Table 3-1 Chemical composition limits per EN 573-3 for AA6063 and AA6082 aluminum alloys (wt%)

Alloy	Mg	Si	Cu	Mn	Cr	Fe	Al
AA6063	0.45-0.9	0.2-0.6	max 0.1	max 0.1	max 0.1	max 0.35	Balance
AA6082	0.6-1.2	0.7-1.3	max 0.1	0.4-1.0	max 0.25	max 0.5	Balance

The as-cast materials were produced by horizontal direct chill casting. The as-cast AA6063 and AA6082 were received as bars with diameters of 95 mm and 100 mm, respectively, while the as-extruded AA6082 material was received as a commercial extruded bar with a diameter of 58 mm. Before being received for this study, the as-cast AA6082 bars had been homogenized at 560 °C for 2 h. For the subsequent thermomechanical processing, all starting bars were longitudinally sectioned to extract the experimental plates, which were then machined to an initial thickness of 5 mm prior to rolling. The starting materials and representative prepared sections are shown in Figure 3-1. The standard chemical composition limits of both aluminum alloys are given in Table 3-1.



Figure 3-1 Initial materials used in this study, including as-cast AA6063 and AA6082 billets and the as-extruded AA6082 bar, together with examples of longitudinally sectioned and machined plates prepared for rolling

3.3 Rolling Process and Equipment

This work forms part of a broader research program on the industrial extrusion process. Extrusion involves a complex thermomechanical history, with pronounced strain and temperature gradients across the material. Consequently, different regions of an extruded profile, such as the shear-dominated zones near the die and the central inflow regions, experience markedly different deformation and restoration conditions (Güzel et al., 2012). This heterogeneity makes it difficult to isolate the fundamental mechanisms governing deformation, recovery, and recrystallization.

To examine these mechanisms under more controlled conditions, rolling was adopted as a model deformation process. Unlike extrusion, rolling provides a simpler and more macroscopically uniform strain path, making it more suitable for following the evolution of microstructure and texture in a systematic manner. This framework facilitates the analysis of restoration during hot deformation, the effects of cold deformation, and the subsequent recrystallization behavior during post-deformation heat treatment. It also offers a reliable basis for interpretation, since rolling behavior has been extensively documented in the literature.

On this basis, both cold and pseudo-hot rolling were applied to the AA6063 and AA6082 alloys. This approach made it possible to follow the material response from the initial as-cast or as-extruded condition, through severe plastic deformation, and finally into partially or fully recrystallized states after heat treatment. The objective was to clarify how the initial state and the imposed deformation path influence the subsequent microstructural and textural evolution. All unidirectional and cross-rolling experiments, under both cold and hot conditions, were carried out at the Centre de métallurgie du Québec (CMQ) in Trois-Rivières using a Stanat two-high laboratory rolling mill equipped with rolls of 152.4 mm (6 inches) in diameter. To ensure accurate and repeatable thickness reductions, the mill was fitted with a digital readout scale connected to the adjustment system for precise control of the roll gap, and the pseudo-hot rolling setup included a Pyradia muffle furnace positioned adjacent to the mill as well as temperature monitoring using thermocouples. These elements of the experimental setup are shown in Figure 3-2. It should be mentioned that the muffle furnace was positioned adjacent to the rolling mill to minimize transfer time and limit temperature loss during successive passes.

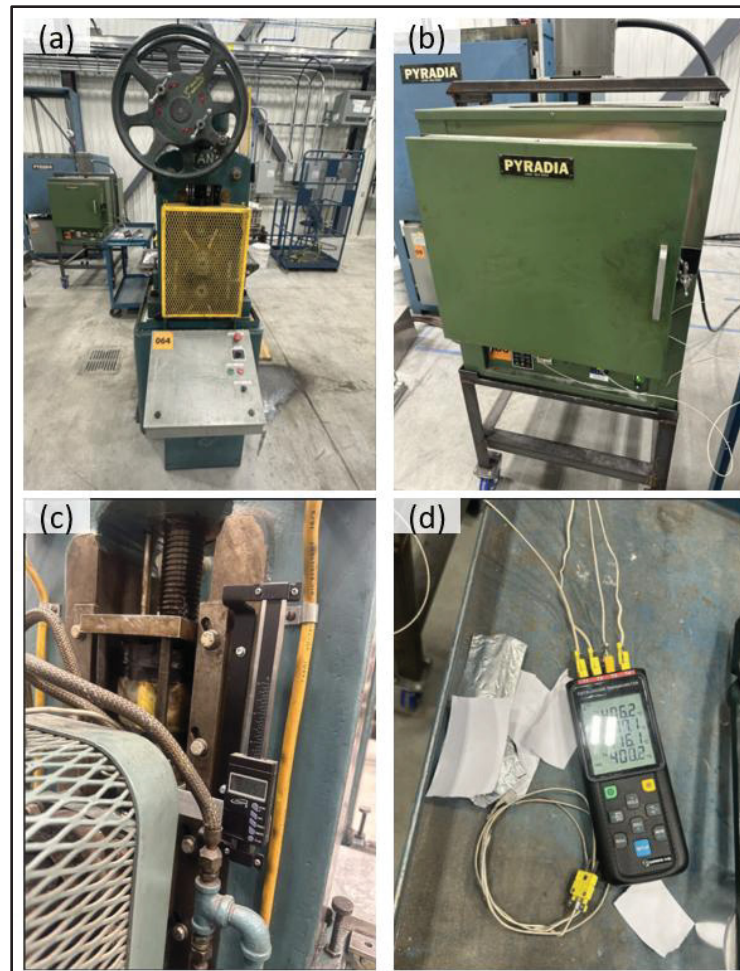


Figure 3-2 Experimental rolling setup used at CMQ: (a) Stanat two-high laboratory rolling mill, (b) Pyradia muffle furnace, (c) digital readout system for roll-gap control, and (d) thermocouple-based temperature monitoring system

3.4 Cold Rolling Process

3.4.1 Cold Rolling Procedures

To investigate the evolution of deformation texture and its dependence on strain path, the prepared plates were cold rolled at room temperature using two distinct strategies: unidirectional cold rolling (UCR) and cross cold rolling (CCR). In the UCR condition, the rolling direction was kept constant throughout all passes. In the CCR condition, the plate was

rotated by 90° about the normal direction between consecutive passes. For the as-extruded AA6082 condition, UCR was performed perpendicular to the initial extrusion direction in order to destabilize the strong starting crystallographic texture. As a result, the original extrusion direction corresponds to the transverse direction of the rolled plates.



Figure 3-3 Representative examples of plates processed under selected cold and pseudo-hot rolling conditions considered in this study

The experimental procedure was designed to establish a detailed baseline for the macroscopic deformation behavior of the AA6082 alloy. Accordingly, both the as-cast and as-extruded AA6082 plates were subjected to incremental cold rolling schedules, covering low to intermediate thickness reductions from 10% to 60%, followed by a high reduction level of

90%, starting from an initial thickness of 5 mm. Furthermore, to provide a comparative analysis with the AA6063 alloy, the as-cast AA6063 was rolled at a specific subset of targeted reductions, from 40% to 90%. This experimental design made it possible to assess the influence of strain level and strain path on the deformation response of the alloys. It also enabled a direct comparison of the role of constituent particle distribution and initial grain structure in the evolution of the deformed microstructure. Examples of plates processed under selected cold and pseudo-hot rolling conditions are shown in Figure 3-3.

3.4.2 Analytical Objectives for Cold Rolling

To comprehensively characterize the severely deformed state and understand the dominant deformation mechanisms, the cold rolled samples were systematically analyzed with a focus on the following parameters:

Incremental Texture Evolution: Tracking the development of crystallographic texture as a function of increasing thickness reduction across the entire rolling schedule.

Strain Path Dependency: Comparing the texture evolution and microstructural changes between the UCR and CCR strain paths.

Initial State Influence: Evaluating how the starting microstructural condition, specifically the as cast versus the as extruded state, dictates the final deformation texture.

Compositional Effects: Assessing the impact of the different chemical compositions of the AA6063 and AA6082 alloys on their respective deformation behaviors.

Grain Morphology: Quantifying the changes in grain size and the progressive evolution of grain aspect ratios as the material undergoes severe plastic strain.

3.5 Pseudo-hot Rolling Process

3.5.1 Pseudo-hot Rolling Procedures

A comprehensive series of Pseudo-hot rolling experiments was conducted to investigate the deformation behavior of the alloys at elevated temperatures. Processing under these thermal conditions fundamentally alters dislocation mobility and activates dynamic restoration mechanisms. The Pseudo-hot rolling was performed at a target deformation temperature of 400 °C.

Given the high thermal conductivity of aluminum alloys and the relatively thin initial gauge of the starting plates (5 mm), rapid heat loss to the ambient environment and the rolling cylinders was a critical experimental challenge. To ensure the deformation occurred as isothermally as possible at 400 °C, the samples were systematically returned to the adjacent muffle furnace between each consecutive rolling pass. Once the material reattained the target temperature, the subsequent rolling pass was executed immediately. Following the final targeted thickness reduction, the samples were instantaneously quenched in water. This rapid quenching process was essential to strictly "freeze" the high-temperature microstructure, thereby preserving the exact dynamically deformed state and preventing any subsequent meta-dynamic or static recrystallization from occurring during a slower cooling cycle.

Consistent with the cold rolling methodology, the hot deformation was applied via two distinct strain paths: unidirectional pseudo-hot rolling (UHR) and cross pseudo-hot rolling (CHR). Both the as-cast AA6063 and as-cast AA6082 plates were subjected to UHR and CHR at targeted intermediate and severe thickness reductions of 40%, 60%, and 90%. For the as-extruded AA6082 material, the UHR process was similarly performed at 40%, 60%, and 90% reductions, while the more complex CHR strain path was specifically evaluated at the maximum 90% reduction level.

3.5.2 Analytical Objectives for Pseudo-hot Rolling

The dynamically deformed and water quenched microstructures were subsequently analyzed to address the following specific research objectives:

Dynamic Restoration Mechanisms: Evaluating the extent of dynamic recovery and identifying the dominant recrystallization mechanisms operating during hot deformation.

High-Temperature Texture Evolution: Characterizing the evolution of crystallographic texture as a function of strain path and thickness reduction during Pseudo-hot rolling.

Microstructural Evolution: Quantifying the changes in grain morphology, grain size, and internal substructure with strain path and reduction level.

Comparative Deformation Analysis: Comparing the texture evolution and restoration behavior observed during Pseudo-hot rolling with the corresponding cold-rolling results to isolate the effect of temperature on the final anisotropic response.

3.6 Post-deformation Annealing Treatment

To systematically investigate the static recrystallization kinetics and the thermal stability of the deformation textures, multiple specimens were extracted from the center of the previously cold-rolled plates and subjected to isothermal annealing treatments. Two different heating routes were used in order to match the thermal requirements of the selected deformation states and annealing conditions.

For the intermediate-temperature analyses, specimens extracted from the 60% rolled as-cast AA6082 and the 60% rolled as-extruded AA6082 were annealed in a standard muffle furnace at 350 °C and 400 °C. These conditions were selected to compare the recrystallization behavior of different starting materials under slower restoration kinetics. To fully capture the onset and progression of static recrystallization, extended soaking times ranging from 5 to 1600 minutes were applied.

For the higher-temperature analyses, annealing was carried out at 520 °C in order to reproduce the typical die-exit temperature of extruded aluminum profiles. Under this condition, specimens from the 60% and 90% rolled as-extruded AA6082 were treated for much shorter times, between 1 s and 10 s, because the restoration kinetics were expected to be significantly faster. To achieve rapid and uniform heat transfer during these short treatments, a specialized molten salt bath at the National Research Council Canada (CNRC) in Chicoutimi was used. This approach made it possible to examine both the effect of higher annealing temperature on the 60% deformed state and the influence of a higher reduction level under identical rapid-heating conditions.

3.7 Microstructural and Texture Analysis

To characterize the microstructural evolution and crystallographic texture changes throughout the different thermomechanical conditions, Scanning Electron Microscopy (SEM), Energy Dispersive X-ray Spectroscopy (EDS), and Electron Backscatter Diffraction (EBSD) were employed. All microstructural observations were performed on the cross-section perpendicular to the rolling direction (the ND–TD plane). For the cross-rolled specimens, this sectioning was specifically performed perpendicular to the primary rolling direction to expose the first cross-section (designated as the RD1 plane).

3.7.1 Metallographic Preparation

Samples were extracted from the center of the deformed metals and cold mounted in epoxy resin (Figure 3-4a). The mounted specimens underwent a rigorous mechanical grinding process using successive grades of silicon carbide (SiC) grinding paper (240, 480, 600, 800, and 1200 grit). Following the completion of the mechanical grinding phase, the samples were thoroughly cleaned in an ultrasonic bath to remove any residual abrasive particles and surface debris.

Since Scanning Electron Microscopy (SEM) and Electron Backscatter Diffraction (EBSD) are highly sensitive to surface deformation, a final electrochemical polishing step was employed

to achieve a strain free surface. The mechanically ground samples were electropolished using a Struers electropolishing unit (Figure 3-4b and Figure 3-4c) equipped with a standard A3 electrolyte solution. The electrolyte mixture consisted of 40 mL perchloric acid, 240 mL butoxyethanol, and 400 mL methanol. The process was conducted at a constant voltage of 22 V and a flow rate setting of 20. The electropolishing duration was carefully monitored and varied between 10 and 20 s depending on the specific sample condition.

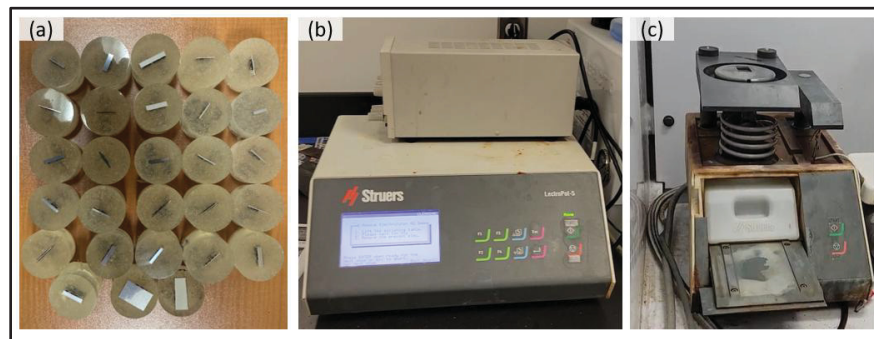


Figure 3-4 Microstructural sample preparation setup showing (a) cold-mounted aluminum specimens in epoxy resin, alongside the (b) control unit and (c) polishing cell of the Struers electropolishing system

3.7.2 Scanning Electron Microscopy and EDS

The microstructural and texture measurements were conducted using a Hitachi SU-70 field emission gun scanning electron microscope (FEG-SEM), as shown in Figure 3-5. Additionally, Energy Dispersive X-ray Spectroscopy (EDS) was utilized in conjunction with the Scanning Electron Microscopy (SEM) to qualitatively analyze the chemical composition of the matrix phases, secondary particles. The SEM and EBSD analyses performed in this study were carried out at an accelerating voltage of 20 kV.

For the EBSD data acquisition, the mapping step size was dynamically adjusted based on the required resolution and the specific microstructural state being analyzed. A step size of 3 μm was utilized at 100x magnification to capture broad grain statistics in larger areas, while a finer step size of 0.7 μm was employed at higher magnifications to resolve highly deformed subgrain structures. However, the heavily deformed 90% cold-rolled samples exhibited severe Kikuchi

pattern degradation and poor indexing rates under conventional Hough-transform-based acquisition. For this reason, advanced EBSD characterization of these specific specimens was carried out at the National Research Council of Canada (NRC) facility in Chicoutimi. To resolve the dense, highly strained substructure, data acquisition was performed at an accelerating voltage of 20 kV with a step size of 0.2 μm . Indexation was achieved using a dynamic pattern-matching approach (MapSweeper), in which experimental diffraction patterns were compared with dynamically simulated templates. Successful indexation was retained only for patterns with a normalized cross-correlation coefficient r greater than 0.15.



Figure 3-5 The Hitachi SU-70 field emission gun Scanning Electron Microscope (FEG-SEM) setup utilized for microstructural characterization, including Electron Backscatter Diffraction (EBSD) and Energy Dispersive X-ray Spectroscopy (EDS) analysis

3.7.3 EBSD Data Processing and Texture Analysis

The raw EBSD data and Kikuchi diffraction patterns were processed and analyzed using ATEX software (version 4.14). The software was used to extract spatial orientation data, determine average grain size, and visualize the crystallographic texture. In addition, ATEX was employed

to calculate the Schmid and Taylor factors and to estimate the primary recrystallized fraction under the different thermomechanical conditions.

To represent the EBSD results, several mapping approaches were used:

- I. Band contrast maps: Band contrast maps were used to visualize the quality of the diffraction patterns. Brighter regions generally correspond to areas with better diffraction quality, while darker regions indicate zones with poorer pattern quality.
- II. EBSD orientation maps: Orientation maps were used to visualize the crystallographic orientations of the grains. A standard inverse pole figure (IPF) color key, shown in Figure 3-6, was used to represent the orientations in the EBSD maps. This color representation provided a consistent basis for comparing orientation changes and texture evolution under the different processing conditions.

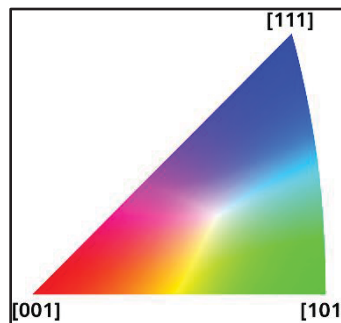


Figure 3-6 Standard inverse pole figure (IPF) color key used to represent crystallographic orientations in the EBSD maps

- III. Grain boundary maps: Subgrains were identified by low-angle grain boundaries (LAGBs) with misorientations between 2° and 15° , shown as red lines. Grains were identified by high-angle grain boundaries (HAGBs) with misorientations greater than 15° , shown as black lines.
- IV. Grain size distribution: Grain size distributions were extracted from the EBSD data to evaluate microstructural refinement and recrystallization. Since deformed and partially recrystallized states often exhibit heterogeneous grain structures, the full distribution provides a more representative basis for comparison than a single average value.

- V. Pole figures (PFs): The macroscopic crystallographic texture was represented using standard $\{111\}$ pole figures. A key aspect of the data processing methodology was the treatment of sample symmetry, which was selected according to the thermomechanical condition of the material. For the hot-rolled samples, standard orthotropic sample symmetry was applied, since hot deformation was accompanied by dynamic restoration and substantial grain growth. In contrast, no sample symmetry (triclinic symmetry) was imposed for the cold-rolled and statically annealed specimens in order to preserve the asymmetrical deformation textures produced by severe rolling.
- VI. Inverse Pole Figures (IPFs): In addition to the spatial maps, statistical IPFs were constructed specifically for the rolled as-extruded AA6082 condition. Given the strong duplex fiber texture inherent to the initial extruded material, IPFs provide a highly suitable representation of the macroscopic orientation distribution, enabling a clear quantitative assessment of its evolution during the cold rolling process.

Table 3-2 Euler angles and Miller indices of common texture components in FCC metals and alloys

Texture Component	Euler Angles (ϕ_1, ϕ, ϕ_2)	Miller Indices
Cube (Cb)	(0, 0, 0)	$\{001\} \langle 100 \rangle$
Goss (Gs)	(0, 45, 0)	$\{011\} \langle 100 \rangle$
Rotated-Goss (RGs)	(90, 45, 0)	$\{011\} \langle 011 \rangle$
Brass (Bs) I	(35, 45, 0)	$\{011\} \langle 2\bar{1}1 \rangle$
Brass (Bs) II	(145, 45, 0)	$\{011\} \langle \bar{2}11 \rangle$
Copper (Cu) I	(90, 35, 45)	$\{112\} \langle \bar{1}11 \rangle$
Copper (Cu) II	(270, 35, 45)	$\{112\} \langle 11\bar{1} \rangle$
Rotated-Copper (RCu) I	(0, 35, 45)	$\{112\} \langle 1\bar{1}0 \rangle$
Rotated-Copper (RCu) II	(180, 35, 45)	$\{112\} \langle \bar{1}10 \rangle$
S I	(61, 34, 64)	$\{2\bar{1}1\} \langle 1\bar{2}4 \rangle$
S II	(241, 34, 64)	$\{21\bar{1}\} \langle 1\bar{2}4 \rangle$
S III	(299, 34, 26)	$\{11\bar{2}\} \langle 4\bar{2}1 \rangle$
S IV	(119, 34, 26)	$\{\bar{2}11\} \langle 1\bar{2}4 \rangle$

- VII. Texture component quantification: To quantify texture evolution, the volume fractions of the principal texture components relevant to each processing series were calculated using a maximum angular deviation of 15° from their ideal $\{hkl\}\langle uvw \rangle$ orientations. The standard ideal orientations and associated variants considered in this study are listed in Table 3-2. In the results, the reported volume fraction of each main texture component corresponds to the sum of the volume fractions of its associated variants.

3.8 Grain Size Distribution

Grain sizes were extracted from EBSD maps as the equivalent-circle diameters (ECD) of the grain. The relationship used throughout is given in (3-1), which converts a grain's mapped area A to the equivalent diameter D of a circle having the same surface:

$$A = \pi \left(\frac{D}{2} \right)^2 \quad (3-1)$$

Grain morphology was quantified using the aspect ratio, defined as the minor-axis length divided by the major-axis length:

$$AR = \frac{\text{minor-axis length}}{\text{major-axis length}} \quad (3-2)$$

According to this definition, represents an equiaxed feature, whereas values approaching zero indicate increasingly elongated morphology. The same aspect-ratio definition was subsequently applied to both grains and second-phase particles.

Distributions are reported as normalized probability density functions (PDFs), together with the cumulative distribution function (CDF) on an area-fraction basis. Following Toth et al.

(Toth, Biswas, Gu, & Beausir, 2013), PDFs were used instead of raw histograms to ensure that the curves are comparable regardless of bin width, with the area under each PDF equal to 1.

3.9 Quantitative Microstructural and Recrystallization Analysis

The EBSD data generated from the rolled and annealed specimens required mathematical filtering and statistical treatment to accurately separate the recrystallized fractions and evaluate the stability of specific texture components. The analytical methodology remained consistent across all heat treatment conditions.

3.9.1 Recrystallization Partitioning

To investigate the onset and progression of recrystallization, grains were reconstructed using a grain-detection tolerance angle of 5° to better capture the evolution of the microstructure. Neighboring pixels with misorientations below this value were assigned to the same reconstructed grain. Grains smaller than three pixels were excluded from the analysis. Recrystallized grains were identified using a combined GOS and GOS/D criterion. Following the approach proposed in (A. Ayad et al., 2021), grains with $GOS < 2.5^\circ$ and $GOS(c)/D < 0.1^\circ/\mu m$ were classified as recrystallized in the present study. Grains that did not satisfy both criteria were classified as non-recrystallized. The thresholds were calibrated so that less than 1% of the grains in the fully deformed state were identified as recrystallized (A. Ayad et al., 2021). This calibration provided a reliable distinction between recrystallized regions and grains containing deformation substructures.

3.9.2 Retention and Resistance of Texture Components

To describe the microstructure evolution of specimens after near-complete recrystallization, two complementary measures were derived from the EBSD-based texture fractions and the distortion metrics. The retention index (RI) quantifies texture persistence by comparing the

weight fraction of each component in the non-recrystallized (NR) fraction with the rolled condition.

At each annealing condition, the NR fraction was calculated as:

$$f_{NR} = 1 - f_{REX} \quad (3-3)$$

where f_{REX} and f_{NR} are the recrystallized and non-recrystallized fractions, respectively.

For each texture component c , the measured fraction within the NR regions V_c^{NR} is weighted by the non-recrystallized fractions to quantify its retention:

$$V_{c,weighted}^{NR} = V_c^{NR} f_{NR} \quad (3-4)$$

The Retention Index (RI) of component c was then calculated as:

$$RI_c = \frac{V_{c,weighted}^{NR}}{V_c^{rolled}} \quad (3-5)$$

where V_c^{rolled} is the initial fraction of texture component c in the 60% cold-rolled reference before annealing. The rolled as-cast condition was used as the reference for the annealed as-cast specimens, while the rolled as-extruded condition was used for the annealed as-extruded specimens. Only components with an initial rolled fraction greater than 1.5% were considered to avoid unreliable RI values caused by very small initial fractions.

The NR substructure was described using a resistance score (RS) that combines the intragranular misorientation (GOS) and misorientation normalized by grain size $((GOS/D)_c)$. These values were normalized for a given annealing condition using the highest (max (x)) and

the lowest ($\min(x)$) GOS and $(GOS/D)_c$ values between S, RG, and Bs(I-II) using the following equation:

$$\tilde{x}_c = \frac{x_c - \min(x)}{\max(x) - \min(x)} \in , \quad x \in \{GOS, GOS/D\} \quad (3-6)$$

The resistance score (RS) of component c was then calculated as:

$$RS_c = \frac{1}{2} \overline{GOS}_c + \frac{1}{2} \left(\frac{\overline{GOS}}{D} \right)_c \quad (3-7)$$

RS_c lies in $[0,1]$ and ranks components by the stability of their NR state. A high value of the RS indicates a relatively highly misoriented substructure compared to the other NR components, whereas a low RS indicates a lower intragranular misorientation and a more uniform internal orientation state.

3.10 Secondary Particle Quantification and Phase Equilibrium Modeling

Second-phase particles were characterized using backscattered electron scanning electron microscopy (BSE-SEM) and energy-dispersive X-ray spectroscopy (EDS). BSE-SEM images were used to examine particle morphology and spatial distribution, while EDS elemental maps were used to identify the chemical characteristics of the observed particle populations.

For the pseudo-hot rolling study, at least three regions were examined for each selected condition to assess the general particle distribution, elemental enrichment, and directional arrangement. Representative BSE-SEM images and corresponding EDS maps are presented in Chapter 4.

For the as-extruded AA6082 specimens subjected to 60% unidirectional cold rolling (UCR) and subsequently annealed at 400 °C for selected times, at least three BSE-SEM/EDS regions were examined for each condition to verify the consistency of the observed particle distribution

and morphology. From these regions, one representative BSE-SEM field was selected for quantitative particle analysis using MATLAB R2021a and is presented in the corresponding results section. The particles were identified from their contrast with the aluminum matrix, followed by image filtering to improve particle detection. Particles intersecting the image boundaries were excluded from the analysis. The particle area fraction was calculated as the total segmented particle area divided by the analyzed BSE-SEM field area. Area-weighted particle parameters were calculated using the projected area of each particle as the weighting factor.

Particle aspect ratio was calculated according to the definition introduced in (3-2). The elongation ratio was defined as its reciprocal:

$$ER = \frac{\text{major-axis length}}{\text{minor-axis length}} = \frac{l}{R} \quad (3-8)$$

Thus, an aspect ratio approaching unity corresponds to a more equiaxed particle, whereas a larger elongation ratio indicates a more elongated morphology.

Particle orientation was measured from the fitted major axis relative to the horizontal image axis, corresponding to TD and the prior extrusion direction. Only particles with $AR \leq 0.80$ were included in the orientation analysis because the orientation of nearly round particles is not reliably defined. This filter was applied only to the orientation analysis.

The accepted orientations were grouped into 10° intervals. The particle number fraction in each interval was calculated relative to the total number of analyzed particles as

$$f_j = \frac{n_j}{N_{\text{orientation}}} \times 100 \quad (3-9)$$

where n_j is the number of accepted particles in interval j , and N_{total} is the total number of analyzed particles. The radial axis of the polar plots therefore represents the percentage of all analyzed particles within each angular interval.

The concentration of the accepted particle orientations was quantified using the area-weighted axial mean resultant length, R . Following the treatment of axial data by Mardia and Jupp (2000) (Mardia & Jupp, 1999), the measured orientations were doubled because θ and $\theta + 180^\circ$ represent the same particle axis. The normalized projected particle area was incorporated as the weighting factor:

$$R = \sqrt{\left[\sum_{i=1}^{N_{\text{orientation}}} w_i \cos(2\theta_i) \right]^2 + \left[\sum_{i=1}^{N_{\text{orientation}}} w_i \sin(2\theta_i) \right]^2} \quad (3-10)$$

where $N_{\text{orientation}}$ is the number of particles accepted for orientation analysis, θ_i is the signed major-axis orientation of particle i relative to TD, and w_i is the normalized projected-area weight, calculated as:

$$w_i = \frac{A_i}{\sum_{i=1}^{N_{\text{orientation}}} A_i} \quad (3-11)$$

where A_i is the projected area of particle i . The mean resultant length, R , ranges from 0 to 1. A value approaching 1 indicates a highly concentrated orientation distribution, meaning that the particle major axes are preferentially aligned along a common direction. In contrast, a value approaching 0 indicates a highly dispersed orientation distribution with no dominant axial direction.

In parallel, the temperature-dependent equilibrium phase distribution in AA6082 and the corresponding theoretical precipitate volume fractions were predicted via thermodynamic

calculations in FactSage 8.4, using the FTlite and FactPS databases for aluminum alloy systems.

3.11 Mechanical Characterization and Digital Image Correlation (DIC)

To relate the different thermomechanical processing histories to the macroscopic anisotropic behavior of the alloys, microtensile tests were carried out on selected conditions corresponding to 40% and 90% reduction. The tests were performed at room temperature using a microtensile testing machine in order to determine the ultimate tensile strength (UTS), the strain hardening exponent (n-value), and the plastic strain ratio (R-value). The tensile specimens were extracted from the rolled plates at various orientations (e.g., 0°, 45°, and 90°) with respect to the rolling direction (Figure 3-7a), and machined according to the sub-sized geometry detailed in Figure 3-7b. To ensure statistical reliability and reproducibility, testing was performed in duplicate for each orientation and experimental condition. Prior to testing, the specimen surface was prepared for DIC measurements by applying a speckle pattern using a boron nitride spray to ensure sufficient contrast for accurate displacement tracking.

To achieve high-precision strain measurements and capture localized deformation, the microtensile machine was equipped with a Digital Image Correlation (DIC) system (Figure 3-7c). This non-contact optical technique enabled continuous tracking of two-dimensional displacement and strain fields over the specimen surface during deformation.

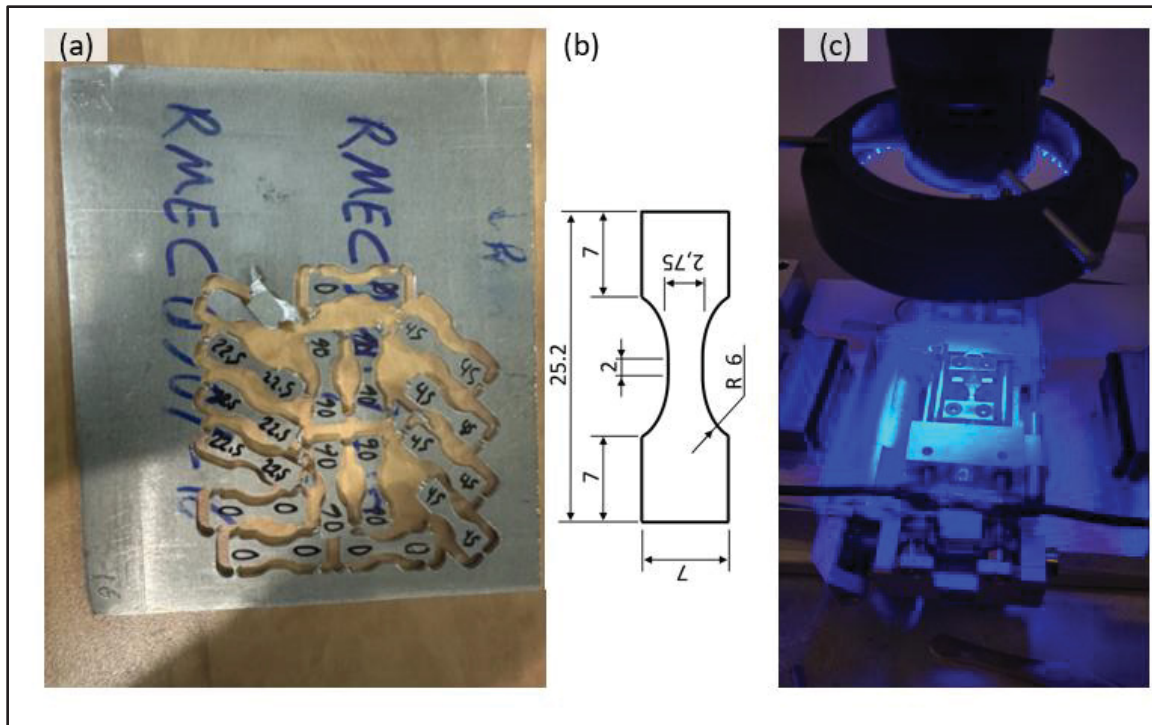


Figure 3-7 Mechanical testing setup and specimen details: (a) rolled aluminum plate illustrating the extraction layout of microtensile specimens at various orientations relative to the rolling direction; (b) schematic geometry and dimensions (in mm) of the sub-sized tensile specimens; and (c) the microtensile testing apparatus equipped with the Digital Image Correlation (DIC) camera and illumination system

3.11.1 DIC Data Acquisition and Processing Workflow

During each tensile test, the DIC camera system captured high-resolution surface images at a frequency of 1 Hz (one frame per second) from the onset of loading until the ultimate fracture or the post-necking regime where the applied force dropped significantly.

The recorded images were subsequently analyzed using a structured data processing procedure to synchronize the optical strain measurements with the mechanical data:

Masking and setup: The region of interest on the specimen surface was defined using ImageJ (FIJI).

Strain field extraction: DIC algorithms were applied to track the surface speckle pattern and compute the two-dimensional displacement fields.

Data synchronization: The displacement data were combined with the corresponding force measurements and time records from the testing system. This allowed the calculation of true stress and true strain throughout the deformation process.

3.11.2 Calculation of Anisotropy Parameters

The synchronized data provided continuous measurements of the longitudinal true strain (ϵ_{xx}) and the transverse width true strain (ϵ_{yy}). Due to the physical limitations of optically measuring the instantaneous thickness reduction during the test, the principle of volume constancy during plastic deformation ($\epsilon_{xx} + \epsilon_{yy} + \epsilon_{zz} = 0$) was applied. The true thickness strain (ϵ_{zz}) was calculated as follows:

$$\epsilon_{zz} = -(\epsilon_{xx} + \epsilon_{yy}) \quad (3-12)$$

This allowed for the plotting of highly accurate true stress-strain curves to extract the UTS and strain hardening (n) parameters. To evaluate the planar anisotropy and sheet formability, the Lankford coefficient (R-value) was calculated. For this study, the R-value was evaluated precisely at a longitudinal true strain of 5% ($\epsilon_{xx} = 0.05$), utilizing the following relationship between the width and thickness strains:

$$R_{5\%} = \frac{\epsilon_{yy}}{\epsilon_{zz}} \quad (3-13)$$

CHAPTER 4

MICROSTRUCTURE AND TEXTURE EVOLUTION UNDER ROLLING CONDITIONS

4.1 Introduction

This section presents the experimental results obtained in this work and discusses the microstructural and texture evolution of the investigated alloys under different thermomechanical conditions. The chapter is organized progressively, beginning with the characterization of the initial materials, followed by the analysis of the deformation processes during cold and Pseudo-hot rolling.

First, the initial states of the materials (Section 4.2) are examined, including the as-cast and as-extruded conditions of AA6082 as well as the as-cast AA6063 alloy. The microstructure and crystallographic texture of these initial materials are characterized in order to establish the reference conditions for the subsequent deformation processes.

The deformation study (Section 4.3) begins with cold rolling of the as-cast AA6082 alloy, followed by cold rolling of the as-extruded AA6082 to evaluate the influence of the initial microstructure and texture on the deformation behavior. The same analysis is then extended to as-cast AA6063 to enable comparison between the two alloys under similar rolling conditions.

After the investigation of cold rolling, the study continues with Pseudo-hot rolling (Section 4.4) performed on the aforementioned initial materials. The purpose of this part is to evaluate the effect of deformation temperature on the microstructure, texture evolution, and the development of typical FCC rolling components.

4.2 Initial Materials

The initial microstructure and texture of the as-cast AA6082 and AA6063 alloys are shown in Figure 4-1 through EBSD orientation maps and corresponding $\{111\}$ pole figures. Both alloys exhibit an equiaxed grain structure typical of cast aluminum alloys, with no evidence of deformation-induced elongation or preferred grain alignment. The grain size appears relatively uniform in both materials, although AA6063 shows slightly finer grains compared to AA6082. The $\{111\}$ pole figures indicate that the crystallographic texture in the as-cast condition is weak and relatively scattered, with low maximum intensities of about 2.6 for AA6082 and 2.0 for AA6063. This low texture intensity confirms the absence of strong preferred orientation, which is consistent with the solidification-controlled microstructure of cast materials. Thereby, both alloys start from a nearly random texture state, providing a suitable reference condition for evaluating the texture evolution during subsequent deformation processes.

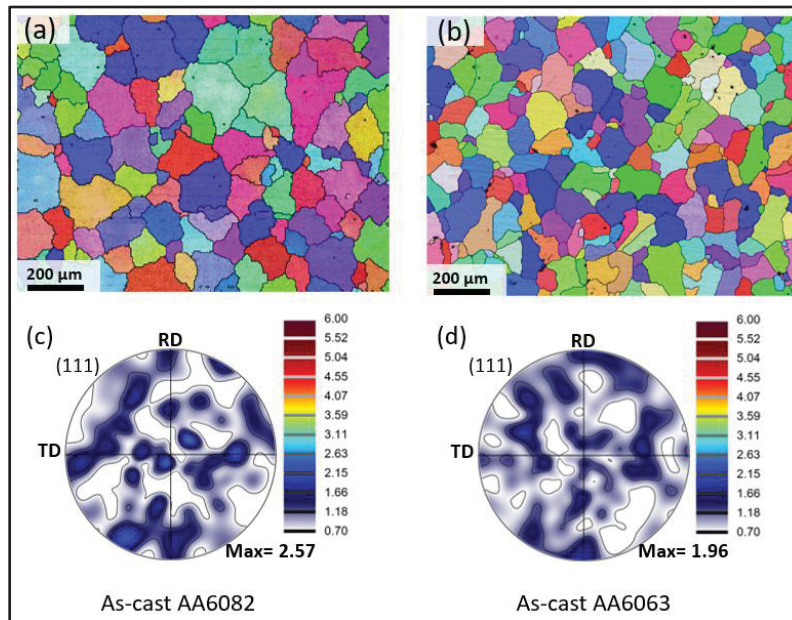


Figure 4-1 EBSD orientation maps of (a) as-cast AA6082 and (b) as-cast AA6063, and the corresponding $\{111\}$ pole figures illustrating the crystallographic texture of (c) AA6082 and (d) AA6063 in the as-cast condition

The initial microstructure and crystallographic texture of the as-extruded AA6082 condition are shown in Figure 4-2 through the EBSD orientation map, the $\{111\}$ pole figure, and the

inverse pole figures along ED, TD, and ND, and is further quantified in Table 4-1. The microstructure exhibits the typical fibrous morphology commonly observed in extruded 6xxx series aluminum alloys, together with a duplex fiber texture in which crystallographic directions close to $\langle 111 \rangle$ and $\langle 100 \rangle$ are preferentially aligned with the extrusion direction. A pronounced texture concentration is observed in the ED direction, where the maximum intensity reaches 12.7 near the $\langle 111 \rangle$ pole, indicating a strong alignment of crystallographic directions along the extrusion axis. In contrast, the TD and ND inverse pole figures show more diffuse intensity distributions, with lower maximum values of 3.3 and 3.1, respectively, reflecting a weaker orientation preference in directions perpendicular to the extrusion axis. This strong extrusion-induced texture is consistent with the component fractions listed in Table 4-1, where the initial condition is dominated by RCu (35%), together with notable contributions from Bs (31%), Cb (16%), and RGs (5%), while the S and Cu components remain negligible.

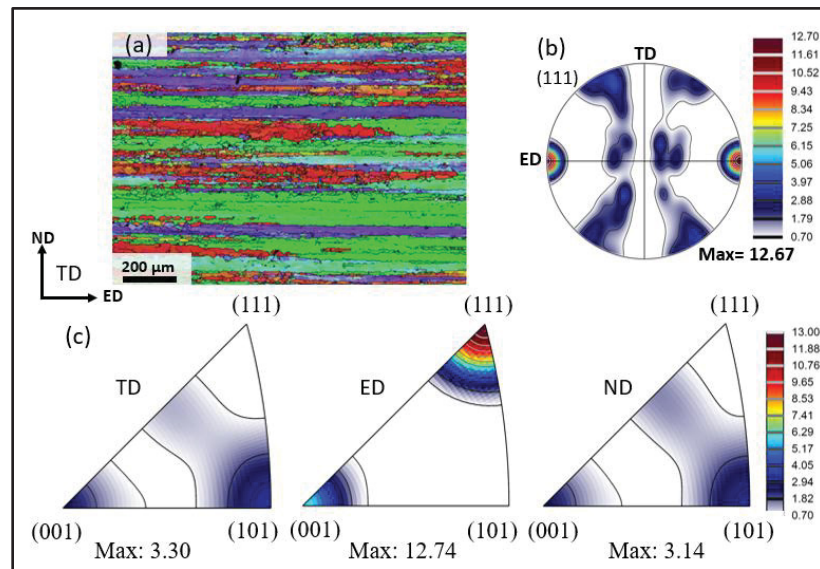


Figure 4-2 (a) EBSD orientation map of the as-extruded AA6082 observed on the ED–TD plane. (b) $\{111\}$ pole figure showing the crystallographic texture of the as-extruded condition. (c) Inverse pole figures along TD, ED, and ND illustrating the orientation distribution with respect to the extrusion reference frame

Table 4-1 Volume fractions of selected texture components in as-extruded AA6082

Sample	S	Bs	Cu	Cb	Gs	RGs	RCu	Gs/Bs
AA6082-As ext	<1	31	0	16	0	5	35	<1

4.3 Cold Rolling

4.3.1 Cold Rolling Results

4.3.1.1 Cold Rolled As-cast AA6082

Unidirectional cold rolling (UCR)

Progressive cold rolling of as-cast AA6082 under UCR results in pronounced microstructural evolution, as revealed by the band contrast and EBSD orientation maps (Figure 4-3 and Figure 4-4). At lower thickness reductions (10–30%), the EBSD maps show clearly indexed, equiaxed grains with distinct orientations. The corresponding band-contrast images exhibit high contrast levels, clearly revealing the coexistence of LAGBs and HAGBs.

As the deformation increases to 40%, 50%, and 60%, the EBSD map quality decreases and the grain boundaries become more blurred. The band contrast maps also show a reduction in image sharpness and contrast. Furthermore, the increasing boundary density is reflected by the rise in total boundary length, from 2.21×10^5 in the as-cast condition to 3.64×10^5 μm at 60% reduction (Figure 4-5). Although elongation along the rolling direction is not visible in the TD–ND cross-sectional view, grain flattening along the normal direction becomes more pronounced with increasing reduction.

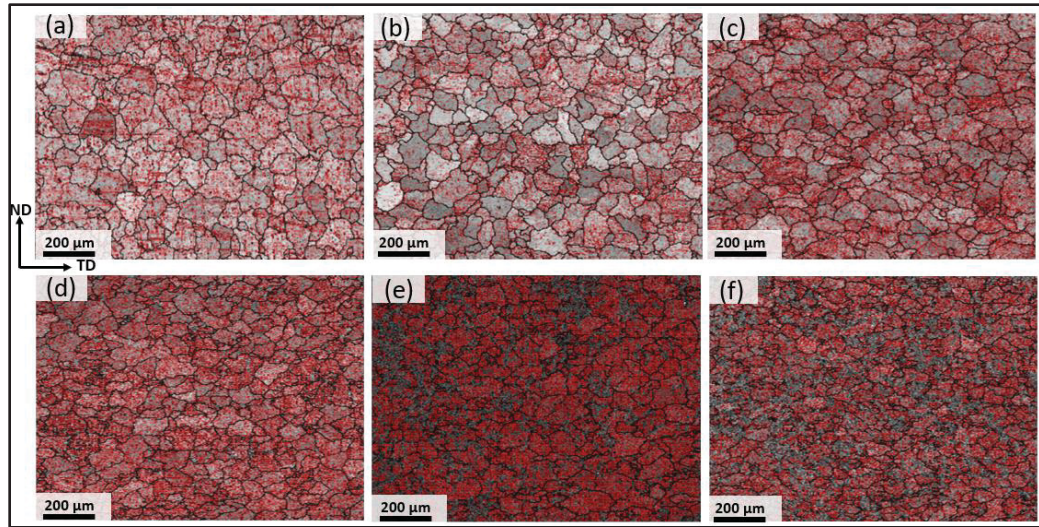


Figure 4-3 Band contrast maps of as-cast AA6082 subjected to unidirectional cold rolling (UCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction (TD–ND plane). Grain boundaries with misorientations between 2° and 15° are shown in red (low-angle boundaries), while those above 15° (high-angle boundaries) are shown in black

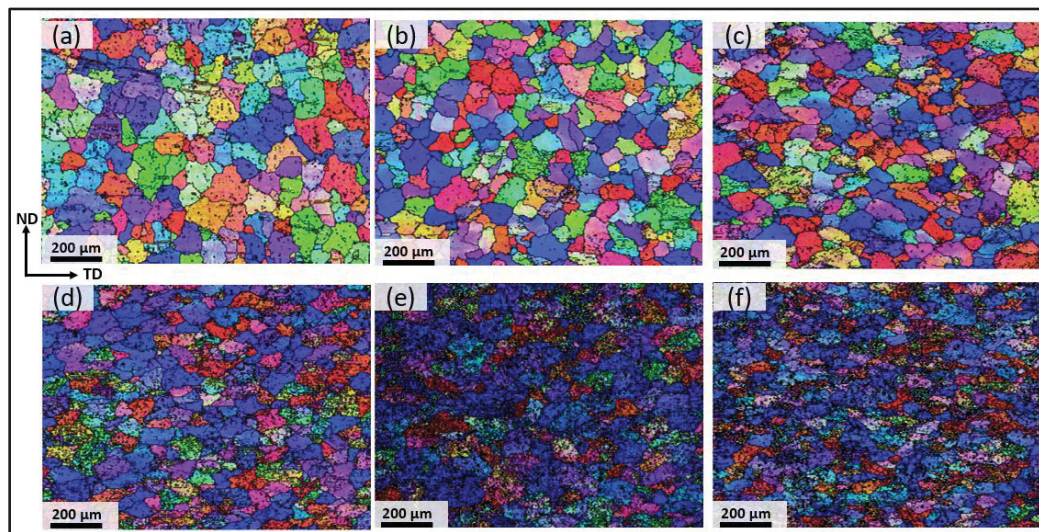


Figure 4-4 EBSD orientation maps of as-cast AA6082 subjected to unidirectional cold rolling (UCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction

The evolution of the grain-boundary misorientation distributions for selected UCR conditions is presented in Figure 4-5. The as-cast condition is strongly dominated by LAGBs, which

account for 86 % of the total boundary population and produce a sharp peak at misorientation angles below 5° . With increasing reduction, the distributions remain concentrated in the low-angle region, but the main peak gradually decreases in intensity and the distributions become broader. The HAGB fraction increases progressively from 14% in the as-cast condition to 15% at 20% reduction, 17% at 40% reduction, and 22% at 60% reduction, while the corresponding LAGB fraction decreases accordingly. These changes indicate the progressive development of deformation-induced substructures and the gradual transformation of low-angle boundaries toward higher misorientation angles.

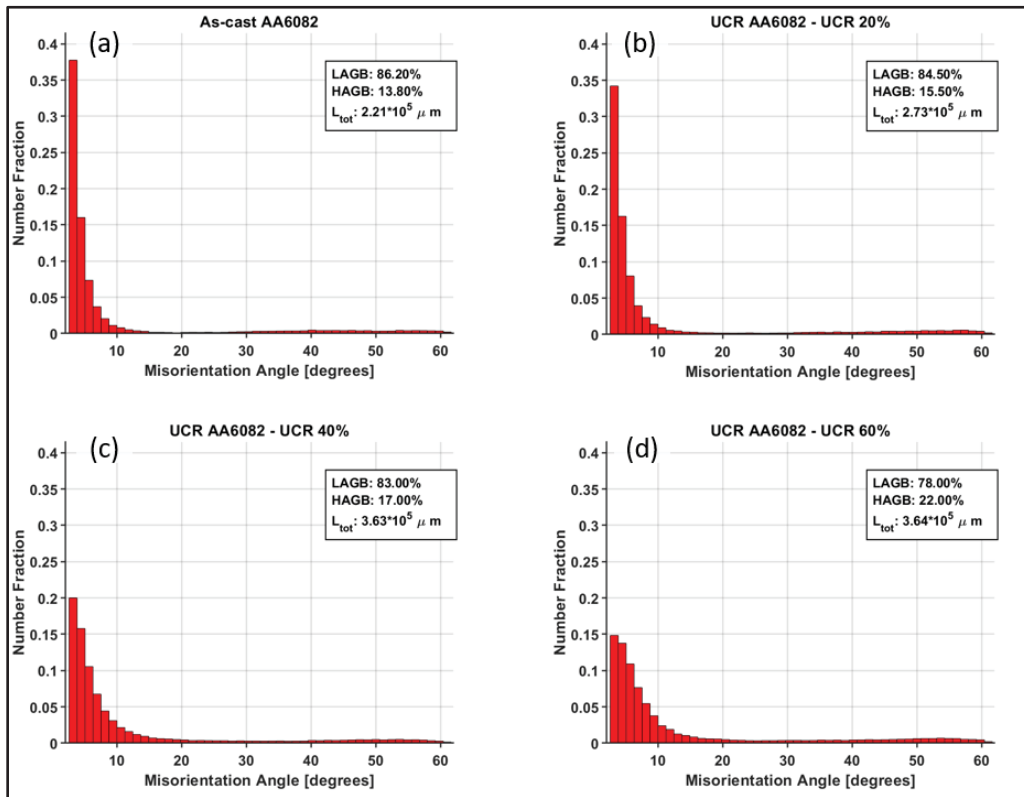


Figure 4-5 Misorientation angle distributions for as-cast AA6082 and selected conditions under unidirectional cold rolling (UCR): (a) as-cast condition, (b) UCR 20%, (c) UCR 40%, and (d) UCR 60%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

The evolution of deformation texture in as-cast AA6082 during UCR is clearly demonstrated in the $\{111\}$ pole figures (Figure 4-6). At low reductions (10–30%), the pole figures exhibit

weak and dispersed texture components with low maximum intensity values (Max $\sim$ 2–2.4), indicating a near-random orientation distribution. As the rolling reduction increases to 40%, 50%, and 60%, the texture becomes progressively sharper and more concentrated along specific crystallographic directions. At 50% and especially 60% reduction, strong β -fiber components aligned parallel to the rolling direction (RD) become evident, with intensity values increasing up to Max = 4.2.

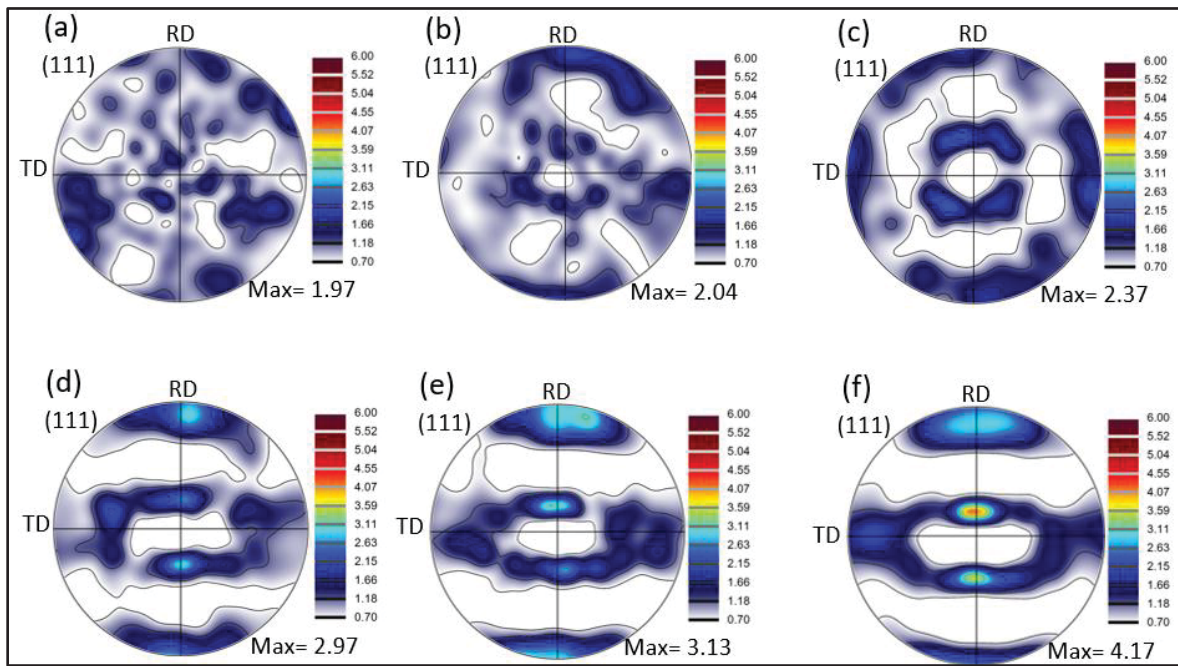


Figure 4-6 $\{111\}$ pole figures of as-cast AA6082 after unidirectional cold rolling (UCR) at thickness reductions of (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%

To further examine the microstructure at severe deformation, EBSD characterization was performed at 90% reduction, as shown in Figure 4-7. Because the microstructure becomes highly refined at this strain level, the EBSD maps were acquired at a higher magnification ($500\times$) compared with the lower reductions. The EBSD orientation map (Figure 4-7a) shows a highly refined banded structure with pronounced grain subdivision and compression along the normal direction. The corresponding band contrast map (Figure 4-7b) reveals a dense boundary network and pronounced substructure formation, with a high density of LAGBs. The associated $\{111\}$ pole figure (Figure 4-7c) indicates a further strengthening of the deformation

texture, with a maximum intensity of approximately 5.1, confirming the continued development of the β fiber texture at the highest strain level.

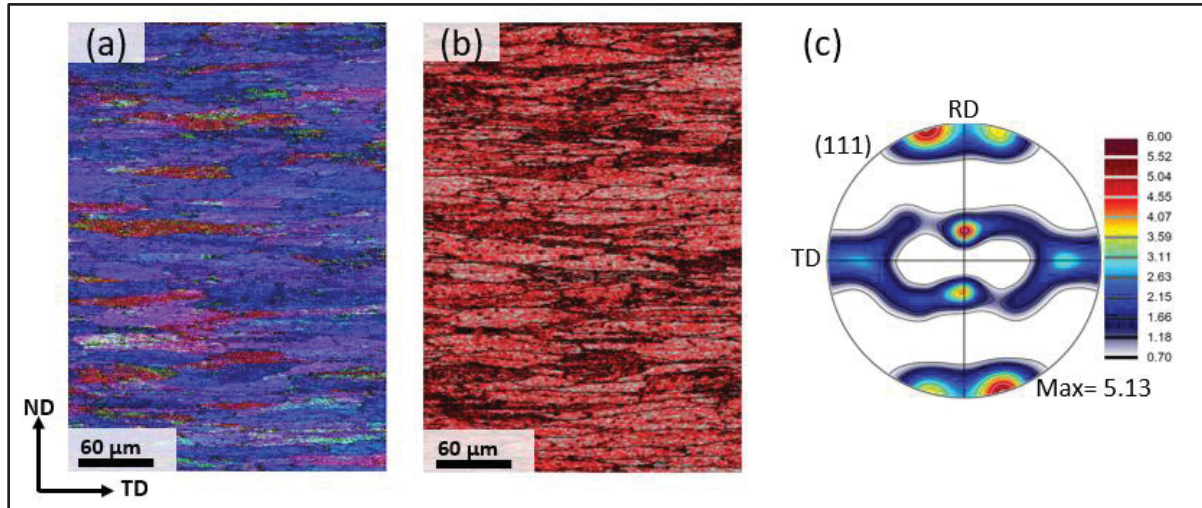


Figure 4-7 High-magnification EBSD characterization of as-cast AA6082 after 90% unidirectional cold rolling (UCR): (a) EBSD orientation map, (b) band contrast map, and (c) $\{111\}$ pole figure

The rolling-texture evolution in as-cast AA6082 is further described through the volume fractions of the main FCC texture components (Figure 4-8). At low reductions (10–30%), the texture remains relatively weak and is distributed among the classical deformation components (S, Cu, and Bs), while recrystallization-related orientations remain limited. With increasing reduction, the S component strengthens progressively, rising from ~10% at 10% reduction to ~20% at 40%, ~24% at 50%, ~26% at 60%, and reaching 45% at 90% reduction. The Cu component also increases from ~5% at 10% to ~11% at 40%, remains at a similar level at 60%, and then rises further to 16% at 90% reduction. In parallel, the Bs component increases from ~3% at 10% reduction to ~13% at 60%, and further to 17% at 90% reduction. By contrast, the components such as RCu decrease from ~5% at 10% reduction to less than 1% at higher reductions. So, these trends indicate a strong shift toward deformation-dominated FCC rolling texture components with increasing strain, with the highest reduction level characterized by the clear dominance of the S component, accompanied by substantial contributions from Bs and Cu.

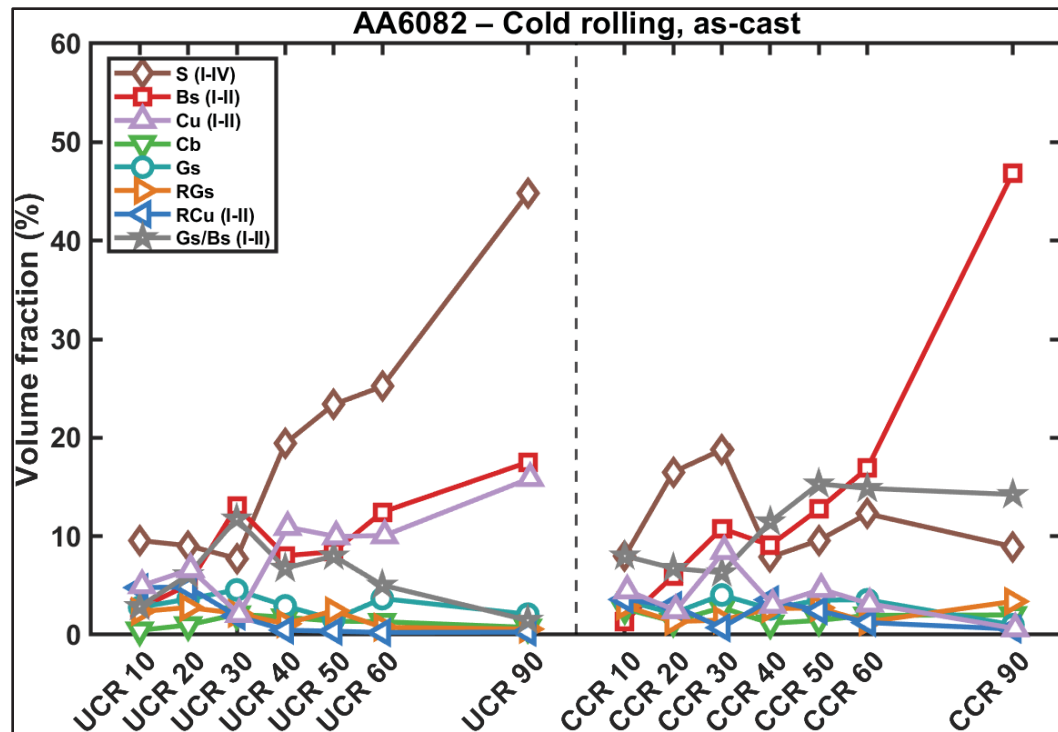


Figure 4-8 Volume fraction of selected texture components in as-cast AA6082 as a function of thickness reduction under unidirectional and cross cold rolling

The evolution of the grain size distribution for as-cast AA6082 during UCR reveals a consistent shift toward a more refined microstructure as deformation increases. The cumulative area fraction curves (Figure 4-9) demonstrate a clear migration toward the left, with the distributions becoming significantly steeper, which indicates a transition from a coarse, heterogeneous as-cast state to a more uniform, fine-grained structure. Quantitatively, the as-cast maximum grain size of 222 μm is progressively reduced to 132 μm at 40% UCR and ultimately to 57 μm at 90% UCR. This refinement is further highlighted by the 85% cumulative size fraction, which drops sharply from 183 μm in the initial state to 36 μm at the highest reduction level (90%).

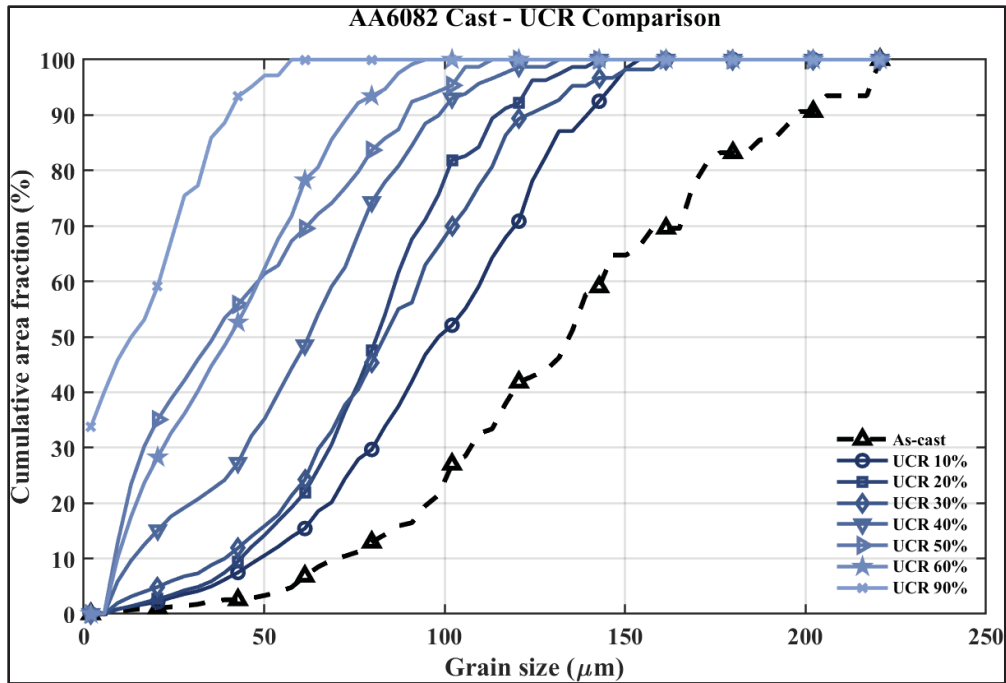


Figure 4-9 Cumulative grain size distribution curves for as-cast AA6082 under unidirectional cold rolling (UCR) at various thickness reductions

Cross cold rolling (CCR)

The microstructural evolution of as-cast AA6082 under CCR is presented in the band contrast (Figure 4-10) and EBSD orientation maps (Figure 4-11). Compared to the UCR condition, cross rolling introduces a multidirectional strain path, which significantly affects grain boundary development and orientation evolution. In the EBSD maps, grains remain relatively equiaxed with limited elongation up to 30% reduction, while at higher reductions (40–60%), gradual orientation alignment and the formation of internal misorientations become more apparent. The band contrast maps display a progressive decrease in image contrast and indexing quality with increasing deformation, particularly at 50% and 60% reduction. Similar to the UCR condition, the density of internal boundaries increases with increasing deformation. This evolution is reflected in the increase in total boundary length, which rises from approximately $2.22 \times 10^5 \mu\text{m}$ at 20% reduction to $3.64 \times 10^5 \mu\text{m}$ at 60% reduction (Figure 4-12). The alternating strain path in cross rolling mitigates grain elongation due to rotation at

each pass, and the effects of this rotation are observable in this TD–ND cross-sectional view (RD1).

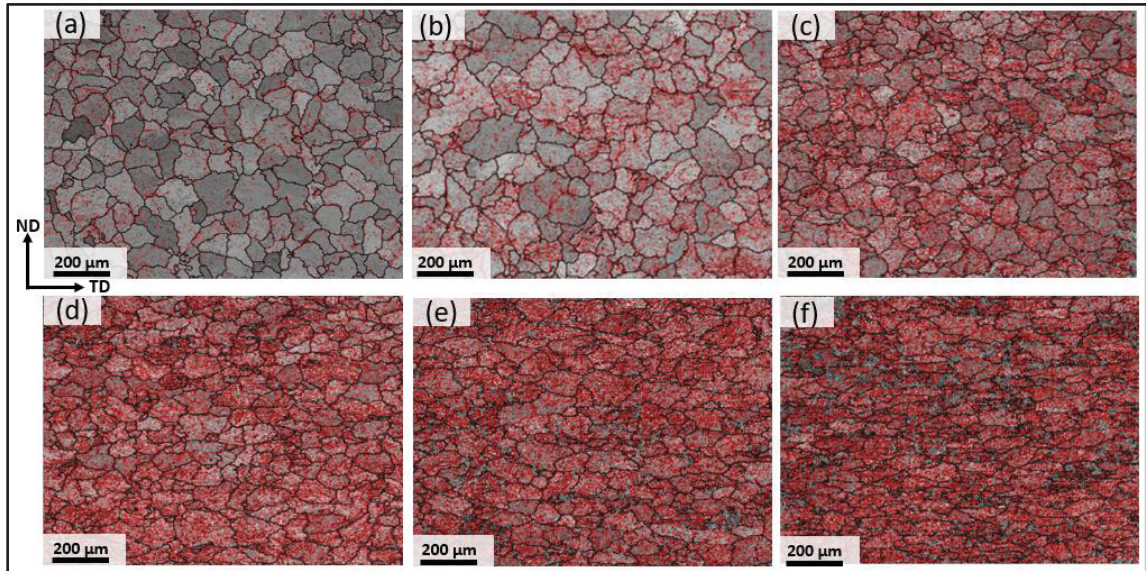


Figure 4-10 Band contrast maps of as-cast AA6082 subjected to Cross cold rolling (CCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction (RD1) - TD–ND plane

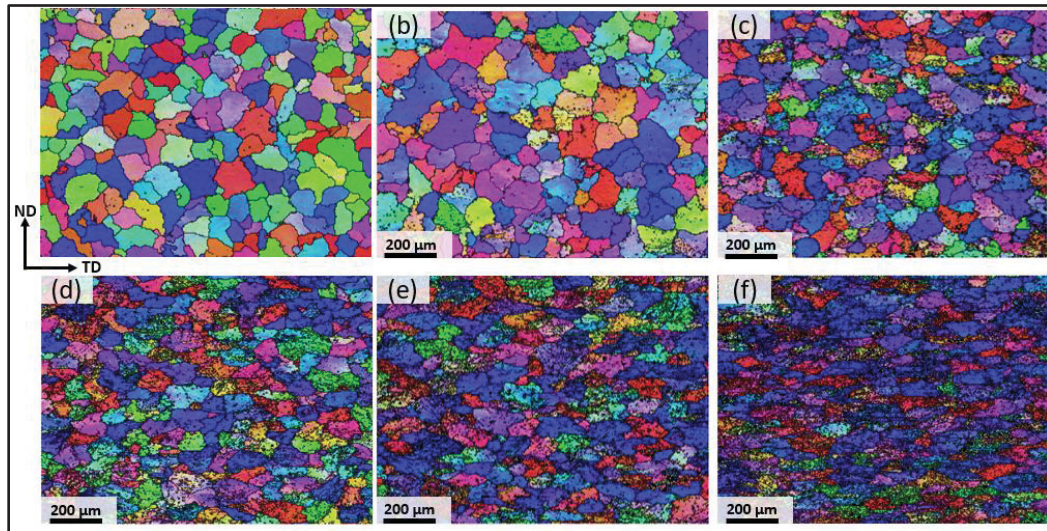


Figure 4-11 EBSD orientation maps of as-cast AA6082 subjected to Cross cold rolling (CCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction (RD1) - TD–ND plane

The evolution of the grain-boundary misorientation distributions for selected CCR conditions is presented in Figure 4-12. Similar to the UCR condition, the misorientation distributions remain strongly concentrated in the low-angle region. With increasing reduction during CCR, the distributions broaden gradually and the fraction of HAGBs increases. At 20% reduction, the HAGB fraction rises to 17%, while the LAGB fraction decreases to 83%. At 40% reduction, the HAGB fraction remains near 16%, and at 60% reduction it increases further to 22%, accompanied by a decrease in LAGBs to 78%. In parallel, the total boundary length increases significantly, reaching $3.64 \times 10^5 \mu\text{m}$ at 60% reduction. These trends indicate the progressive development of deformation-induced substructures and the gradual rotation of sub-boundaries toward higher misorientation angles during cross rolling.

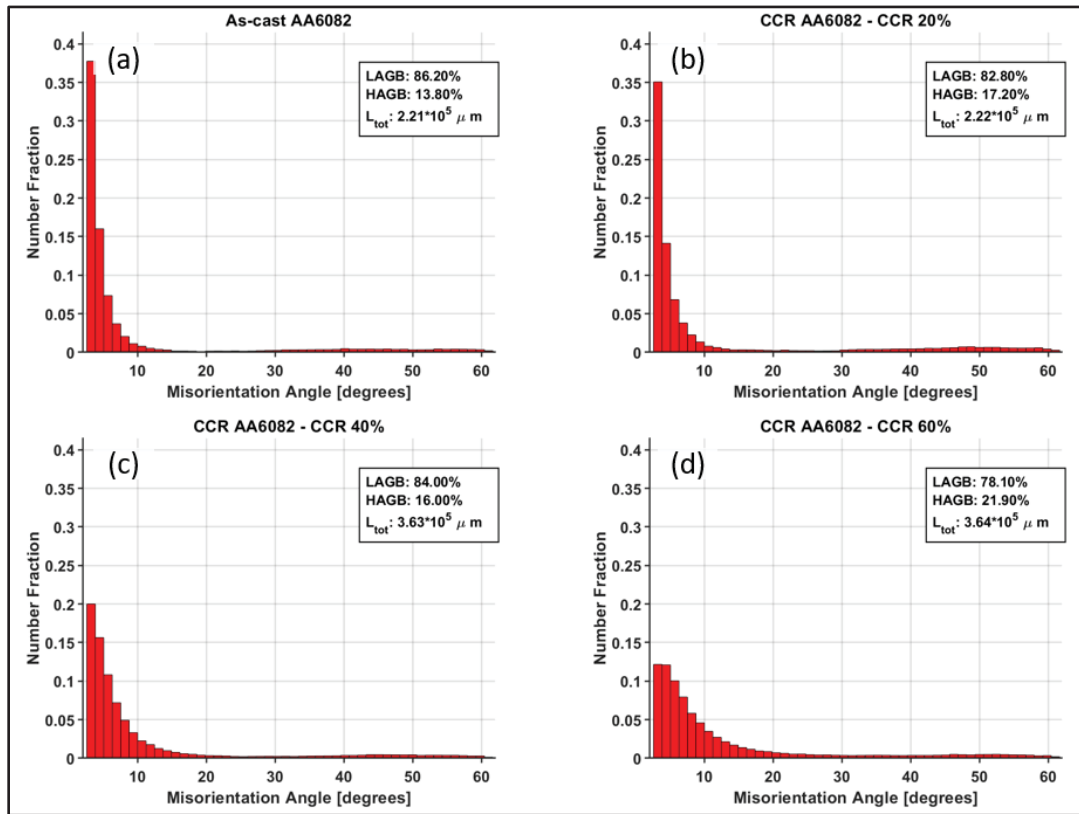


Figure 4-12 Misorientation angle distributions for as-cast AA6082 and selected conditions under cross cold rolling (CCR): (a) as-cast condition, (b) CCR 20%, (c) CCR 40%, and (d) CCR 60%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

The $\{111\}$ pole figures of as-cast AA6082 after CCR (Figure 4-13) illustrate the evolution of crystallographic texture as a function of thickness reduction. At low strains (10%), the texture is weak and relatively random (Max intensity ~ 2), indicating a lack of preferred orientation. As reduction increases to 20% and 30%, initial deformation features emerge, reaching a local intensity peak at 20% (Max ~ 3). However, between 40% and 60% reduction, the intensity plateaus (ranging from 2.2 to 2.8) and the orientations appear broadly spread across the pole figure. This behavior confirms that the alternating strain path in cross-rolling effectively interrupts the continuous buildup of the standard rolling fiber, maintaining a more balanced orientation distribution compared to the UCR (Figure 4-6) at similar intermediate strains.

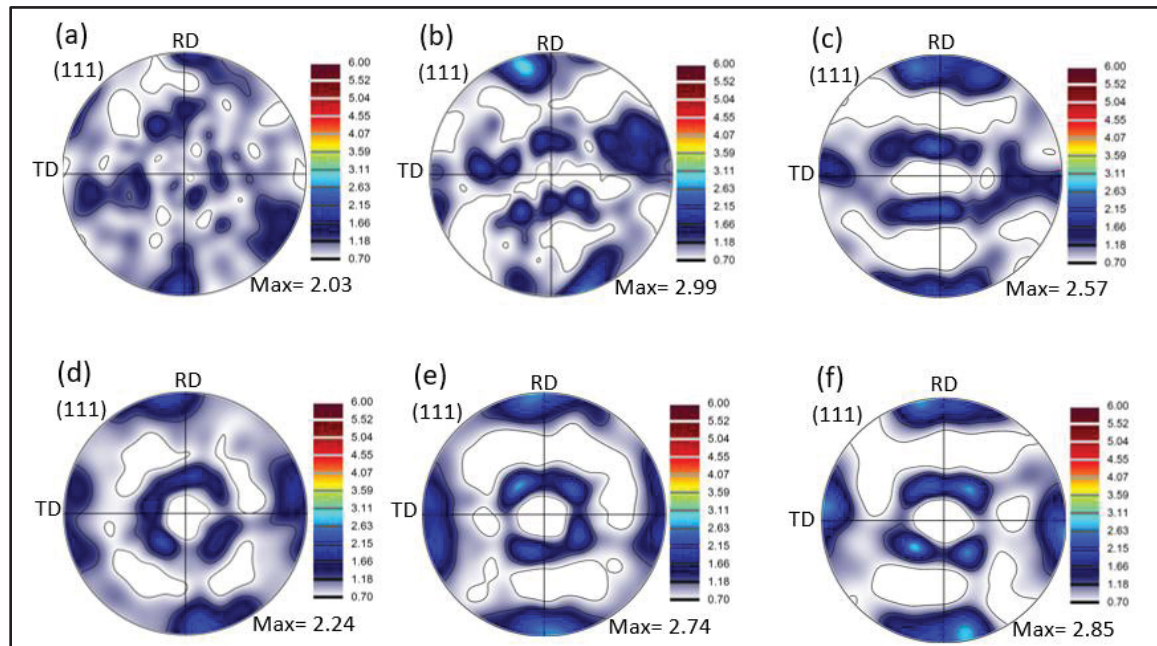


Figure 4-13 $\{111\}$ pole figures of as-cast AA6082 after Cross cold rolling (CCR) at thickness reductions of (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%

To evaluate the microstructure under extreme deformation, EBSD characterization was performed at 90% reduction (Figure 4-14). Due to the significant grain refinement and high dislocation density at this strain level, maps were acquired at higher magnification to resolve the internal structure. The orientation map (Figure 4-14a) reveals a highly fibrous microstructure characterized by fine deformation bands and significant internal misorientation. The corresponding band contrast map (Figure 4-14b) shows a dense network of boundaries

and pronounced grain subdivision. Notably, the associated $\{111\}$ pole figure (Figure 4-14c) exhibits a maximum intensity of approximately 6.0. This high value indicates that while CCR suppresses texture intensification at intermediate strains, extreme deformation eventually triggers a sharp, localized crystallographic alignment.

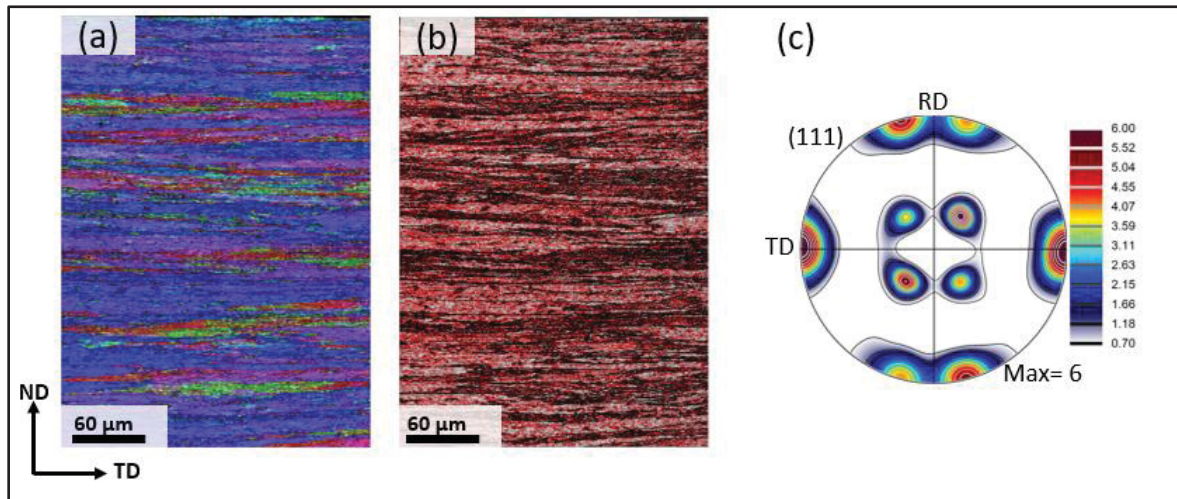


Figure 4-14 High-magnification EBSD characterization of as-cast AA6082 after 90% cross cold rolling (CCR): (a) EBSD orientation map, (b) band contrast map, and (c) $\{111\}$ pole figure

The evolution of rolling texture in as-cast AA6082 under CCR is further described through the volume fractions of the main FCC texture components (Figure 4-8). At low reductions (10–20%), the texture is partitioned among several weak components without a clear dominant orientation. With increasing strain, a divergence between the S and Brass (Bs) components becomes evident. While the S component in CCR fluctuates and ultimately plateaus (remaining below 13% at intermediate and high strains), the Bs component increases progressively.

By 60% reduction, the Bs component reaches approximately 17%, but at 90% reduction, it undergoes a dramatic intensification to approximately 47%. In contrast, the Cu and RCu components remain negligible throughout the process. This signifies a fundamental shift in the rolling texture: whereas UCR is dominated by the S-component of the β -fiber at high reductions, the 90% CCR condition is defined by a transition to a singular, dominant Brass-

type texture. This transition suggests that the cross-rolling path eventually favors the stability of the Brass orientation over the S-orientation at severe plastic strains.

The evolution of the grain size distribution for as-cast AA6082 during CCR indicates progressive refinement with increasing deformation. The cumulative area fraction curves (Figure 4-15) shift progressively toward smaller grain sizes as the rolling reduction increases. This leftward shift is accompanied by steeper curves, indicating the development of a finer and more uniform grain structure. Quantitatively, the maximum grain size decreases from 222 μm in the as-cast condition to 163 μm at 40% reduction, 118 μm at 60% reduction, and 67 μm at 90% reduction. A similar trend is observed for the 85% cumulative grain size, which decreases from 183 μm in the as-cast state to 102 μm at 40% reduction, 77 μm at 60% reduction, and 51 μm at 90% reduction. These values show the progressive reduction of the coarse as-cast grain structure as deformation increases. Compared with UCR, the grain size evolution under CCR follows the same overall refinement trend, but the refinement remains less pronounced across the deformation range. At comparable reductions, the CCR condition generally retains larger maximum and cumulative grain-size values than UCR, indicating that grain fragmentation proceeds more gradually under the cross-rolling strain path.

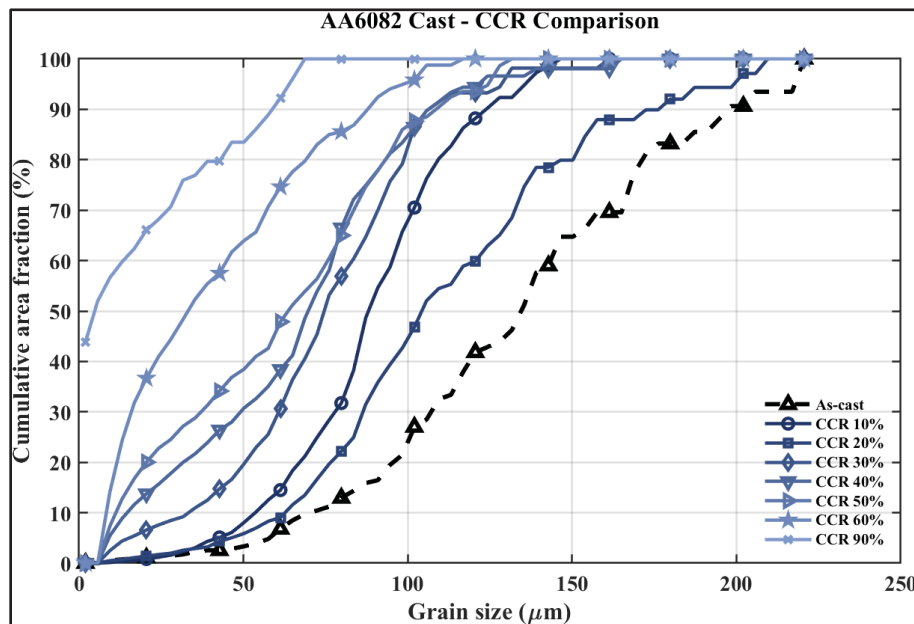


Figure 4-15 Cumulative grain size distribution curves for as-cast AA6082 under cross cold rolling (CCR) at various thickness reductions

4.3.1.2 Cold Rolled As-extruded AA6082

Unidirectional cold rolling (UCR)

The microstructural evolution of as-extruded AA6082 under UCR, perpendicular to the initial extrusion direction, is illustrated in the band contrast and EBSD orientation maps (Figure 4-16 and Figure 4-17). Since the sections are taken perpendicular to the rolling direction (TD–ND plane), grain elongation along the rolling direction is not directly visible. Instead, what becomes evident is the progressive thinning of deformation bands (fibrous grains) with increasing reduction, reflecting the compressive nature of cold rolling along the normal direction. The EBSD maps (Figure 4-17) show a transition from a broad color distribution at 10% to a more saturated pattern of purple-toned bands at higher reductions, indicating the development of preferred orientations. The band contrast maps (Figure 4-16) reveal a gradual reduction in indexing quality, especially beyond 40% reduction, though less severe than in the as-cast condition. The dominance of red-colored boundaries at higher reductions indicates an increasing density of LAGBs due to progressive substructure formation.

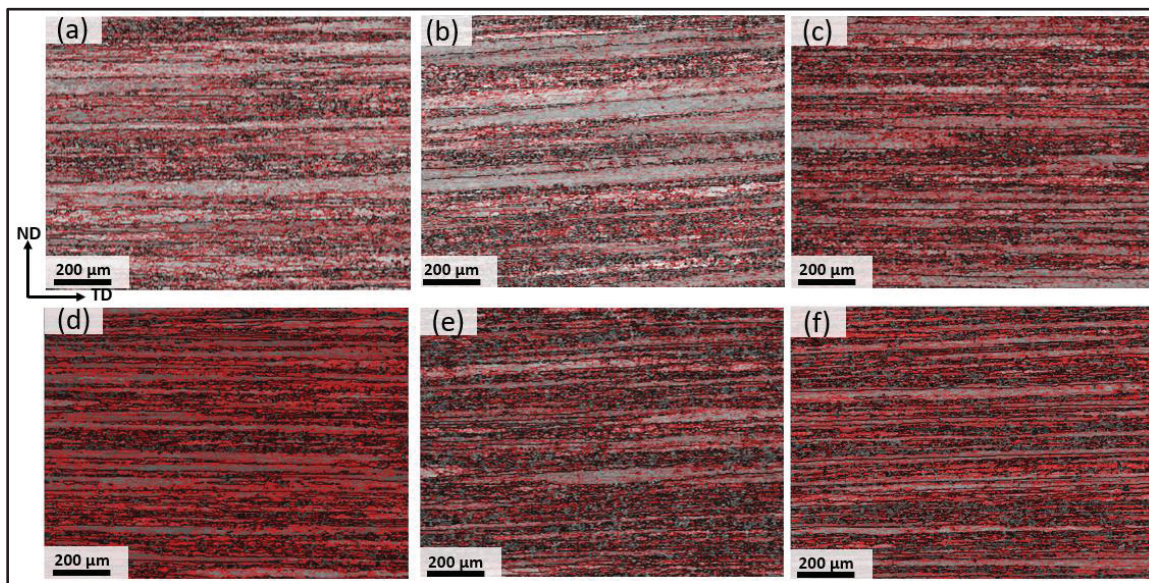


Figure 4-16 Band contrast maps of as-extruded AA6082 subjected to unidirectional cold rolling (UCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction

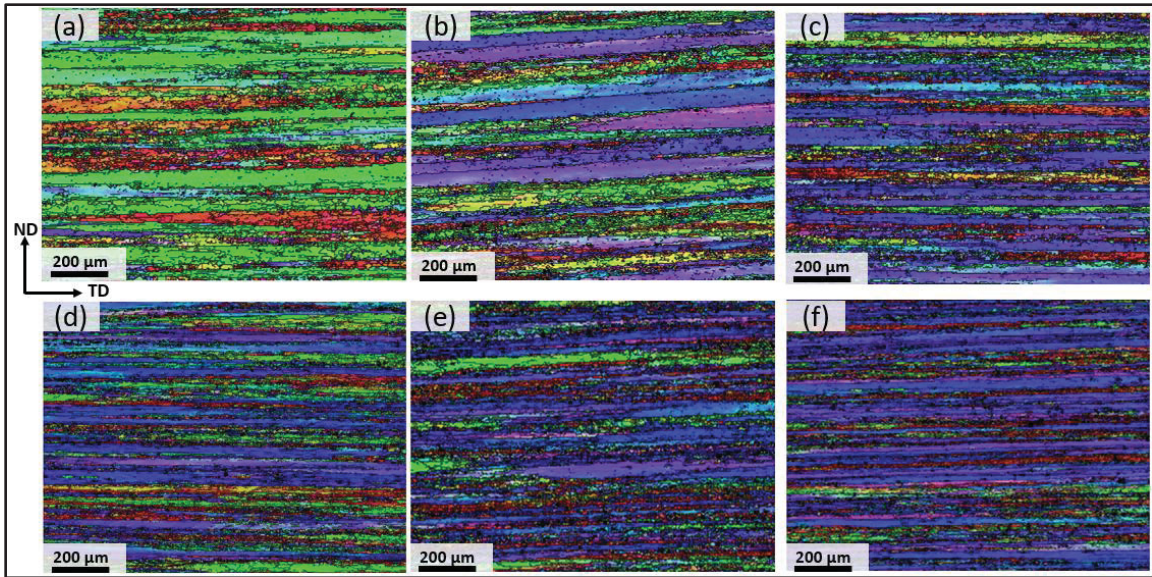


Figure 4-17 EBSD orientation maps of as-extruded AA6082 subjected to Unidirectional cold rolling (UCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction

The evolution of the grain-boundary misorientation distributions for rolled as-extruded AA6082 under UCR is presented in Figure 4-18. The as-extruded condition shows a mixed boundary character, with LAGBs representing 66% of the total boundary population and HAGBs accounting for 34%, resulting in a broader distribution than that of the as-cast material. With increasing thickness reduction, the misorientation distributions remain concentrated in the low-angle region, but the main peak at small angles becomes more pronounced and the overall distribution gradually broadens toward higher misorientation angles. At 20% reduction, the LAGB fraction increases to 76% and the HAGB fraction decreases to 24%, reflecting the strong increase in the proportion of LAGBs during the initial stage of deformation. At 40% reduction, the LAGB fraction decreases to 70% and the HAGB fraction increases to 30%, and at 60% reduction the HAGB fraction rises further to 32% while the LAGB fraction decreases to 68%. In parallel, the total boundary length increases significantly from $2.31 \times 10^5 \mu\text{m}$ in the as-extruded condition to $2.66 \times 10^5 \mu\text{m}$ at 20% reduction and reaches approximately $3.70 \times 10^5 \mu\text{m}$ at 40–60% reduction. These results indicate the progressive subdivision of the original grains into smaller domains and the continuous development of deformation-induced boundaries with increasing cold-rolling reduction.

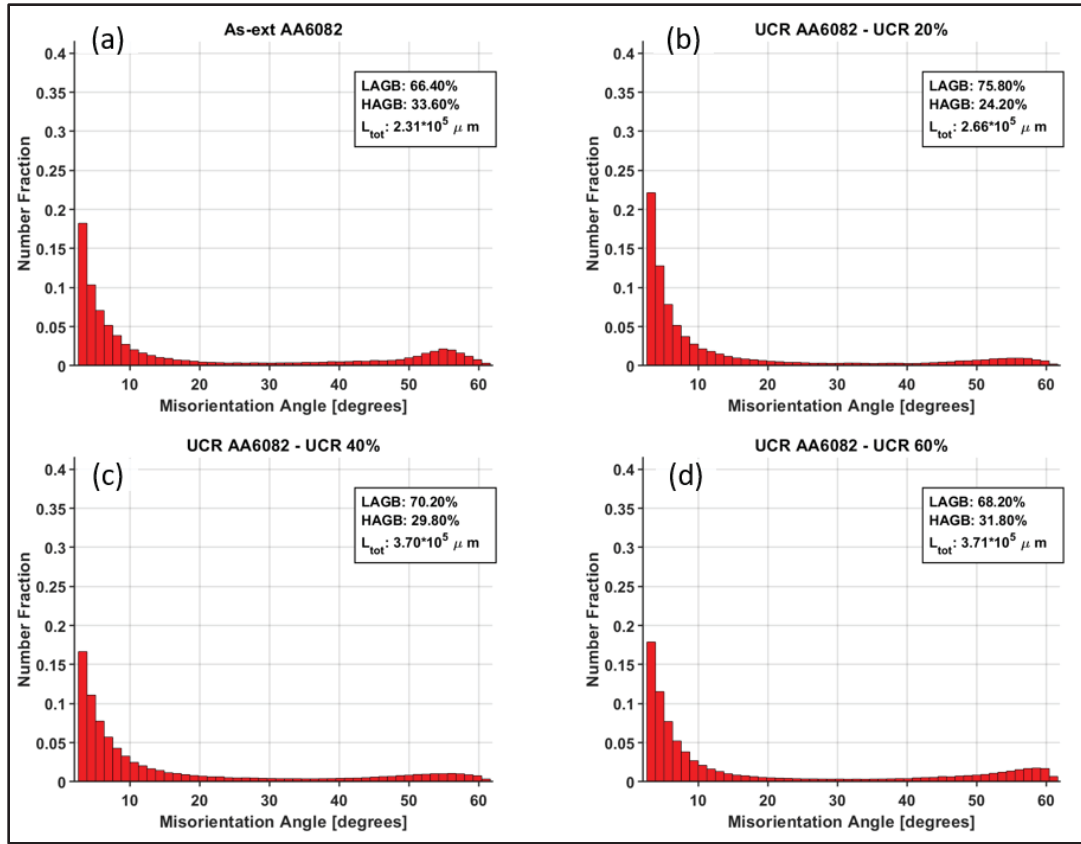


Figure 4-18 Misorientation angle distributions for as-extruded AA6082 and selected conditions under unidirectional cold rolling (UCR): (a) as-extruded condition, (b) UCR 20%, (c) UCR 40%, and (d) UCR 60%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

The $\{111\}$ pole figures of as-extruded AA6082 after UCR (Figure 4-19) reveal the evolution of crystallographic texture across increasing thickness reductions. At 10% reduction, the pole figure shows a strong fibrous deformation texture aligned near the TD axis, which corresponds to the normal orientation of the fibrous grains in the as-extruded material. This high initial intensity (Max = 10.4) reflects the strong inherited texture from the extrusion process. As rolling progresses to 20% and 30%, the texture remains sharp and intense, reaching a maximum of 11.7, indicating continued dominance of the original extrusion-induced orientation. However, from 40% onward, the maximum intensity decreases noticeably dropping to 6.2 at 50% before slightly increasing to 7.9 at 60%. This reduction in intensity can be attributed to the fact that rolling was performed perpendicular to the initial extrusion direction, which likely

destabilized the well-aligned fibrous texture and promoted orientation redistribution. As a result, the pole figure becomes broader and less peaked, reflecting the gradual transition from a strong extrusion texture to a more balanced, rolling-induced one.

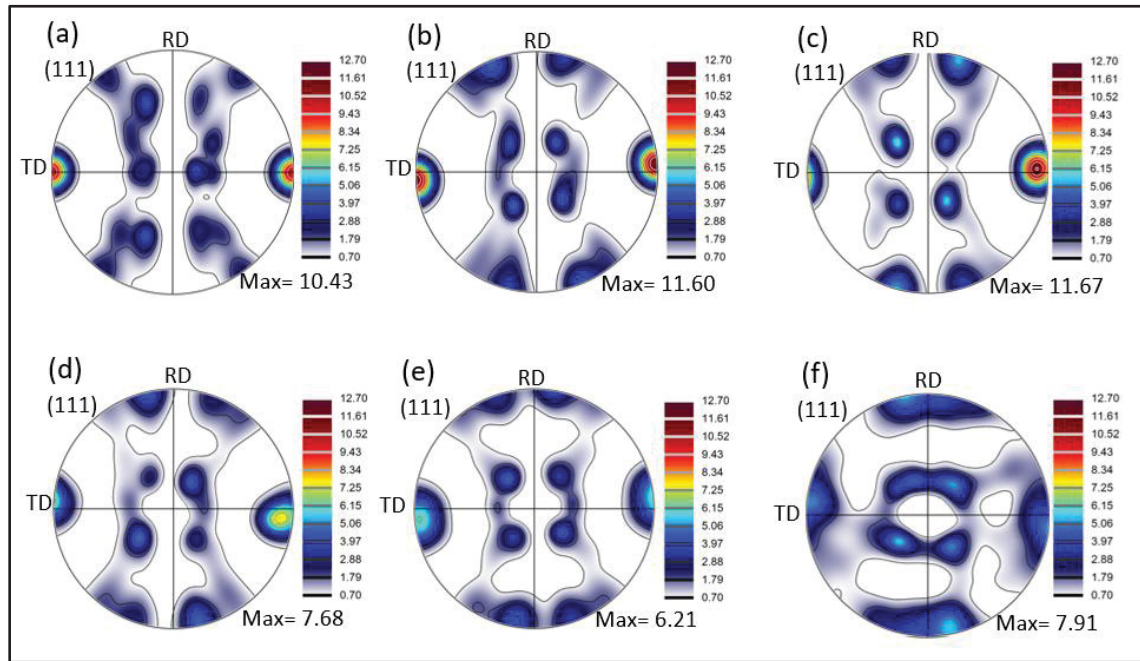


Figure 4-19 $\{111\}$ pole figures of as-extruded AA6082 after unidirectional cold rolling (UCR) at thickness reductions of (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%

When compared to the as-cast condition under UCR (Figure 4-6), the as-extruded material starts with a much stronger initial texture due to prior processing. However, while the as-cast UCR texture steadily sharpens with increasing deformation, the as-extruded UCR texture initially intensifies but then weakens after 40% reduction, suggesting that the perpendicular rolling direction disrupts the pre-existing orientation framework.

To further examine the microstructure at severe deformation, EBSD characterization was performed at 90% reduction, as shown in Figure 4-20. Because the microstructure becomes highly refined at this strain level, the EBSD maps were acquired at a higher magnification (500 $\times$) compared with the lower reductions. In the TD–ND cross-sectional view, the EBSD orientation map (Figure 4-20a) shows a highly refined banded structure with pronounced grain subdivision and compression along the normal direction. The corresponding band contrast map

(Figure 4-20b) reveals a dense boundary network and pronounced substructure formation, with a high density of LAGBs. The associated $\{111\}$ pole figure (Figure 4-20c) shows a maximum intensity of approximately 5.6. Compared with the 10–60% UCR conditions shown in Figure 4-19, this indicates that the strong inherited texture of the as-extruded state is reduced at the highest strain level, while a pronounced deformation-induced texture remains at 90% reduction.

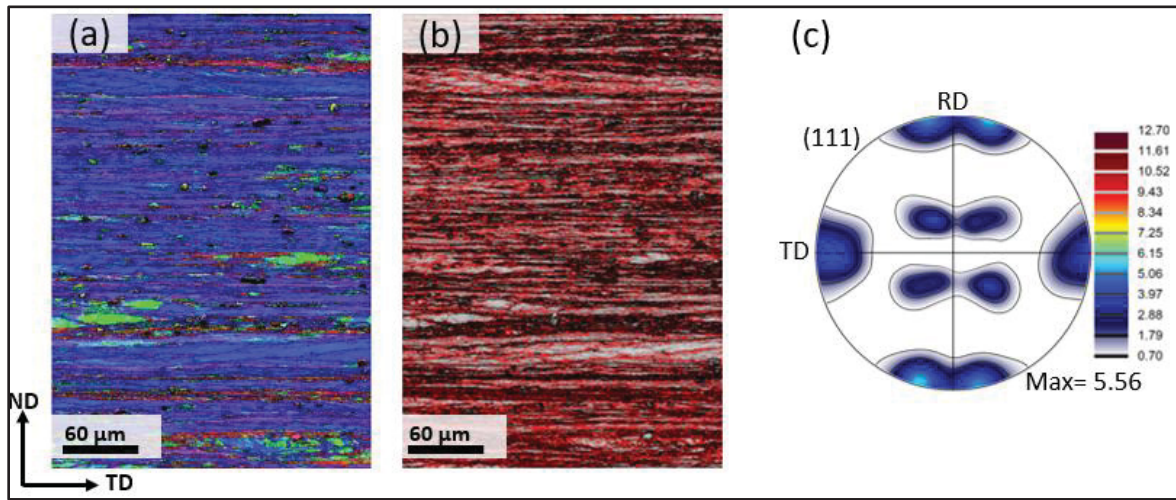


Figure 4-20 High-magnification EBSD characterization of as-extruded AA6082 after 90% unidirectional cold rolling (UCR): (a) EBSD orientation map, (b) band contrast map, and (c) $\{111\}$ pole figure

The texture evolution of as-extruded AA6082 during UCR can be also described using the inverse pole figures shown in Figure 4-21 together with the volume fractions of the main FCC texture components presented in Figure 4-22. As mentioned earlier for the initial condition (Figure 4-2), the as-extruded material exhibits a duplex fiber texture in which crystallographic directions close to $\langle 111 \rangle$ and $\langle 100 \rangle$ are preferentially aligned with the extrusion direction. In the present work, rolling was performed perpendicular to the extrusion direction and parallel to the prior transverse direction, so the imposed deformation was applied across the initial extrusion-induced texture.

With increasing reduction, the inverse pole figures show a progressive breakdown of the inherited crystallographic alignment. The most pronounced change occurs in the TD inverse

pole figure, where the maximum intensity decreases from 12.7 in the initial state to 9.9 at 20% reduction and further to 5.6 at 40% reduction, indicating a rapid weakening of the extrusion fiber. Over the same reduction range, the volume fraction data reveal a transition away from the extrusion-related components, with the RCu fraction decreasing continuously, while the Bs component increases sharply to about 47% at 20% reduction and remains high at intermediate reductions. At higher reductions, the inverse pole figures indicate the stabilization of a deformation-induced rolling texture. In the RD inverse pole figure, the intensity progressively concentrates near the $\langle 111 \rangle$ pole, reaching a maximum of 3.7 at 90% reduction. At the same time, the TD maximum continues to decrease, reaching 3.9 at the highest reduction level. These changes are accompanied by a clear evolution in the texture components: the S component increases from below 1% at 10% reduction to approximately 10% at 40–50%, about 11% at 60%, and rises sharply to about 38% at 90% reduction, while the Bs component remains significant and reaches about 27% at the highest reduction. The Cu component remains negligible throughout the deformation range. So, the strong extrusion-induced texture is progressively replaced by a deformation-induced FCC rolling texture, with the highest reduction level characterized by the coexistence of dominant S and Bs components.

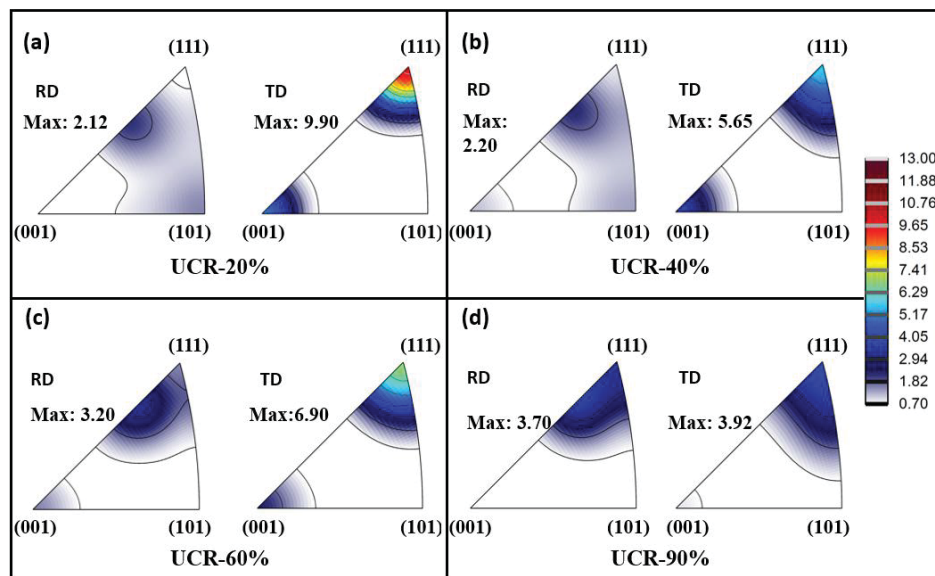


Figure 4-21 Inverse pole figures along RD and TD showing the texture evolution of the as-extruded AA6082 during unidirectional cold rolling at (a) 20%, (b) 40%, (c) 60%, and (d) 90% thickness reduction

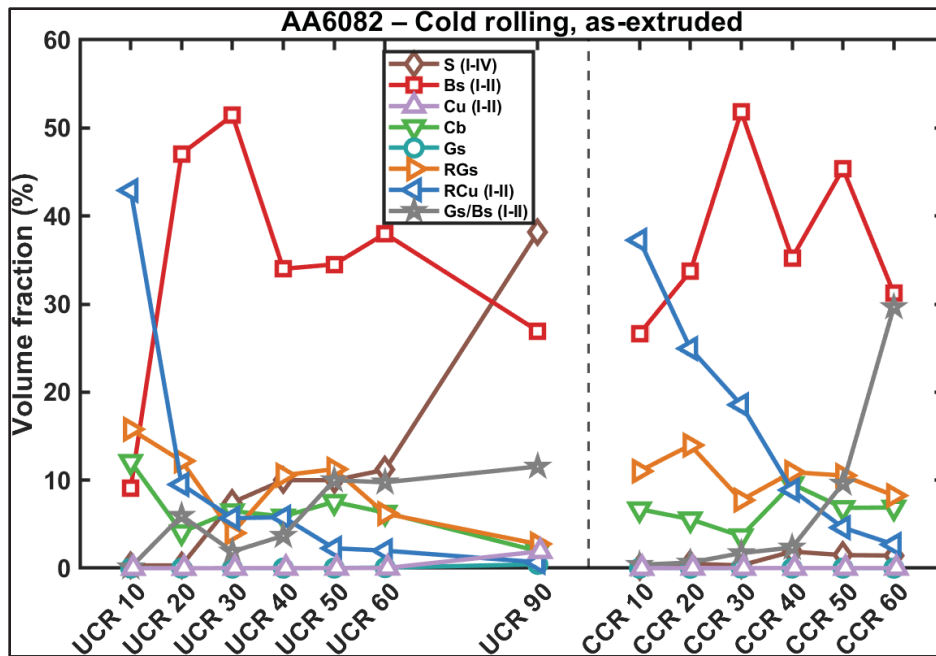


Figure 4-22 Volume fraction of selected texture components in as-extruded AA6082 as a function of thickness reduction under unidirectional and cross cold rolling

The evolution of the grain size distribution for as-extruded AA6082 during UCR shows a progressive refinement of the microstructure with increasing deformation. The cumulative area-fraction curves (Figure 4-23) shift steadily toward smaller grain sizes, and the distributions become steeper as the reduction increases, indicating a transition from the relatively coarse as-extruded structure to a finer and more uniform grain structure. Quantitatively, the maximum grain size of about 379 μm in the as-extruded condition decreases to about 220 μm at 40% reduction and is further reduced to about 72 μm at 90% reduction. A similar trend is observed for the 85% cumulative grain size, which decreases from about 243 μm in the initial state to about 192 μm at 40% reduction, 177 μm at 60% reduction, and finally to about 43 μm at 90% reduction, confirming the strong refinement at the highest strain level.

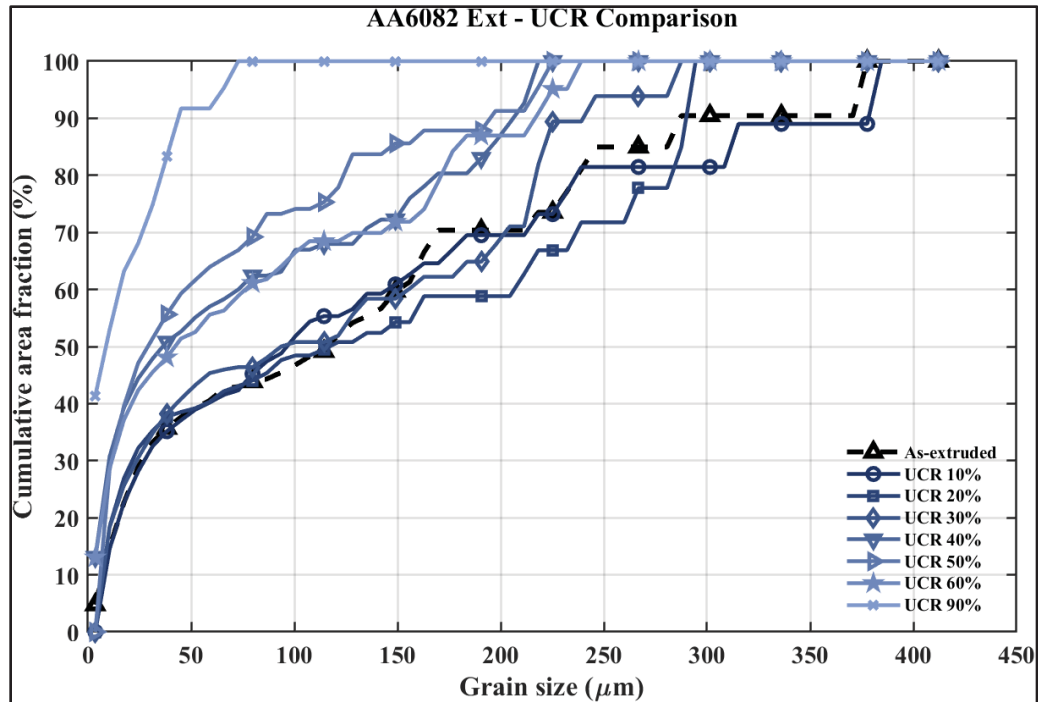


Figure 4-23 Cumulative grain size distribution curves for as-extruded AA6082 under unidirectional cold rolling (UCR) at various thickness reductions

Cross Cold Rolling (CCR)

The microstructural evolution of as-extruded AA6082 under CCR is illustrated in the band contrast and EBSD orientation maps (Figure 4-24 and Figure 4-25). Unlike UCR, the CCR process involves alternating rolling passes rotated 90° relative to each other. The cross-sectional orientation shown here is taken parallel to the initial extrusion direction. Since the rolling direction changes by 90° after each pass, every second deformation step was applied along the direction observed in this cross section. Unlike performance of this deformation, cross rolling does not severely disrupt or destabilize the microstructure; instead, it tends to preserve elongated features aligned with the extrusion direction. However, the overall compression effect along the normal direction appears lower compared to UCR (Figure 4-17). Concurrently, the band contrast maps reveal increasingly refined substructures and higher densities of red boundaries, signifying the accumulation of low-angle misorientations.

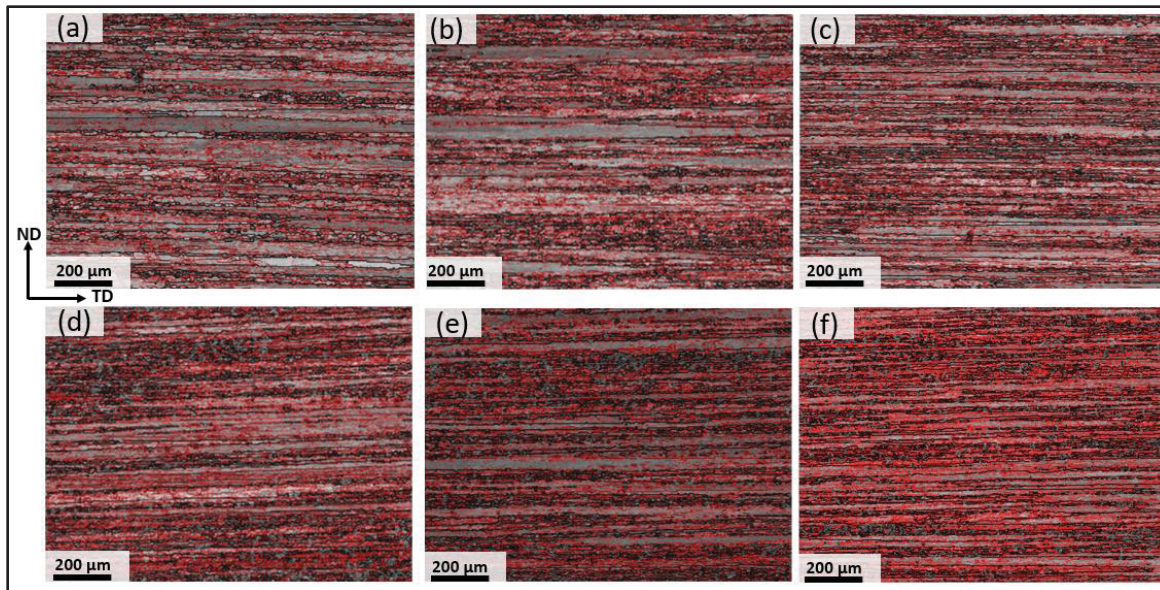


Figure 4-24 Band contrast maps of as-extruded AA6082 subjected to cross cold rolling (CCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction e.g. TD–ND plane

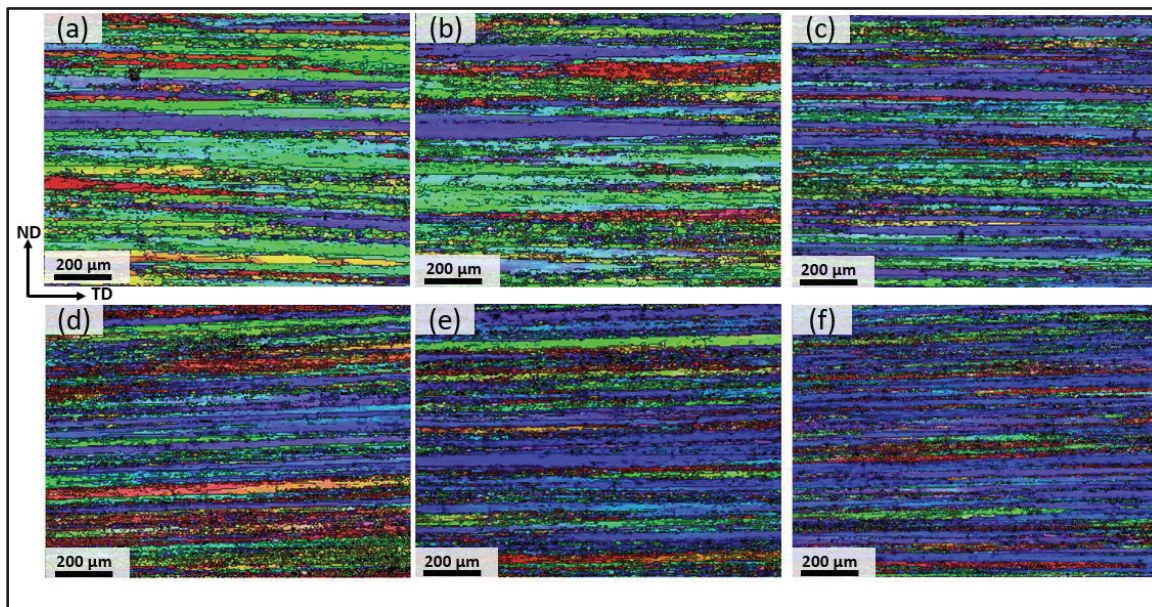


Figure 4-25 EBSD orientation maps of as-extruded AA6082 subjected to Cross cold rolling (CCR) at various thickness reductions: (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%. All sections are taken perpendicular to the rolling direction e.g. TD–ND plane

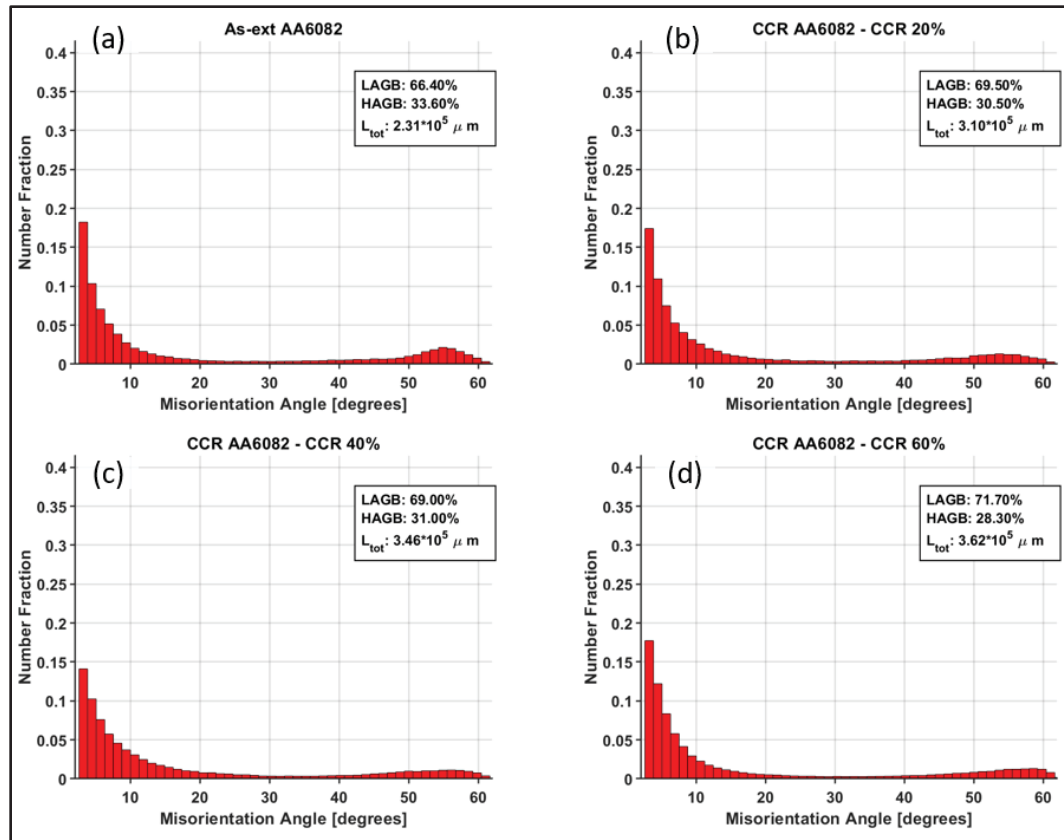


Figure 4-26 Misorientation angle distributions for as-extruded AA6082 and selected conditions under cross cold rolling (CCR): (a) as-extruded condition, (b) CCR 20%, (c) CCR 40%, and (d) CCR 60%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

The evolution of the grain-boundary misorientation distributions for as-extruded AA6082 during CCR is shown in Figure 4-26. Similar to the UCR condition, the misorientation distributions remain concentrated in the low-angle region throughout deformation, indicating the progressive formation of deformation-induced substructures. With increasing thickness reduction, the distributions gradually broaden and the relative fraction of HAGBs changes moderately. At 20% reduction, the LAGB fraction is 69% and the HAGB fraction is 30%, showing the formation of additional LAGBs compared with the initial condition. At 40% reduction, the boundary fractions remain nearly balanced, with 69% LAGBs and 31% HAGBs, indicating that cross rolling promotes continuous subdivision of the grains while maintaining a relatively stable proportion of high-angle boundaries. At 60% reduction, the LAGB fraction

slightly increases to 72%, while the HAGB fraction decreases to 28%, suggesting that further deformation mainly produces additional sub-boundaries rather than forming new HAGBs. In parallel, the total boundary length increases significantly with deformation, rising from $3.10 \times 10^5 \mu\text{m}$ at 20% reduction to $3.62 \times 10^5 \mu\text{m}$ at 60% reduction. This increase reflects the progressive subdivision of the microstructure caused by the multidirectional strain path during cross rolling, which distributes the deformation over successive rolling directions and promotes the formation of a refined boundary network.

The $\{111\}$ pole figures (Figure 4-27) reveal a texture evolution defined by the competitive balance between the inherited extrusion fiber and the imposed rolling strain. At 10% reduction, the texture displays a high maximum intensity (Max = 13.1), inherited from the extrusion process. As deformation progresses to 40%, a significant transient destabilization occurs, where the maximum intensity drops to 6.3. This redistribution suggests that the 90° rotation between passes effectively disrupts the stable extrusion fiber at intermediate strains.

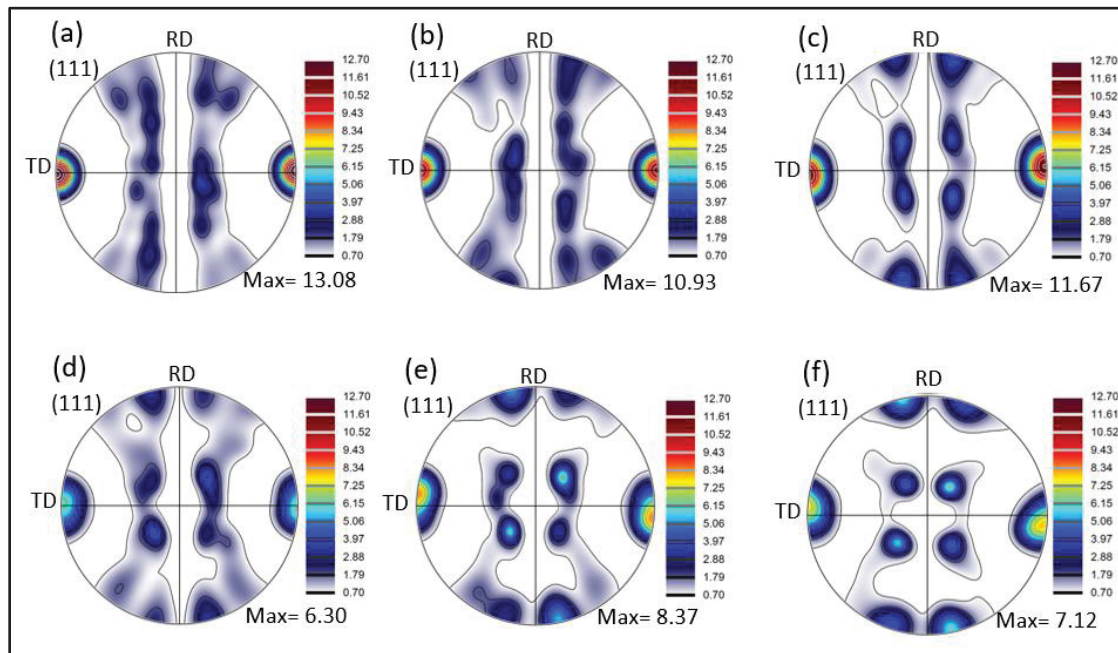


Figure 4-27 $\{111\}$ pole figures of as-extruded AA6082 after cross cold rolling (CCR) at thickness reductions of (a) 10%, (b) 20%, (c) 30%, (d) 40%, (e) 50%, and (f) 60%

Between 50% and 60% reduction, the texture re-intensifies (Max = 7.1 at 60%), with the peaks becoming highly localized along the TD axis. Unlike the UCR condition, which shows a broader spreading of intensity, the CCR pole figures remain focused, suggesting a more specialized orientation convergence.

The rolling texture evolution of the as-extruded AA6082 under cross cold rolling can be interpreted using the inverse pole figures shown in Figure 4-28 together with the volume fractions of the main FCC texture components presented in Figure 4-22. At low reductions (10–20%), the inverse pole figures still show features related to the inherited extrusion texture, particularly the strong intensity near the $\langle 111 \rangle$ corner in TD, while the RD distribution remains relatively diffuse.

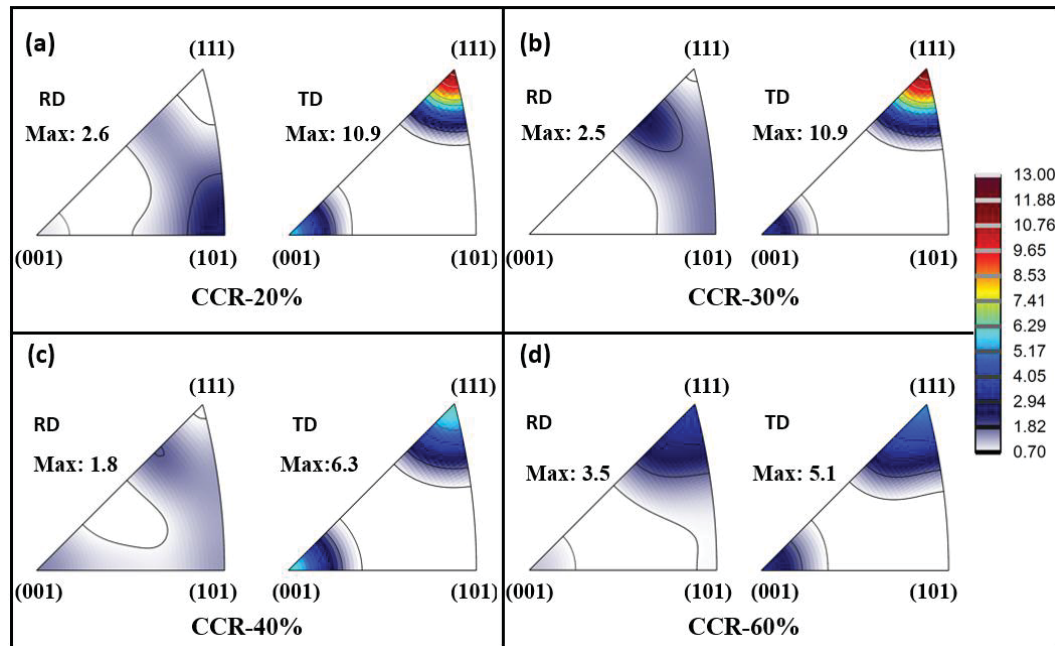


Figure 4-28 Inverse pole figures along RD and TD showing the texture evolution of the as-extruded AA6082 during cross cold rolling at (a) 20%, (b) 30%, (c) 40%, and (d) 60% thickness reduction

The RD inverse pole figure in Figure 4-28 at 20% reduction shows a maximum located close to the $\langle 101 \rangle$ region, which is consistent with the relatively high fraction of the RCu component inherited from the extrusion state. At 30% reduction, the intensity shifts toward the $\langle 112 \rangle$ –

$\langle 111 \rangle$ region, indicating the progressive weakening of the extrusion-related texture and the development of rolling-texture orientations, particularly the Brass component toward the $\langle 112 \rangle$ region. This evolution is consistent with the texture-component fractions: RCu remains high at low reductions ($\sim 37\%$ at 10% and $\sim 25\%$ at 20%) but decreases continuously with increasing reduction, reaching below $\sim 3\%$ at 60%. In contrast, the Bs fraction increases from approximately 27% at 10% reduction to about 52% at 30% and remains above $\sim 30\%$ up to 60% reduction, becoming the dominant deformation component. RG also contributes to the evolving texture, whereas the S component remains below $\sim 2\%$ and Cu remains negligible throughout the investigated reduction range. The progressive redistribution of intensity in the inverse pole figures therefore reflects the gradual replacement of the extrusion-induced RCu-dominated texture by deformation-induced rolling components, mainly Brass.

A comparison of the texture evolution in as-cast (Figure 4-8) and as-extruded (Figure 4-22) AA6082 under UCR and CCR highlights the strong influence of both the initial microstructure and the strain path on the development of rolling texture. In the as-extruded condition, the initial texture contains strong extrusion-related components, particularly RCu and RGs, whereas the as-cast condition begins from a weaker and more dispersed orientation state. Under UCR, both materials evolve toward deformation-dominated FCC rolling textures, but the as-extruded material shows a more rapid transition from extrusion-related components to strong Bs and S components. In contrast, the as-cast material exhibits a more gradual increase in the rolling components with increasing reduction. Under CCR, the texture evolution remains more distributed than in UCR. In the as-cast state, CCR promotes a moderate development of the rolling components without a strong dominance of a single orientation until high reduction, whereas in the as-extruded state the Bs component becomes dominant while the S component remains limited.

A notable difference between UCR and CCR appears at high reductions. In the as-extruded condition, UCR leads to a strong increase of the S component at large strain, reaching high values at 60% and becoming dominant at 90% reduction, while the Bs component remains significant but lower. In contrast, under CCR the Bs component continues to increase with reduction and becomes the dominant component at high strain, while the S fraction remains

very small even at 60% reduction. A similar tendency is observed for the as-cast condition, where UCR promotes a stronger growth of the S component at high reductions, whereas CCR favors the development of Bs-dominated textures. This difference indicates that the strain path strongly influences the balance between S and Bs components, making the evolution of these two orientations a critical aspect of the texture development in both initial conditions.

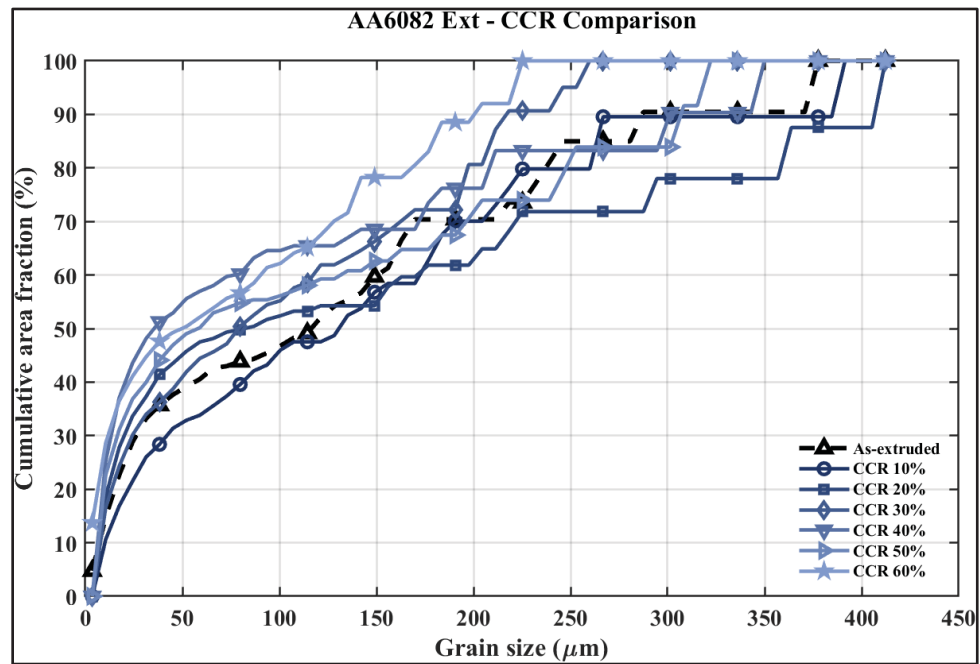


Figure 4-29 Cumulative grain size distribution curves for as-extruded AA6082 under cross cold rolling (CCR) at various thickness reductions

The evolution of the grain size distribution for as-extruded AA6082 during CCR also shows progressive refinement with increasing deformation, as illustrated by the cumulative area-fraction curves in Figure 4-29. The curves shift gradually toward smaller grain sizes and become steeper with increasing reduction, indicating the development of a finer microstructure. Quantitatively, the maximum grain size of about 379 μm in the as-extruded condition remains relatively high at low reductions, before decreasing to about 347 μm at 40% reduction and about 223 μm at 60% reduction. A similar trend is observed for the 85% cumulative grain size, which shows a slight initial increase to about 345 μm at 20% reduction, followed by a decrease to about 232 μm at 40% and about 181 μm at 60% reduction. This

temporary increase at 20% reduction is more likely associated with local microstructural heterogeneity in the as-extruded state than with actual grain growth. These values indicate that grain refinement during CCR occurs more gradually compared with UCR.

4.3.1.3 Cold Rolled As-cast AA6063

To explore the impact of chemical composition and compare the results with AA6082, rolling was conducted on AA6063 at thickness reductions of 40%, 60% and 80%. This allowed for a deeper understanding of how the different compositions influence the texture and microstructure.

Unidirectional Cold Rolling (UCR)

Progressive cold rolling of as-cast AA6063 under UCR leads to significant microstructural changes, as shown by the EBSD orientation maps and band contrast images in Figure 4-30 together with the misorientation distributions in Figure 4-31.

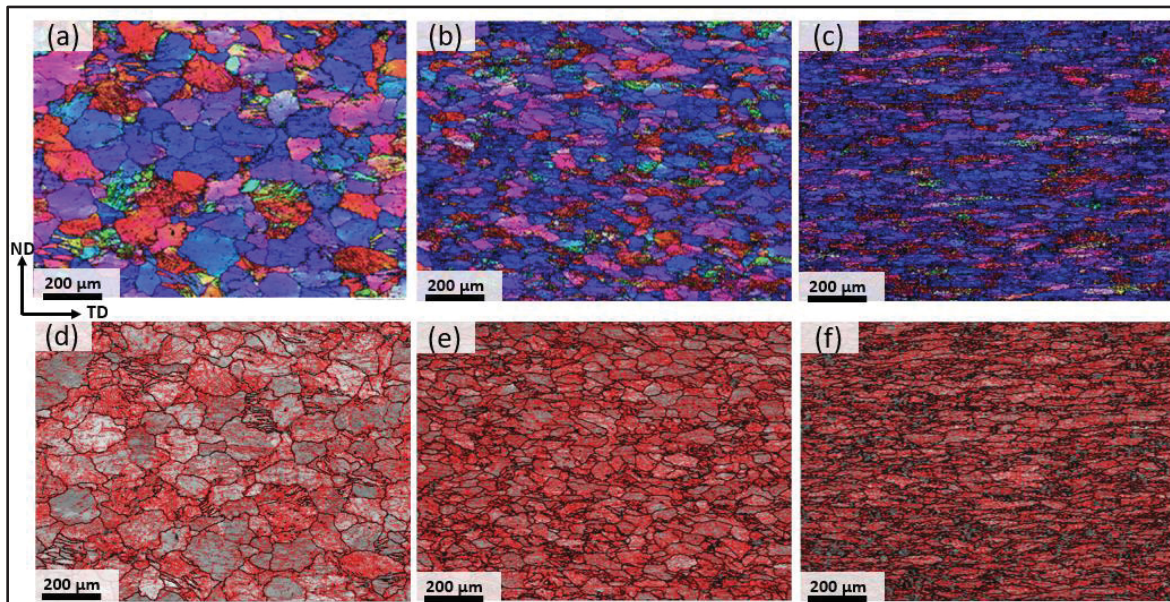


Figure 4-30 EBSD orientation maps (a–c) and corresponding band contrast images (d–f) of as-cast AA6063 after unidirectional cold rolling (UCR) to (a,d) 40%, (b,e) 60%, and (c,f) 80% thickness reduction, observed on the TD–ND plane

At 40% reduction, the EBSD map still shows partially equiaxed grains, although the orientation contrast becomes less uniform compared to the initial state (Figure 4-1). The band contrast image indicates an increase in boundary density, with both LAGBs and HAGBs visible, the former appearing in red and the latter in black. This evolution is consistent with the misorientation histogram, where the fraction of LAGBs increases significantly, reaching about 87%, while the total boundary length rises from $4.42 \times 10^4 \mu\text{m}$ in the as-cast condition to $2.31 \times 10^5 \mu\text{m}$ at 40% reduction (Figure 4-31).

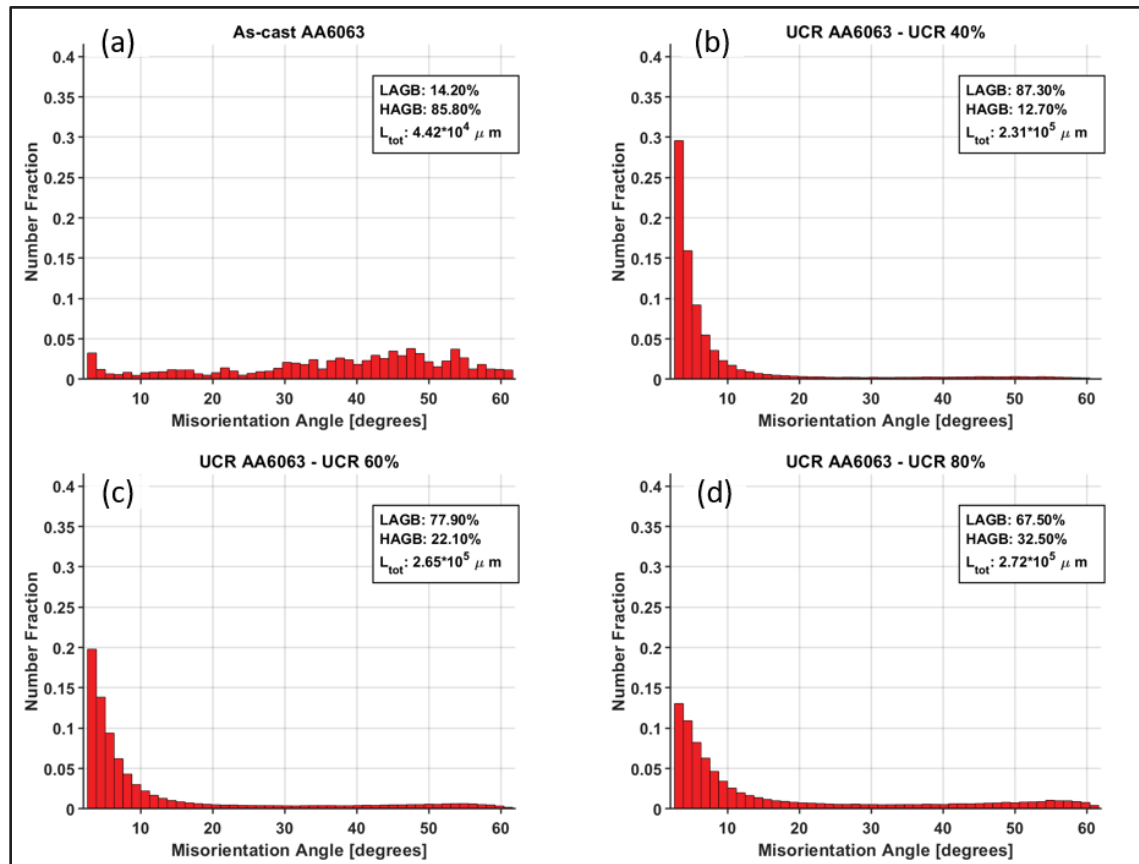


Figure 4-31 Misorientation angle distributions for as-cast AA6063 and selected conditions under unidirectional cold rolling (UCR): (a) as-cast condition, (b) UCR 40%, (c) UCR 60%, and (d) UCR 80%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

With further deformation to 60%, the EBSD map shows a more pronounced banded morphology together with a higher density of internal boundaries. The misorientation distribution indicates a slight increase in HAGBs compared to 40%, and the total boundary length increases further to about $2.65 \times 10^5 \mu\text{m}$, confirming the progressive subdivision of the grains during deformation.

At 80% reduction, the microstructure becomes strongly flattened along the normal direction, and the EBSD map shows heavily banded features typical of severe rolling deformation in the TD–ND section. The band contrast image reveals a very high boundary density with numerous subgrain boundaries. The misorientation histogram shows a marked increase in the HAGB fraction, reaching about 32%, while the total boundary length increases to approximately $2.72 \times 10^5 \mu\text{m}$. This behavior indicates that at high strain the microstructure is dominated by deformation-induced substructures together with the formation of new high-angle boundaries, leading to a highly refined and strongly banded grain morphology.

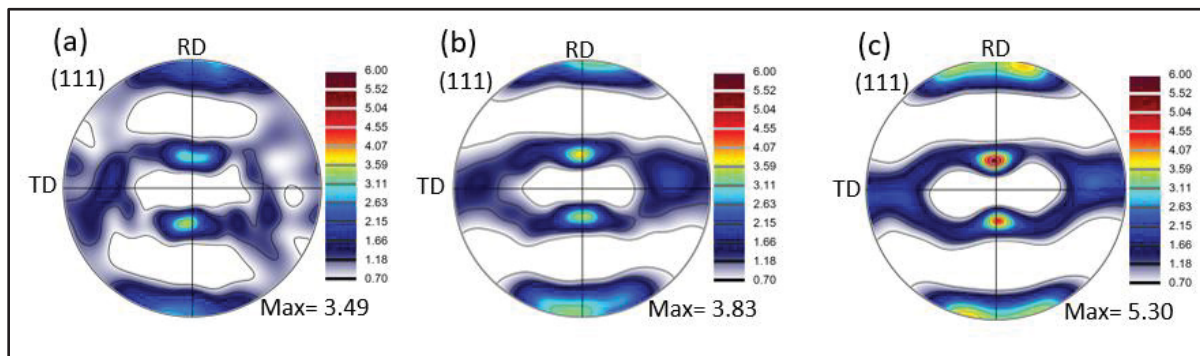


Figure 4-32 $\{111\}$ pole figures showing the evolution of crystallographic texture in as-cast AA6063 during unidirectional cold rolling (UCR) at (a) 40%, (b) 60%, and (c) 80% thickness reduction

The evolution of deformation texture in as-cast AA6063 during UCR is illustrated by the $\{111\}$ pole figures shown in Figure 4-32. At 40% reduction, the pole figure still exhibits a relatively broad and moderately dispersed texture distribution, with a maximum intensity of 3.5. As the reduction increases to 60%, the texture becomes more concentrated along the rolling direction, and the maximum intensity increases slightly to 3.8. At 80% reduction, the pole figure shows

a much sharper and more localized intensity distribution, with strong β -fiber aligned parallel to the rolling direction and a maximum intensity reaching 5.3. These results indicate progressive texture strengthening with increasing deformation, together with the gradual development of a more pronounced rolling texture in AA6063 under UCR.

To further examine the microstructure at severe deformation, EBSD characterization was performed at 90% reduction, as shown in Figure 4-33. Because the microstructure becomes highly refined at this strain level, the EBSD maps were acquired at higher magnification compared with the lower reductions. The EBSD orientation map (Figure 4-33a) shows a strongly banded structure with pronounced grain subdivision and marked flattening along the normal direction in the TD–ND section. The corresponding band contrast map (Figure 4-33b) reveals a very dense boundary network and well-developed substructure, characterized by a high density of LAGBs together with numerous fragmented grains. The associated $\{111\}$ pole figure (Figure 4-33c) indicates a further strengthening of the deformation texture, with a maximum intensity of about 5.5, confirming the continued development of the rolling texture with strong $\{111\}$ fiber components at the highest strain level.

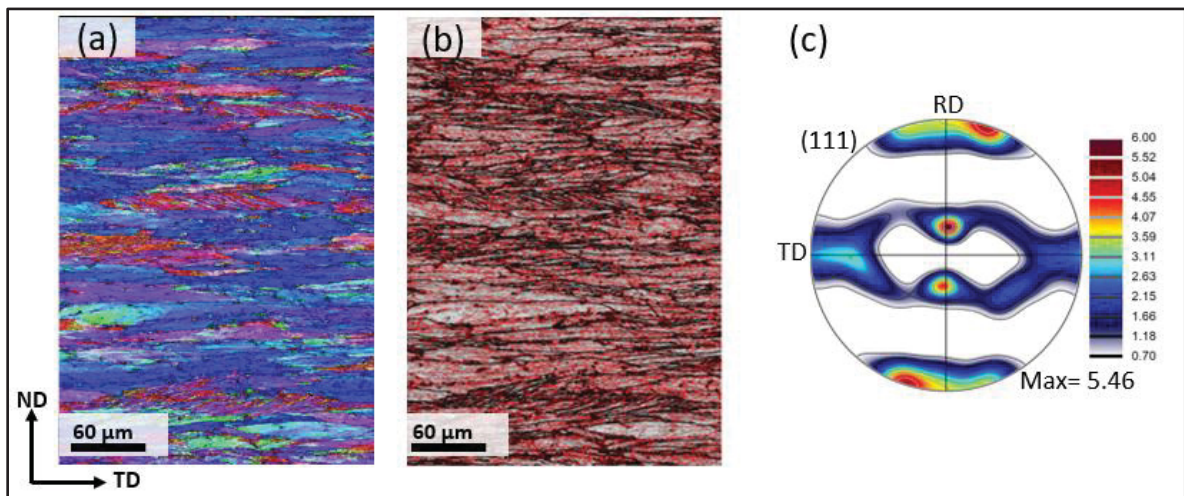


Figure 4-33 High-magnification EBSD characterization of as-cast AA6063 after 90% unidirectional cold rolling (UCR): (a) EBSD orientation map, (b) band contrast map, and (c) $\{111\}$ pole figure

The rolling-texture evolution in as-cast AA6063 under unidirectional cold rolling is further described through the volume fractions of the main FCC texture components (Figure 4-34). With increasing reduction, the texture becomes progressively dominated by the classical deformation components, particularly S, Cu, and Bs. The S component increases continuously from about 21% at 40% reduction to about 28% at 60%, about 37% at 80%, and reaches approximately 42% at 90% reduction, becoming the dominant texture component at the highest strain level. The Cu component also increases progressively, from about 10% at 40% reduction to about 14% at 60%, about 16% at 80%, and about 19% at 90% reduction. In parallel, the Bs component rises from about 10% at 40% reduction to about 13% at 60%, about 17% at 80%, and about 20% at 90% reduction. By contrast, the non-dominant components remain relatively low throughout the deformation range. The Cb component decreases from about 2% at 40–60% reduction to below 1% at 90%, while RGs remains very limited and RCu stays negligible at all reductions. Thereby, these trends indicate a clear shift toward a deformation-dominated FCC rolling texture with increasing strain, with the highest reduction level characterized by the strong predominance of the S component together with notable contributions from Cu and Bs.

The evolution of the grain size distribution for as-cast AA6063 during UCR shows a progressive refinement of the microstructure with increasing deformation. The cumulative area-fraction curves (Figure 4-35) shift gradually toward smaller grain sizes and become steeper as the rolling reduction increases, indicating a transition from the relatively coarse as-cast structure to a finer and more uniform grain distribution. Quantitatively, the maximum grain size is about 153 μm in the as-cast condition and shows an apparent increase to about 173 μm at 40% reduction, followed by a decrease to about 125 μm at 60%, about 103 μm at 80%, and about 61 μm at 90% reduction. A similar trend is observed for the 85% cumulative grain size, which increases from about 127 μm in the initial state to about 146 μm at 40% reduction, before decreasing markedly to about 70 μm at 60%, about 55 μm at 80%, and about 41 μm at 90% reduction. This initial increase at 40% reduction does not indicate physical grain growth, but is more likely associated with sampling heterogeneity and the limitations of EBSD boundary detection at low strains, where distorted original grain boundaries may not yet be fully resolved as high-angle grain boundaries. These results indicate that grain refinement

becomes clearly pronounced beyond 40% reduction, with the strongest refinement occurring at the highest strain level.

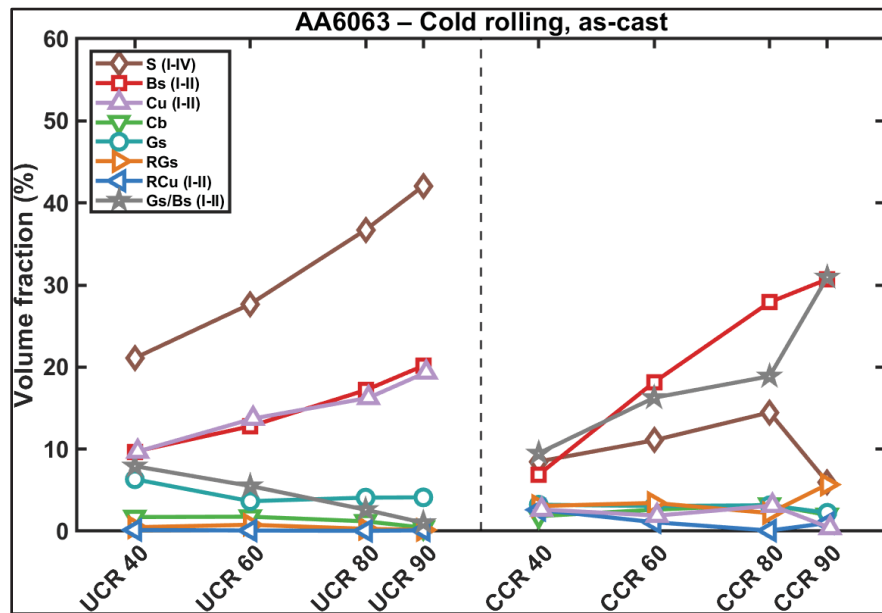


Figure 4-34 Volume fraction of selected texture components in as-cast AA6082 as a function of thickness reduction under unidirectional and cross cold rolling

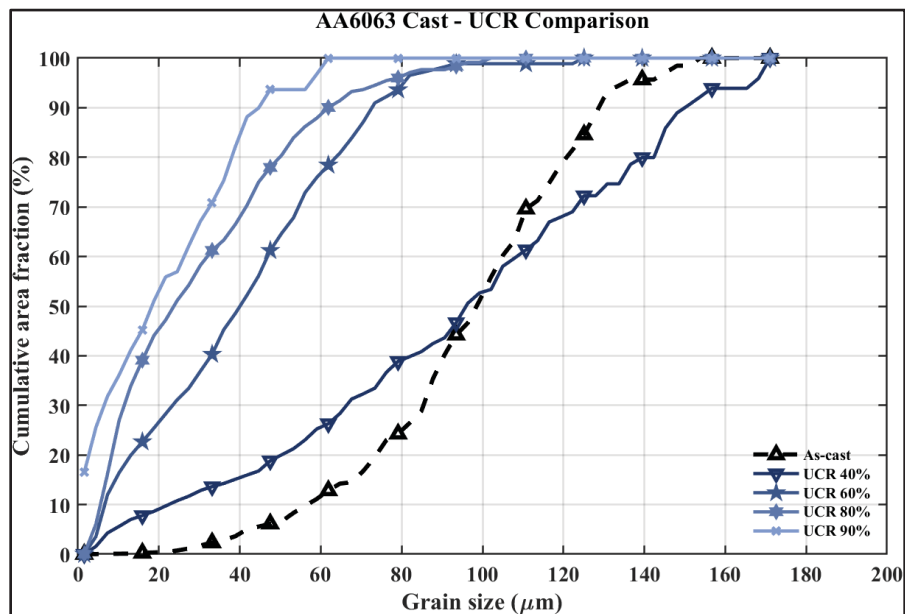


Figure 4-35 Cumulative grain size distribution curves for as-cast AA6063 under unidirectional cold rolling (UCR) at various thickness reductions

Cross Cold Rolling (CCR)

Progressive cold rolling of as-cast AA6063 under CCR also leads to pronounced microstructural evolution, as shown by the EBSD orientation maps and band contrast images in Figure 4-36 together with the misorientation distributions in Figure 4-37. At 40% reduction, the EBSD map still shows a relatively equiaxed grain morphology with only limited elongation in the TD–ND section. The band contrast image reveals a clear increase in boundary density, with both LAGBs and HAGBs visible, the former shown in red and the latter in black. The misorientation histogram confirms that the microstructure is dominated by LAGBs, reaching about 86.5%, while the total boundary length increases strongly from $4.42 \times 10^4 \mu\text{m}$ in the as-cast state to about $2.33 \times 10^5 \mu\text{m}$ at 40% reduction.

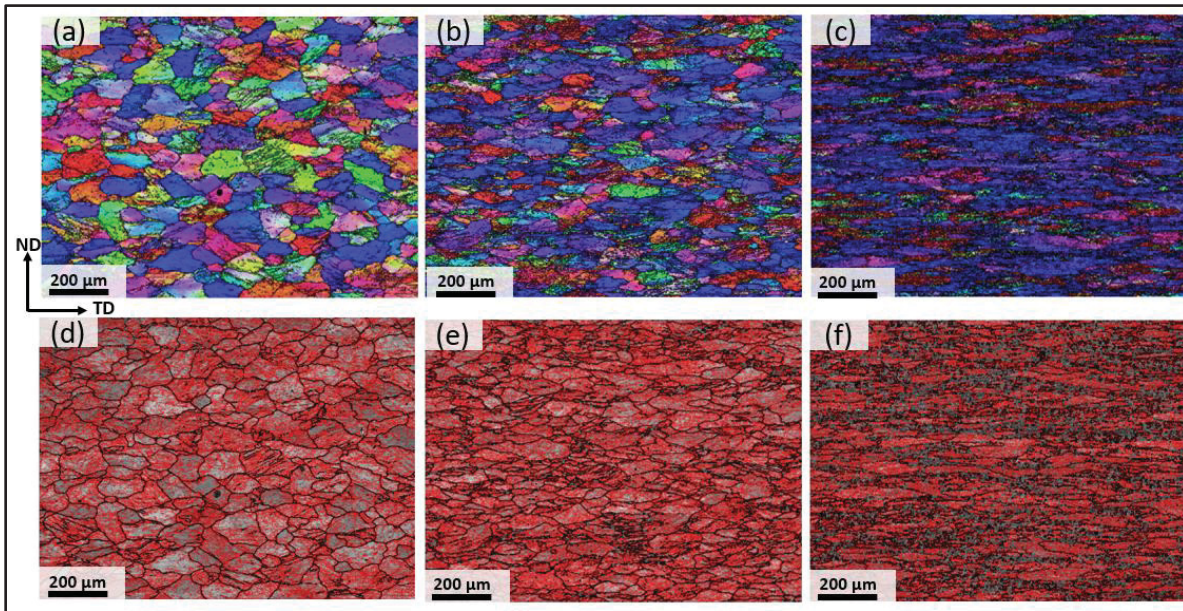


Figure 4-36 EBSD orientation maps (a–c) and corresponding band contrast images (d–f) of as-cast AA6063 after cross cold rolling (CCR) to (a,d) 40%, (b,e) 60%, and (c,f) 80% thickness reduction, observed on the TD–ND plane

With increasing deformation to 60%, the EBSD map shows a more pronounced banded morphology together with a higher density of internal boundaries. Because the rolling direction changes between passes, the effect of grain elongation becomes visible in the TD–ND section, in contrast to the UCR condition where such elongation is not directly observed in this plane.

The misorientation distribution indicates a slight increase in the HAGB fraction to about 21.3%, while the total boundary length rises further to approximately $2.55 \times 10^5 \mu\text{m}$, confirming the progressive formation of deformation-induced substructures during CCR.

At 80% reduction, the microstructure becomes strongly refined and shows a well-developed banded structure in the TD–ND cross-section. The microstructure also exhibits flattening along the normal direction, although the compression effect appears less pronounced than in the UCR condition. The band contrast image reveals a very dense boundary network with numerous subgrain boundaries. The misorientation histogram shows a further increase in the HAGB fraction to about 29.1%, while the total boundary length increases to about $2.73 \times 10^5 \mu\text{m}$. These results indicate that at high strain the alternating strain path in cross rolling promotes continuous grain subdivision and gradual conversion of LAGBs into HAGBs, leading to a highly refined microstructure while partially suppressing the strong grain elongation typically observed in UCR.

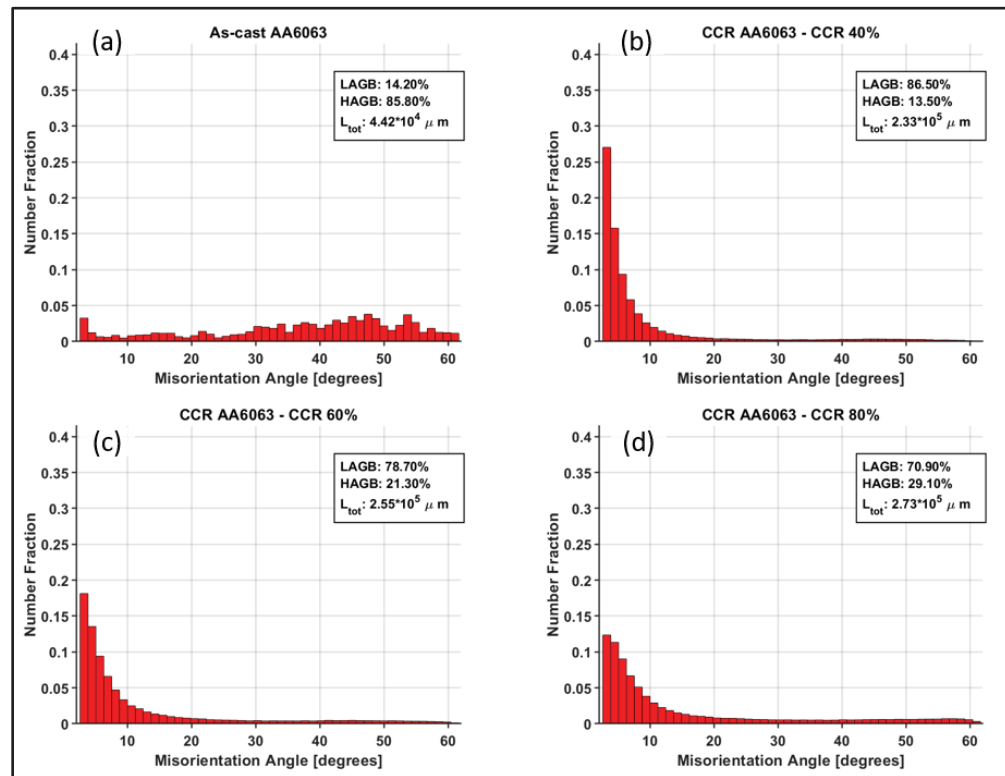


Figure 4-37 Misorientation angle distributions for as-cast AA6063 and selected conditions under cross cold rolling (CCR): (a) as-cast condition, (b) CCR 40%, (c) CCR 60%, and (d)

CCR 80%. The histograms show the number fraction of grain-boundary misorientation angles. The corresponding LAGB/HAGB fractions and total boundary length (L_{tot}) are indicated for each condition

The evolution of deformation texture in as-cast AA6063 during CCR is illustrated by the $\{111\}$ pole figures shown in Figure 4-38. At 40% reduction, the pole figure exhibits a relatively weak and dispersed texture distribution with a maximum intensity of about 2.3, indicating that the crystallographic orientations are still broadly distributed. With increasing reduction to 60%, the texture becomes more concentrated and symmetric, and the maximum intensity increases to about 3.2, reflecting the progressive development of deformation-induced orientations. At 80% reduction, the pole figure shows a noticeably sharper intensity distribution, with stronger $\{111\}$ fiber components aligned along the rolling directions and a maximum intensity reaching about 4.2. Compared with the UCR condition (Figure 4-32), the texture developed during CCR remains more distributed and less sharply concentrated, which is consistent with the alternating strain path in cross rolling that redistributes deformation and delays the formation of a strongly aligned rolling texture.

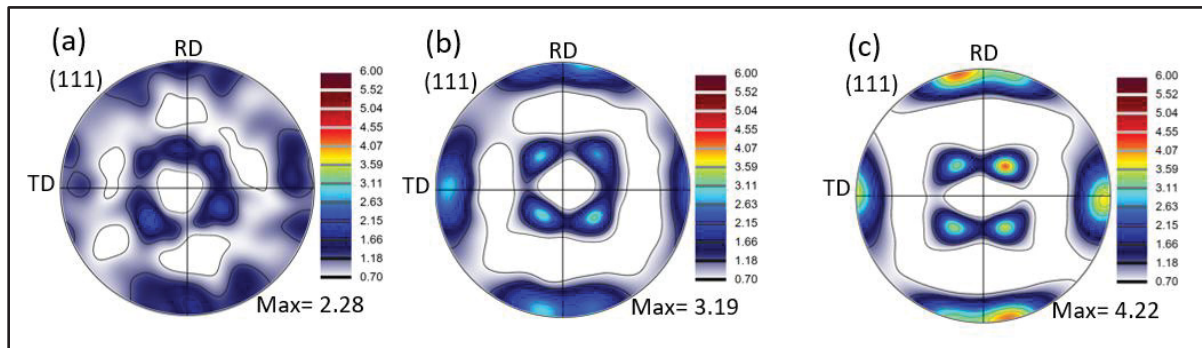


Figure 4-38 $\{111\}$ pole figures showing the evolution of crystallographic texture in as-cast AA6063 during cross cold rolling (CCR) at (a) 40%, (b) 60%, and (c) 80% thickness reduction

To further examine the microstructure at severe deformation, EBSD characterization was performed at 90% reduction, as shown in Figure 4-39. Because the microstructure becomes highly refined at this strain level, the EBSD maps were acquired at higher magnification compared with the lower reductions. The EBSD orientation map (Figure 4-39a) shows a

strongly banded and fibrous microstructure with pronounced grain subdivision and visible flattening along the normal direction in the TD–ND section. The corresponding band contrast map (Figure 4-39b) reveals a very dense boundary network and well-developed substructure, characterized by numerous LAGBs and fragmented domains. The associated $\{111\}$ pole figure (Figure 4-39c) indicates a further strengthening of the deformation texture, with a maximum intensity of about 4.7, confirming the continued development of rolling texture at the highest strain level under cross cold rolling.

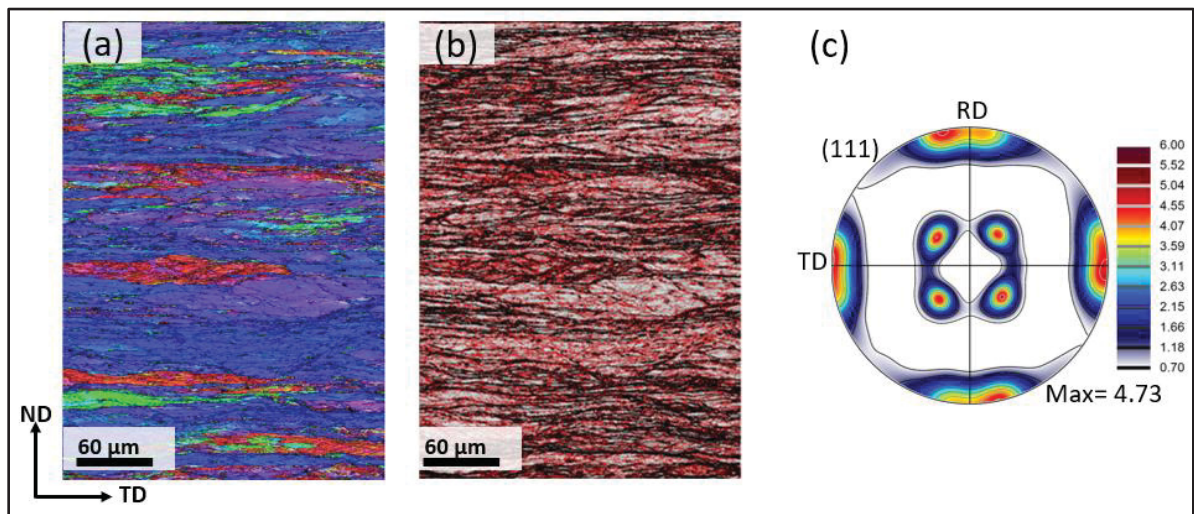


Figure 4-39 High-magnification EBSD characterization of as-cast AA6063 after 90% cross cold rolling (CCR): (a) EBSD orientation map, (b) band contrast map, and (c) $\{111\}$ pole figure

The evolution of rolling texture in as-cast AA6063 under CCR is further described through the volume fractions of the main FCC texture components (Figure 4-34). At 40% reduction, the texture is distributed among several deformation components, with relatively low fractions of S ($\sim 8\%$), Bs ($\sim 7\%$), and Cu ($\sim 3\%$), while the recrystallization-related components remain limited. With increasing reduction to 60% and 80%, the Bs component becomes progressively dominant, increasing from about 18% at 60% to about 28% at 80%, while the S component increases more moderately to about 11% and 14%, respectively. In contrast, the Cu component remains low and decreases slightly with deformation, and the RCu component stays negligible at all reductions. At the same time, the Gs/Bs component shows a clear increasing trend, rising from about 10 at 40% reduction to about 16 at 60% and about 19 at 80%

At 90% reduction, the texture is clearly dominated by the Bs component, which reaches about 31%, while the S component decreases to about 6%, indicating a redistribution of deformation texture under the alternating strain path of CCR. At the same time, the RGs component increases to about 6%, and the Gs/Bs component rises strongly to about 31%, reflecting the development of shear-related orientations at high strain. Overall, compared with the UCR condition where the S component becomes dominant at high reduction, the 90% CCR condition in AA6063 is characterized by the predominance of the Bs component together with a noticeable contribution from shear-related components, indicating that cross rolling promotes a more distributed deformation texture.

The evolution of the grain size distribution for as-cast AA6063 during CCR also indicates progressive refinement with increasing deformation. The cumulative area-fraction curves (Figure 4-40) shift gradually toward smaller grain sizes as the thickness reduction increases, reflecting the gradual breakdown of the initial cast structure. Quantitatively, the maximum grain size decreases from about 153 μm in the as-cast condition to about 133 μm at 40% reduction, about 129 μm at 60%, about 113 μm at 80%, and about 101 μm at 90% reduction. A similar trend is observed for the 85% cumulative grain size, which decreases from about 127 μm in the initial state to about 109 μm at 40%, about 81 μm at 60%, and about 62 μm at 80%. At 90% reduction, a slight increase to about 68 μm is observed. This apparent increase does not indicate actual grain coarsening, but is more likely related to statistical fluctuation and the reduced EBSD indexing quality at high deformation, where very fine subgrains may not be fully resolved and can therefore be merged into larger measured grains. These results indicate that grain refinement progresses steadily during CCR, although the refinement rate is slower than in UCR.

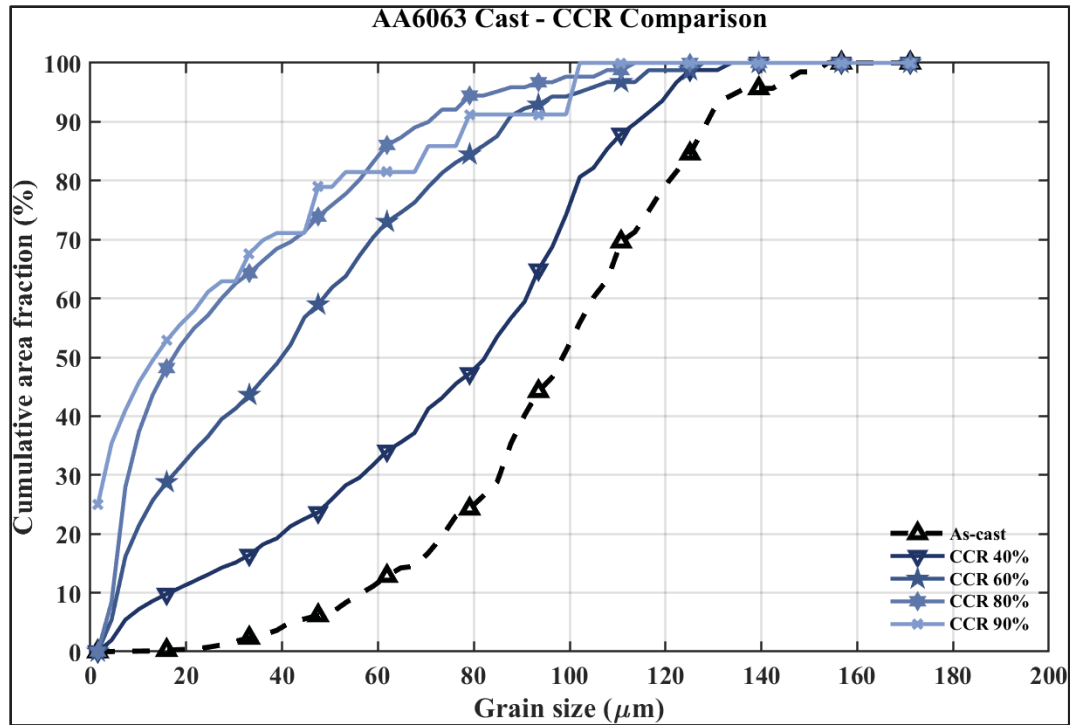


Figure 4-40 Cumulative grain size distribution curves for as-cast AA6063 under cross cold rolling (CCR) at various thickness reductions

4.3.2 Cold Rolling Discussion

4.3.2.1 Microstructure Evolution during Cold Rolling

The structural refinement of these 6xxx series alloys follows a hierarchical path of grain subdivision, driven primarily by the accumulation and reorganization of dislocations into stable substructures. In the early stages of cold rolling, typically up to 40% reduction, the transition from the initial equiaxed as-cast or fibrous extruded state to a deformation-dominated regime is marked by a sharp decline in EBSD indexing quality and band contrast. This degradation reflects the increase in lattice distortion and dislocation density within the grain interiors. Quantitatively, this is captured by a significant increase in absolute boundary length for both alloys and a massive jump in LAGB fraction for AA6063 (from 14% in as-cast to over 87% after rolling), while AA6082 maintains a high LAGB dominance (above 78%)

throughout the refinement process. This fragmentation stage is associated with a significant increase in total boundary length, as the original grains are subdivided into subgrains separated by dense networks of LAGBs to accommodate the imposed plastic strain.

The morphological evolution in the TD–ND section is characterized by progressive thinning of grains and deformation bands along the normal direction. In the as-cast alloys, the initially coarse, equiaxed grains are gradually compressed, evolving into a strongly banded structure at higher reductions. For the as-extruded material, which already exhibits a fibrous morphology, cold rolling further reduces the thickness of these pre-existing bands. By 90% reduction, the microstructure in both UCR and CCR conditions appears as a dense, highly refined network of nearly parallel boundaries. High-magnification observations show that the final state consists of fine deformation domains with thickness on the order of a few micrometers, reflecting the intensity of plane-strain compression regardless of the rolling path.

The initial microstructural state influences the kinetics of this subdivision process. In the as-cast condition, the breakdown of the equiaxed structure proceeds progressively, with boundary density increasing steadily as a function of strain. In contrast, the as-extruded material starts with a higher fraction of HAGBs and a fibrous structure, which provides effective sites for dislocation storage. During the early stages of rolling, this leads to a rapid increase in the LAGB fraction as the material adjusts to the new stress state. Although the initial HAGB fraction is higher in the extruded condition, the rapid formation of new LAGBs at low reductions reduces their relative proportion, indicating that early-stage deformation is dominated by substructure formation rather than by rotation of existing grains.

At high deformation levels, the microstructure approaches a saturated state of refinement where the influence of the initial condition and rolling path becomes less pronounced in the TD–ND plane. By 90% reduction, both alloys exhibit a highly refined, fibrous morphology with a dense boundary network. The misorientation distributions indicate that, although LAGBs remain dominant, there is a progressive increase in misorientation leading to the gradual transformation of some LAGBs into HAGBs. This cumulative process represents the final stage of cold deformation, where the original grains are replaced by a complex network of deformation-induced boundaries that define the refined and strongly banded microstructure.

4.3.2.2 Texture Evolution during Cold Rolling

Rolled as-cast condition

For both as-cast AA6063 and AA6082 alloys, the comparison between unidirectional cold rolling (UCR) and cross cold rolling (CCR) reveals that the strain path exerts a decisive influence on orientation evolution along the β -fiber. Under UCR, the volume-fraction results show a progressive strengthening of the S ($\{123\}\langle 634 \rangle$) component with increasing thickness reduction, while Bs and Goss/Bs remain comparatively weak. This behavior is consistent with the classical rotation sequence predicted for FCC rolling textures under near plane-strain compression, where grains tend to rotate along the β -fiber from Bs toward S and, at larger strains, toward Cu. According to the Taylor–Bishop–Hill framework, such a monotonic strain path promotes continuous lattice rotation controlled by the activity of the $\{111\}\langle 110 \rangle$ slip systems, allowing orientations to move toward the stable S position without necessarily reaching Cu at the investigated strain levels (Mondal & Singh, 2012).

When the same alloys are processed by CCR, the texture evolution changes significantly. The experimental volume-fraction plots indicate that Bs and mixed Goss/Bs components become dominant at higher reductions, whereas the S fraction collapses to roughly 6–9% at 90% reduction, with the Brass component jumping to 47% in AA6082 (compared to just 18% under UCR). Importantly, this demonstrates that the transition to an alloy-type texture is not solely dependent on a low stacking fault energy (SFE), as the path change alone is sufficient to drive this phenomenon in high-SFE aluminum alloys. A similar strain-path-dependent transition has been reported for Ni–16Cr–8Fe alloys, where cross-rolling sequences lead to a Bs-dominated “alloy-type” texture, while monotonic rolling promotes stronger S and Cu components (Mehta, 2023). The introduction of alternating rolling directions modifies the effective loading path and prevents the continuous rotation toward S, resulting instead in the stabilization of intermediate orientations along the β -fibre.

This difference between UCR and CCR can be rationalized in terms of changes in the active slip systems and their relative resolved shear stresses. Mehta et al. (Mehta, 2023) showed that altering the rolling plane modifies the relative resolved shear stress (RRSS) on the octahedral

slip systems, which determines whether deformation proceeds by single slip, double slip, or cross-slip. Under monotonic rolling, the RRSS distribution favors slip combinations that drive orientations toward S along the β -fiber. In contrast, when the rolling direction is changed between passes, the stress state is repeatedly reoriented, increasing the probability of activating multiple slip systems simultaneously. This reduces the stability of the S orientation and promotes the accumulation of orientations around Bs and related variants.

In addition, cross rolling alters the geometrical relationship between the elongated cast microstructure and the imposed strain path. The repeated 90° rotation relaxes the strict plane-strain condition that exists in UCR and increases the likelihood of double-slip and cross-slip deformation. As suggested by Mehta et al. (Mehta, 2023), such non-proportional strain paths favour orientations that accommodate deformation through planar slip on fewer systems, which stabilizes Bs-type components while suppressing the continuous rotation toward S. Consequently, the transition from an S-dominated texture in UCR to a Bs/Goss-dominated texture in CCR can be interpreted as a direct consequence of the strain-path-dependent redistribution of RRSS and the resulting change in slip activity along the β -fibre.

Although the same fundamental crystallographic mechanisms apply to both alloys, the relative intensity of the resulting texture components differs because of their different alloy chemistry. In AA6082, the higher solute content and lower stacking-fault energy enhance planar slip and restrict cross-slip, which makes Bs-type orientations more stable once formed. As a result, the Bs dominance under CCR is generally more pronounced in AA6082 than in AA6063. A comparable transition from “metal-type” textures to “alloy-type” textures under cross-rolling has been reported by Mehta et al. (Mehta, 2023), who showed that non-proportional strain paths combined with lower effective stacking-fault energy promote planar slip and stabilize Bs-type orientations. Nevertheless, the overall interpretation remains the same for both alloys: UCR promotes the classical Bs $\rightarrow$ S rotation path under a monotonic strain path, whereas CCR redistributes the RRSS and slip activity in a way that suppresses continuous rotation toward S and favors the stabilization of Bs and Goss/Bs components.

Rolled as-extruded AA6082

The transition from the $\langle 111 \rangle$ extrusion fiber to a mature rolling texture occurs through progressive lattice reorientation driven by plastic deformation. Unlike the as-cast material, which exhibits a nearly random initial orientation distribution and rotates directly toward stable rolling components, the as-extruded AA6082 alloy, the extrusion direction (ED) is characterized by a strong inherited fiber texture, mainly composed of $\langle 111 \rangle$ and $\langle 001 \rangle$ orientations parallel to the extrusion direction (ED). Because cold rolling is performed perpendicular to the extrusion direction ($RD \perp ED$), this initial crystallographic orientation becomes highly unstable under the imposed rolling strain state. The fundamental deformation mode, consisting of compression along the normal direction (ND) and extension along the rolling direction (RD), requires the widespread activation of $\{111\}\langle 110 \rangle$ slip systems. The activation of these systems forces a gradual, geometric rotation of the crystal lattice away from the extrusion fiber and toward orientations that are compatible with plane-strain compression.

Under unidirectional cold rolling (UCR), the monotonic strain path promotes a continuous and predictable rotation of grains along the β -fiber. Grains belonging to the initial $\langle 111 \rangle$ extrusion fiber progressively rotate toward typical metal-type rolling components, primarily the S ($\{123\}\langle 634 \rangle$) and Copper ($\{112\}\langle 111 \rangle$) orientations. Because this required rotation path is relatively long in orientation space, the inherited extrusion texture undergoes a temporary weakening at intermediate strain levels. This phenomenon is clearly observed as a distinct decrease in overall pole-figure intensity at approximately 40% thickness reduction. This intermediate decrease represents a critical transition stage where the original extrusion fiber is heavily destabilized, yet the terminal β -fiber rolling texture has not yet fully matured.

In contrast, cross cold rolling (CCR) fundamentally modifies the crystallographic rotation path due to the 90° change in strain direction between successive passes. The alternating RD/TD strain state actively interrupts the continuous rotation along the β -fiber and alters the thermodynamic stability of the developing rolling components. As mentioned in the as-cast condition, this alternating strain path periodically redistributes the relative resolved shear stresses (RRSS) on the active slip systems. Under these non-proportional loading conditions, the continuous path toward the S component is disrupted, rendering it significantly less stable.

Instead, the Brass orientation ($\{011\}\langle 211\rangle$) exhibits a much higher compatibility with the alternating strain path. Consequently, the rotation of the grain structure tends to converge toward and lock at the Brass component. This mechanism explains the overwhelmingly strong Brass volume fractions observed at high reductions in the CCR condition, ultimately confirming that the macroscopic strain path exerts a direct and decisive influence on the crystallographic rotation sequence during cold deformation.

4.3.2.3 Grain Refinement during Cold Rolling

The grain-size distributions support the microstructural evolution observed in the EBSD maps, showing progressive refinement with increasing reduction. In the AA6082-UCR condition, accelerated grain refinement beyond the 40% reduction threshold is driven by severe strain-induced subdivision. Rather than deforming uniformly, different regions of the coarse as-cast grains rotate toward distinct stable orientations, a phenomenon driven by the monotonic strain path discussed in Section 4.3.2.2. This localized lattice rotation builds up high internal misorientations, effectively partitioning the original grains into smaller, independent units separated by newly formed deformation-induced high-angle grain boundaries (HAGBs). Coupled with geometric thinning along the normal direction (ND), this continuous subdivision explains the drastic drop in the 85% cumulative grain size fraction observed at 90% UCR.

Conversely, the cross-cold rolling (CCR) strain path alters this fragmentation behavior. While CCR also drives progressive refinement and structural homogenization at higher strain levels, the 90° rotation between passes repeatedly shifts the principal stress axes, which delays uniform grain fragmentation at intermediate stages. This alternating deformation suppresses the sharp crystallographic alignment and rapid misorientation buildup seen in UCR, instead promoting a more equiaxed microstructure. Ultimately, the synergy between this distributed lattice rotation and the alternating strain path explains the transition of the AA6082 alloy into a more stable and isotropic state under CCR conditions.

4.3.3 Conclusions on Cold Rolling

4.3.3.1 Microstructural Evolution and Grain Refinement

Regardless of the initial material state or alloy composition, cold rolling drives progressive structural refinement through the accumulation of dislocations, the formation of dense LAGB networks, and their gradual conversion into HAGBs. By 90% reduction, both UCR and CCR conditions converge into a highly refined, heavily banded morphology. However, the kinetics of this fragmentation depend on the strain path. Under UCR, the monotonic continuous deformation promotes rapid, severe strain-induced subdivision, leading to a reduction in grain size early in the deformation process. Conversely, the 90° rotation between successive passes in CCR repeatedly shifts the principal stress axes. This alternating strain path disrupts the continuous buildup of aligned deformation boundaries, effectively delaying the fragmentation process at intermediate stages. As a result, while CCR ultimately achieves a similarly saturated fibrous structure at severe strains, it proceeds at a more gradual rate, yielding characteristic grain sizes that remain slightly larger than those achieved under UCR at comparable deformation levels.

4.3.3.2 Texture Evolution and Crystallographic Rotation

The most profound impact of the strain path is observed in the crystallographic texture. Under UCR, all investigated conditions exhibit a classic "metal-type" rolling texture progression. The monotonic plane-strain compression facilitates continuous lattice rotation along the β -fiber, ultimately leading to a strong dominance of the S-component at high reductions. In contrast, CCR triggers a fundamental transition to an "alloy-type" texture dominated by the Brass (Bs) component. The orthogonal deformation path interrupts the continuous rotation toward the S-position by periodically redistributing the relative resolved shear stresses (RRSS) on the active slip systems, effectively destabilizing the S-component and locking the grains into the Brass orientation. This confirms that the transition to a Brass-dominated texture is not strictly reliant

on a low stacking-fault energy, but can be driven entirely by a non-proportional strain path in high-SFE aluminum alloys.

4.3.3.3 Influence of Initial State and Alloy Chemistry

While the strain path dictates the final state, the initial microstructure and alloy chemistry regulate the transition kinetics. The strong initial $\langle 111 \rangle$ and $\langle 001 \rangle$ fiber textures in the as-extruded AA6082 temporarily resist reorientation, requiring an intermediate destabilization phase before conforming to the rolling-induced paths. Chemically, while both AA6063 and AA6082 follow the same fundamental crystallographic rules, the higher solute content in AA6082 enhances planar slip and restricts cross-slip. This chemical difference amplifies the stabilization of the Brass component under CCR, resulting in a much sharper alloy-type texture in AA6082 than in the leaner AA6063 matrix.

4.3.3.4 Statement on Cold Rolling

Ultimately, these findings establish that varying the cold-rolling strain path provides a highly effective thermomechanical tool. While UCR maximizes rapid grain fragmentation and S-texture alignment, CCR can be strategically employed to produce a more isotropic structural morphology and a Brass-dominant texture, offering a direct pathway to tailor the mechanical anisotropy of 6xxx series aluminum alloys without altering their chemical composition.

4.4 Pseudo-hot Rolling

4.4.1 Introduction

This chapter focuses on pseudo-hot rolling and aims to examine how AA6082 and AA6063 respond to hot deformation in terms of microstructure and texture evolution. The primary objective is to evaluate how these two alloys behave when processed through different rolling routes, specifically unidirectional (UHR) and cross (CHR) pseudo-hot rolling, and when

starting from distinct initial states, including as-cast and as-extruded conditions. By systematically investigating these variables, this section establishes a clear framework for comparing the influence of strain path and initial material condition on the resulting deformation and restoration behavior.

4.4.2 Pseudo-hot Rolling Results

4.4.2.1 Pseudo-hot Rolled As-cast AA6082

Microstructure Evolution during Pseudo-hot Rolling

At 40% reduction, the UHR microstructure of rolled as-cast AA6082 remains heterogeneous. The EBSD orientation map shows the coexistence of coarse and finer grains, with clear orientation contrast between neighboring regions (Figure 4-41a). The corresponding band contrast map reveals a well-developed intragranular substructure, where LAGBs subdivide the grain interiors and HAGBs define the primary grain boundaries (Figure 4-41d). This morphology is consistent with the grain-size statistics. Relative to the initial as-cast state, which exhibited a maximum grain size of about 222 μm and an 85% cumulative size of about 183 μm , the 40% UHR condition still shows a broad grain-size distribution, with a maximum grain size of about 202 μm , while 85% of the grains are smaller than about 121 μm (Figure 4-42).

At 60% reduction, the microstructure becomes more deformed, and the grains show a more pancaked morphology, with slight elongation along the TD (Figure 4-41b). The corresponding band contrast map indicates a lower density of intragranular LAGBs compared with the 40% condition, suggesting partial recovery or boundary rearrangement during continued hot deformation (Figure 4-41e). This change is also reflected in the grain-size statistics. The cumulative grain-size distribution shifts toward finer sizes relative to 40%, with the maximum grain size decreasing slightly to about 189 μm and the 85% cumulative grain size decreasing to about 106 μm (Figure 4-42).

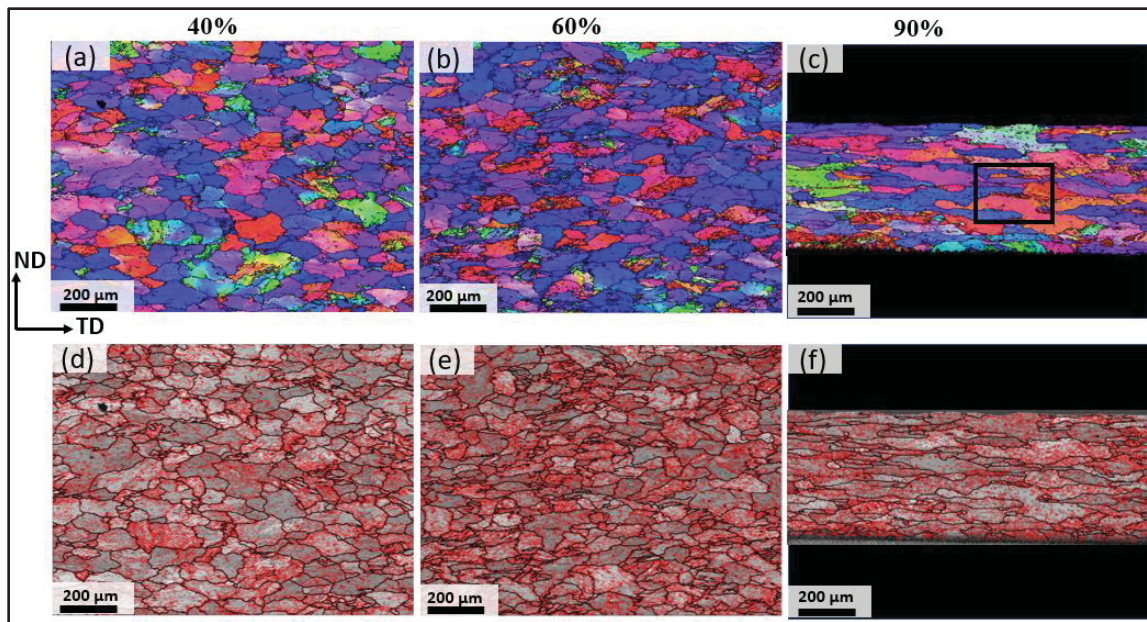


Figure 4-41 EBSD orientation maps (a–c) and corresponding band contrast maps (d–f) of rolled as-cast AA6082 processed by unidirectional pseudo-hot rolling (UHR) at thickness reductions of 40% (a,d), 60% (b,e), and 90% (c,f). The black rectangular box in (c) highlights the region shown at higher magnification in Figure 4-43a. All maps are acquired on sections perpendicular to the rolling direction e.g. TD-ND plane

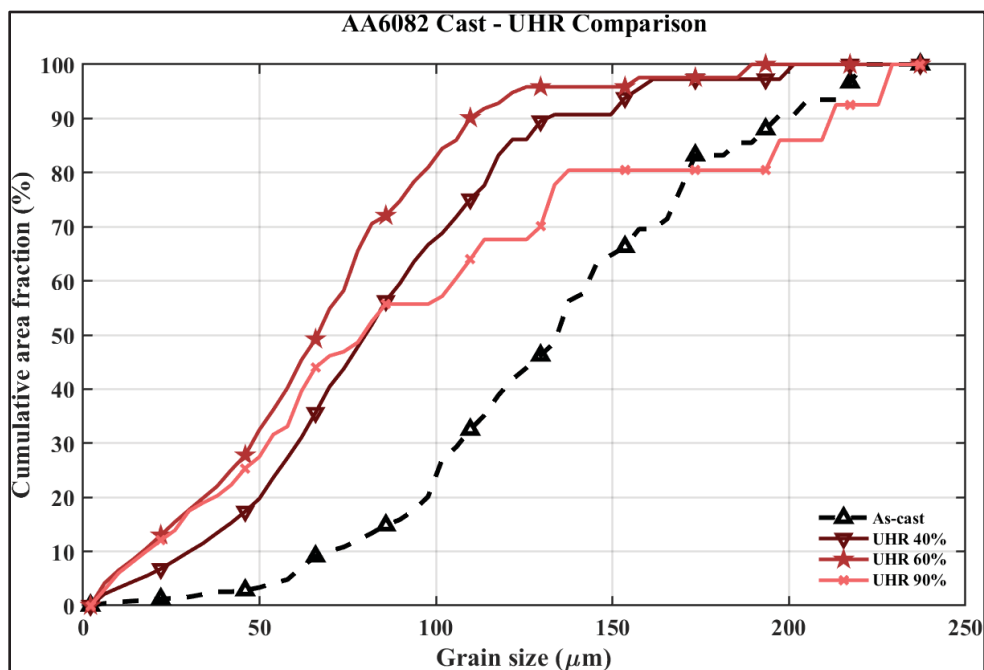


Figure 4-42 Cumulative grain size distribution curves for as-cast AA6082 under unidirectional pseudo-hot rolling (UHR) at various thickness reductions

At 90% reduction, a clear transition in the UHR microstructure is observed. The EBSD orientation map, which captures the entire sample thickness at this high reduction level, reveals a strongly deformed, flattened grain structure in the TD–ND section, together with pronounced grain subdivision (Figure 4-41c). The corresponding band contrast map shows a dense and continuous boundary network, where HAGBs delineate the elongated grain boundaries and LAGBs are distributed throughout the grain interiors (Figure 4-41f). A higher-magnification view of the highlighted rectangular region from Figure 4-41c reveals local grain-boundary bulging within the circled area within the circled area (Figure 4-43a), characterized by a localized protrusion and curvature of the boundary between highly flattened grains. This morphology suggests the onset of local boundary migration, potentially associated with an early stage of recrystallization. Despite this boundary instability, the grain-size statistics indicate that the refinement trend under UHR is not monotonic with increasing reduction. Although the 60% condition shows the strongest shift toward finer grain sizes, the area-weighted distribution at 90% remains strongly influenced by the persistence of large elongated grains. Consequently, the grain-size distribution shifts back toward coarser sizes relative to the 60% condition, with a maximum grain size of about 227 μm and 85% of the grains remaining below about 185 μm (Figure 4-42).

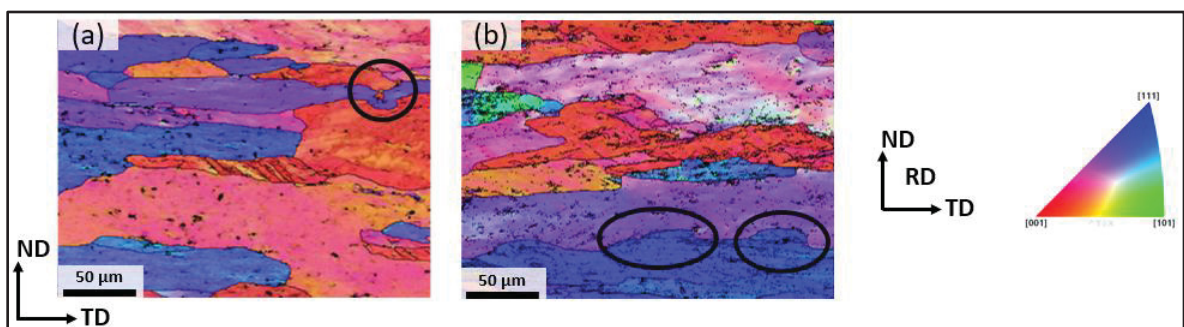


Figure 4-43 High-magnification EBSD orientation maps of rolled as-cast AA6082 at 90% reduction: (a) UHR and (b) CHR. The circled regions highlight local boundary protrusions consistent with grain-boundary bulging. For the CHR condition, the analyzed cross-section is taken perpendicular to RD1.

For the CHR condition, the microstructure at 40% reduction is characterized by more elongated grains compared to the UHR condition. This is due to the intrinsic nature of cross rolling, where after each pass the sample is rotated 90° to do the next pass (Figure 4-44a). The band contrast map reveals a dense and homogeneously distributed boundary network, with both HAGBs and LAGBs continuously present across the scanned area (Figure 4-44d). The cumulative grain-size distribution confirms this refinement relative to the initial as-cast state, with a maximum grain size of $\sim 162\ \mu\text{m}$ and 85% of the grains smaller than $\sim 105\ \mu\text{m}$ (Figure 4-45).

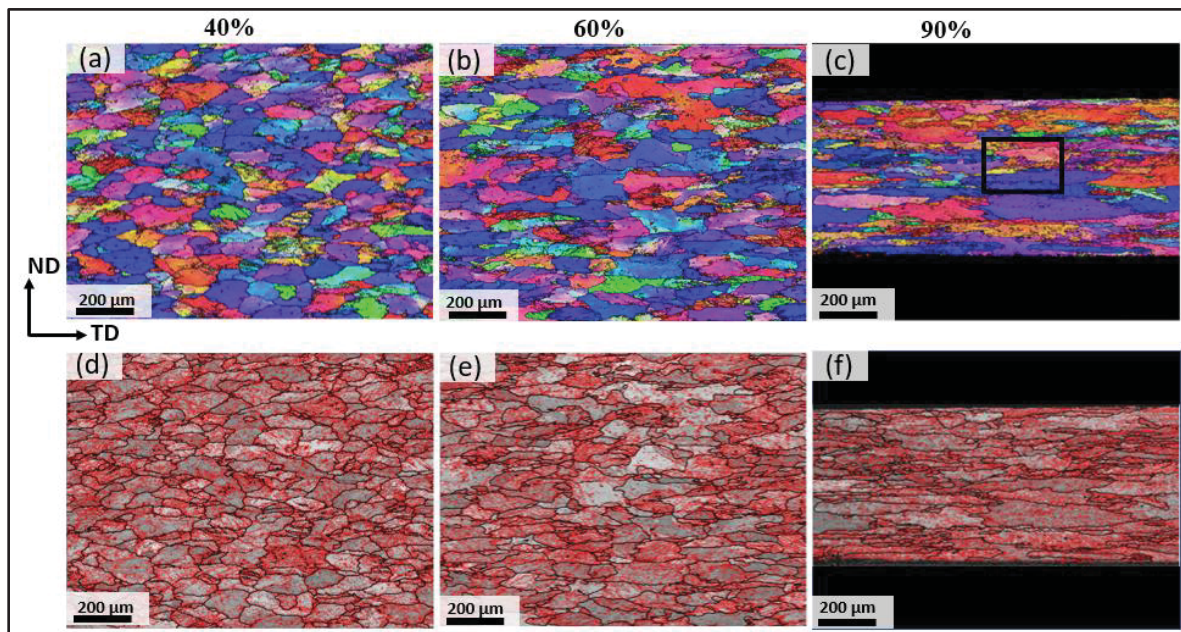


Figure 4-44 EBSD orientation maps (a–c) and corresponding band contrast maps (d–f) of rolled as-cast AA6082 processed by cross pseudo-hot rolling (CHR) at thickness reductions of 40% (a,d), 60% (b,e), and 90% (c,f). All maps are acquired on sections perpendicular to the primary rolling direction(R1) in the TD-ND plane

This elongation is more visible at 60% however it seems that the grains have a bit larger size compared to 40%. At 60% CHR, the EBSD map indicates a modest increase in area-weighted grain size relative to 40%, with elongated grains becoming more visible (Figure 4-44b). The band contrast map shows a reduced density of intragranular LAGBs, suggesting partial annihilation or rearrangement of boundaries during subsequent cross-rolling passes (Figure

4-44e). Quantitatively, the cumulative grain-size distribution shifts toward larger sizes, with a maximum grain size of $\sim 239 \mu\text{m}$ and 85% of the grains below $\sim 134 \mu\text{m}$ (Figure 4-45).

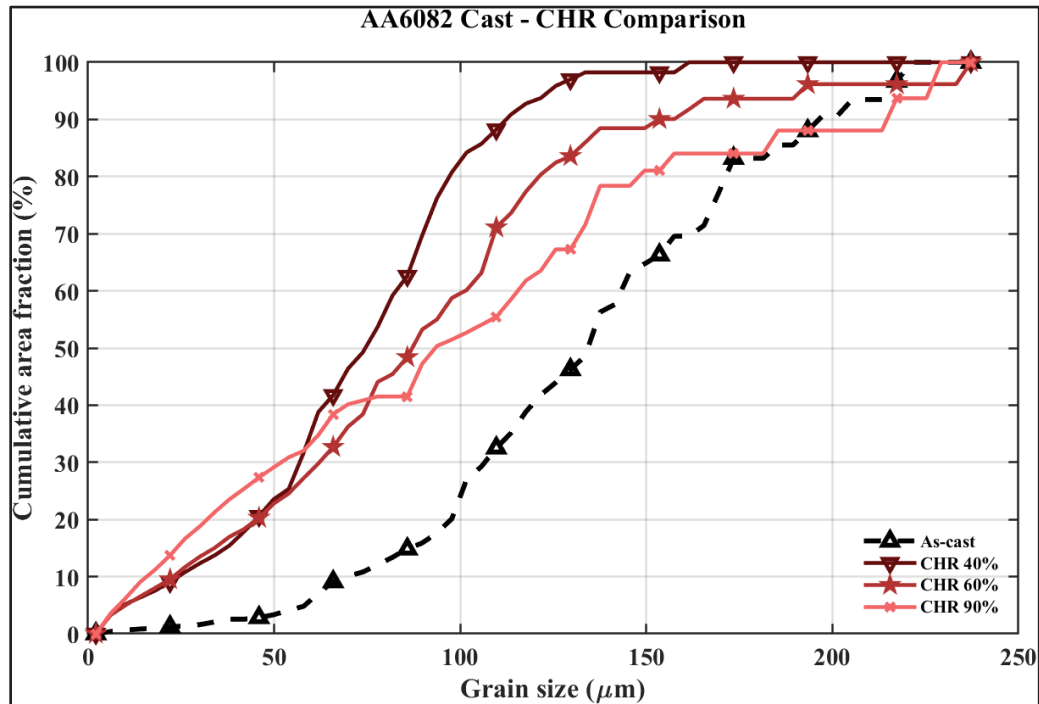


Figure 4-45 Cumulative grain size distribution curves for as-cast AA6082 under cross pseudo-hot rolling (CHR) at various thickness reductions

At 90% reduction, the CHR microstructure again exhibits localized grain-boundary instability. The EBSD orientation map shows elongated grains across the full cross-section in the TD-ND plane, but with less extreme elongation than in the UHR condition (Figure 4-44c). The band contrast map displays a well-developed and continuous boundary network (Figure 4-44f). A higher-magnification view of the highlighted rectangular region from Figure 4-44c reveals local grain-boundary bulging within the circled areas within the circled areas ((Figure 4-43b). Specifically, several localized boundary protrusions are observed along the elongated grain boundaries, suggesting that boundary migration occurs at multiple locations under the CHR condition. However, similar to the UHR condition, the area-weighted grain-size statistics do not reflect an overall refinement. Instead, the area-weighted distribution shifts further toward coarser sizes compared to the 60% condition, with the maximum grain size at $\sim 230 \mu\text{m}$ and 85% of the grains remaining smaller than $\sim 165 \mu\text{m}$ (Figure 4-45).

Texture analysis during pseudo-hot rolling

For rolled as-cast AA6082, the $\{111\}$ pole figures (Figure 4-46) show a clearer strengthening from 40% to 60% in UHR, followed by a redistribution at 90%. The UHR intensity increases from $\text{Max} \approx 3.1$ at 40% to $\text{Max} \approx 4$ at 60%, then remains pronounced at 90% ($\text{Max} \approx 3.4$). In CHR, the pole figures stay more moderate and more evenly distributed ($\text{Max} \approx 2.5\text{--}2.8$), indicating that the 90° rotation between passes limits strong sharpening along a single rolling path and promotes a more balanced texture.

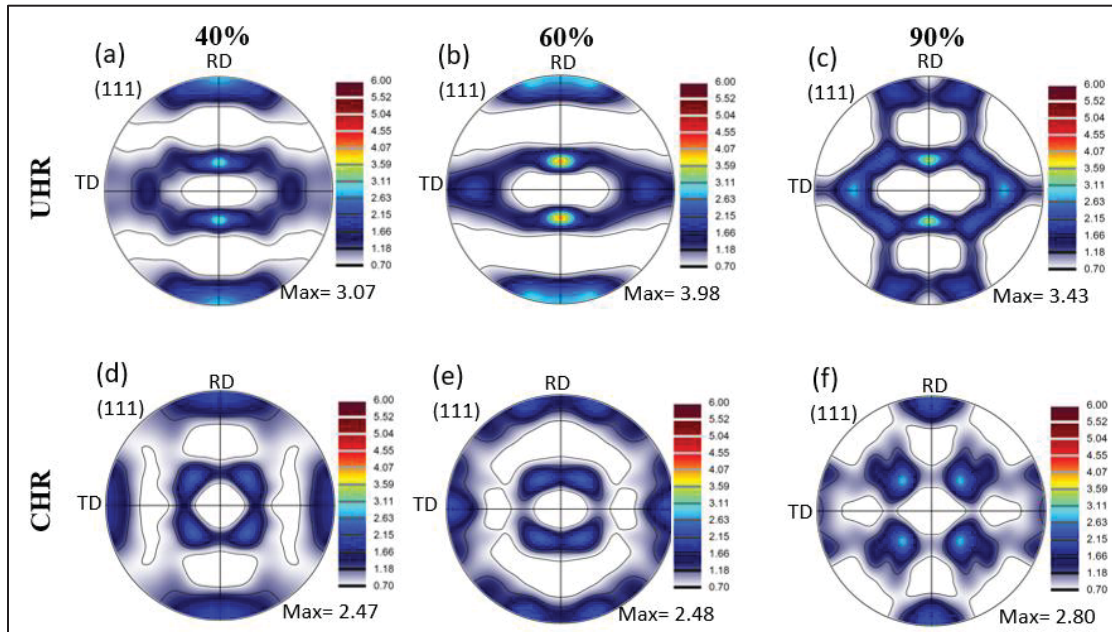


Figure 4-46 $\{111\}$ pole figures of as-cast AA6082 subjected to unidirectional pseudo-hot rolling (UHR) at various thickness reductions: (a) 40%, (b) 60%, (c) 90%, and for Cross pseudo-hot rolling (CHR) (d) 40%, (e) 60%, and (f) 90%

The volume-fraction results for as-cast AA6082 (Figure 4-47) show that S and Bs are again key components, with a noticeably higher S fraction in UHR than in AA6063. In UHR, S is dominant across 40–90% and peaks around $\sim 30\%$ at 60%, while Bs stays relatively constant in the $\sim 9\text{--}12\%$ range. In CHR, Bs is more prominent at 40–60% ($\sim 19\%$) and decreases toward 90% ($\sim 14\%$), whereas S remains lower than in UHR ($\sim 10\%$ at 40% and $\sim 7\%$ at 90%). Overall,

UHR promotes a stronger, more concentrated rolling texture, while CHR maintains a more moderate and redistributed texture state.

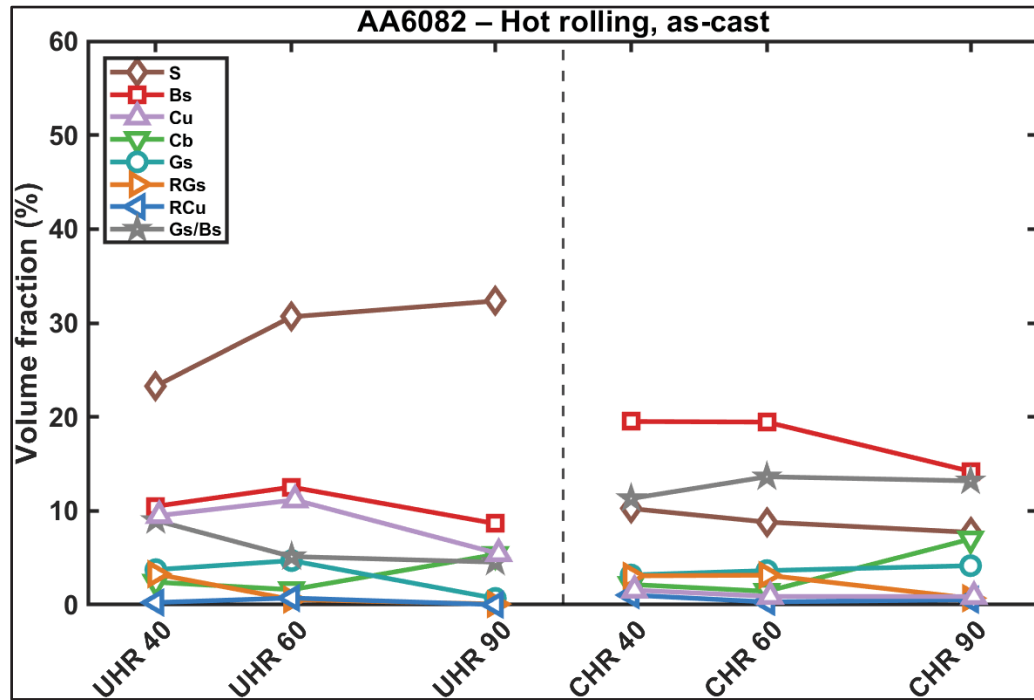


Figure 4-47 Volume fraction of selected texture components in as-cast AA6082 as a function of thickness reduction under unidirectional and cross pseudo-hot rolling

4.4.2.2 Pseudo-hot Rolled As-extruded AA6082

Microstructure Evolution during Pseudo-hot Rolling

The EBSD orientation and corresponding band contrast maps of hot-rolled as-extruded AA6082 are presented in Figure 4-48 and Figure 4-49. Figure 4-48 shows the UHR conditions at 40% and 60% reductions, while Figure 4-49 compares the UHR and CHR conditions at 90% reduction. All maps were acquired on TD–ND sections, perpendicular to the rolling direction. For the UHR samples, the rolling direction is parallel to the transverse direction of the extruded bar. For the CHR sample, the section perpendicular to the primary rolling direction (R1) was selected.

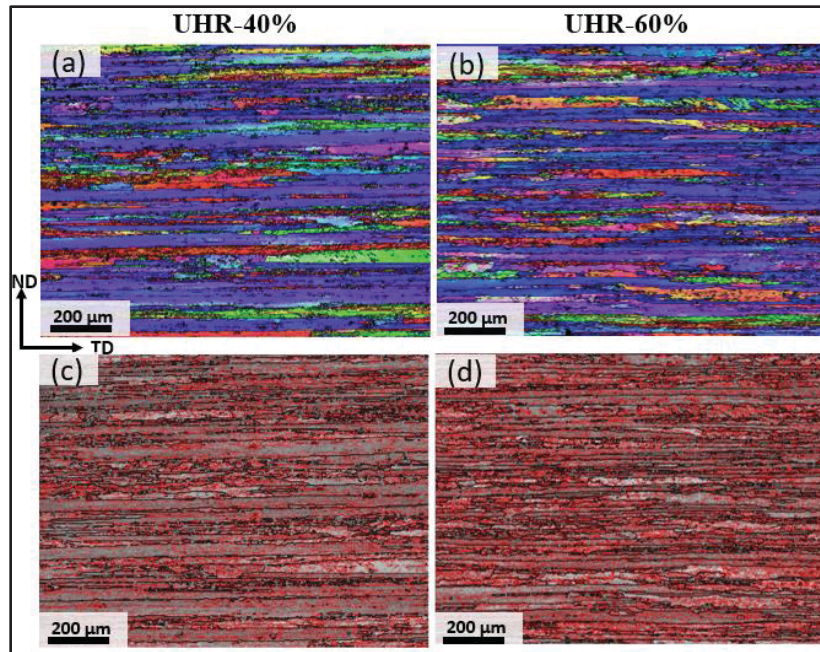


Figure 4-48 EBSD orientation maps (a, b) and corresponding band contrast maps (c, d) of as-extruded AA6082 after unidirectional pseudo-hot rolling (UHR) to 40% (a, c) and 60% (b, d) thickness reductions. All sections are taken on the TD–ND plane, perpendicular to the rolling direction

At UHR 40% reduction, the microstructure is dominated by strongly elongated grains aligned along the TD (Figure 4-48a), which corresponds to the extrusion direction prior to rolling (Figure 4-2). This elongated morphology reflects the inherited extrusion texture and grain shape, indicating that rolling perpendicular to the extrusion direction has a limited effect on disrupting the pre-existing grain structure at this reduction level. The band contrast map (Figure 4-48c) reveals a dense intragranular substructure, with numerous LAGBs aligned within elongated grains, while HAGBs define the primary grain boundaries.

At 60% reduction, the elongated grain morphology remains clearly visible in the TD-ND section (Figure 4-48b), indicating that the extrusion-induced grain shape is still largely preserved. With increasing reduction, the grains appear more flattened (pancaked), while the overall elongation along TD persists. The band contrast map (Figure 4-48d) shows a dense and continuous LAGB network within grains, suggesting increased deformation substructure;

however, the dominant grain shape remains strongly influenced by the prior extrusion rather than by the rolling deformation.

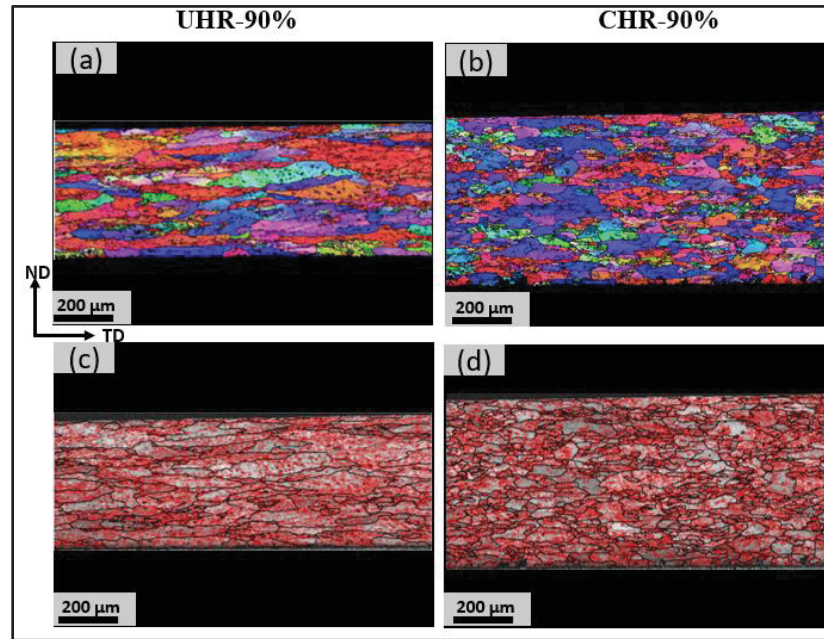


Figure 4-49 EBSD orientation maps (a, b) and band contrast maps (c, d) of as-extruded AA6082 after unidirectional pseudo-hot rolling (UHR) and cross pseudo-hot rolling (CHR) to 90% thickness reduction. The UHR condition is shown in (a, c), and the CHR condition is shown in (b, d). All sections are taken on the TD–ND plane, perpendicular to the rolling direction

At 90% reduction, a clear change in grain morphology is observed across the entire sample thickness (Figure 4-49a). The elongated grains inherited from extrusion are noticeably shortened, and the microstructure appears less directionally elongated in the TD-ND plane. The band contrast map (Figure 4-48c) reveals a dense and well-distributed subgrain network, indicating extensive grain subdivision and the increasing dominance of rolling-induced deformation over the initial extrusion structure. This suggests that only at high accumulated strain does UHR become effective in reducing extrusion-induced grain elongation.

Unlike the UHR-90% condition, the CHR-90% microstructure does not exhibit a pronounced elongation along the original extrusion direction (Figure 4-49b). Instead, the grains appear more equiaxed and less directionally aligned in the TD-ND plane across the entire cross

section, indicating that the repeated 90° rotation between rolling passes further weakens the inherited extrusion-induced anisotropy. The corresponding band contrast map (Figure 4-49d) displays a dense and uniformly distributed boundary network, with LAGBs and HAGBs well developed throughout the scanned area. Thereby, these observations indicate that when rolling is performed perpendicular to the extrusion direction, very high reductions are required under UHR conditions to noticeably reduce the inherited elongated grain structure, whereas CHR at the same reduction level is more effective in suppressing extrusion-induced anisotropy and promoting a more isotropic microstructure in as-extruded AA6082.

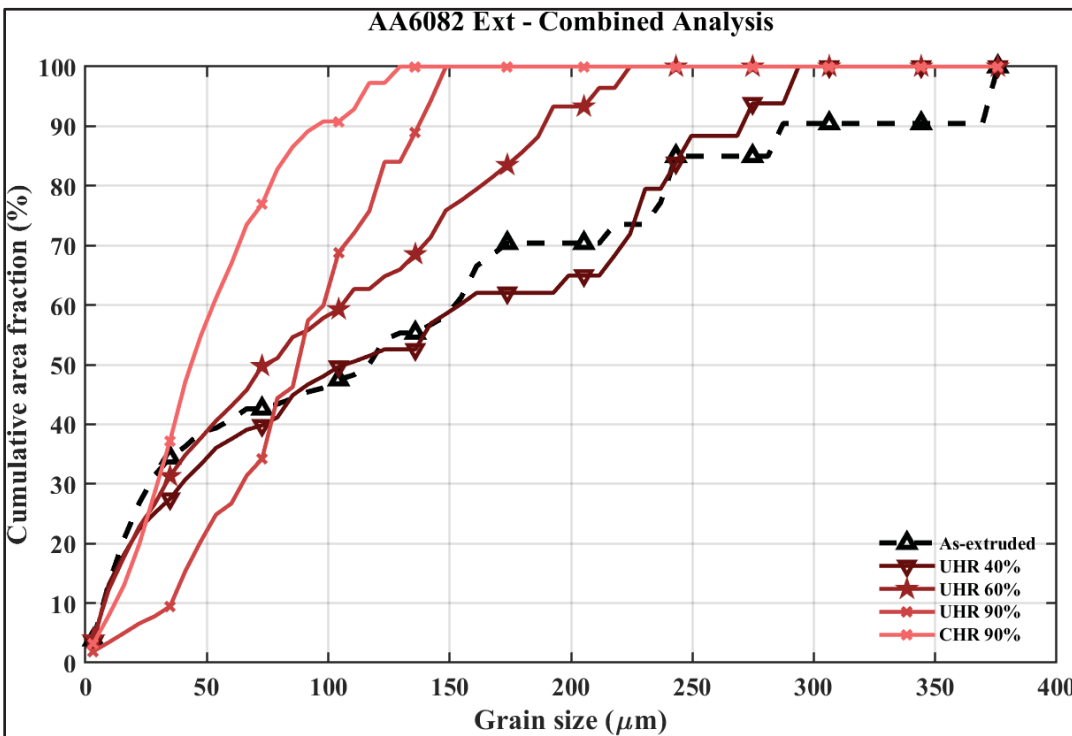


Figure 4-50 Cumulative grain size distribution curves for as-extruded AA6082 under unidirectional pseudo-hot rolling (UHR) and cross pseudo-hot rolling (CHR) at various thickness reductions

Grain-size statistics further support these observations (Figure 4-50). Compared with the initial as-extruded state, which had a maximum grain size of about $379 \mu\text{m}$ and an 85% cumulative threshold of about $243 \mu\text{m}$, the UHR-40% and UHR-60% conditions still show broad cumulative grain-size distributions. This is consistent with the presence of large, elongated grains occupying a substantial fraction of the scanned area. The maximum grain size decreases

from about 292 μm at UHR-40% to about 226 μm at UHR-60%, while the 85% threshold decreases from about 247 μm to about 177 μm . At UHR-90%, the maximum grain size further decreases to about 147 μm , and the 85% threshold decreases to about 127 μm , indicating a greater contribution from finer grains. In the CHR-90% condition, the maximum grain size is reduced to about 130 μm , and 85% of the grains are smaller than about 88 μm . This confirms a substantially finer and more homogeneous grain-size distribution compared with UHR-90%.

Texture analysis during pseudo-hot rolling

For the as-extruded AA6082, the $\{111\}$ pole figures (Figure 4-51) show the strongest texture at the lower UHR reductions, followed by a clear weakening by 90% reduction. The UHR-40% condition exhibits the highest intensity (Max ≈ 8.3), which decreases at UHR-60% (Max ≈ 5.7) and drops substantially at UHR-90% (Max ≈ 3.0), indicating a marked reduction in texture sharpness at high accumulated strain. The CHR-90% condition is weaker still (Max ≈ 2.2), consistent with high-temperature deformation combined with pass-to-pass rotation redistributing orientations and reducing the dominance of the initial extrusion-related texture.

The component volume fractions (Figure 4-52) are consistent with this evolution. At UHR-40% and UHR-60%, the texture is strongly dominated by Bs, which is very high at 40% ($\approx 55\%$) and remains substantial at 60% ($\approx 39\%$). By UHR-90%, Bs drops to an almost negligible level, while Cb increases markedly to about 21%, and the remaining components remain at low-to-moderate levels. The S component is initially low ($\approx 4\%$), increases to $\approx 12\%$ at 60%, and then decreases again at 90%. At CHR-90%, the component distribution is more balanced and the overall texture remains weak. In this condition, S becomes the largest component ($\approx 11\%$) and Gs/Bs is also significant ($\approx 9\%$). The cube component is around 7% in CHR-90%, while it remains much higher in UHR-90% ($\approx 21\%$). In addition, Bs is still present in CHR-90% ($\approx 6\%$), whereas it is nearly absent in UHR-90%. Thereby, these trends support the pole-figure observation that cross rolling is more effective at suppressing the inherited extrusion texture at the highest reduction.

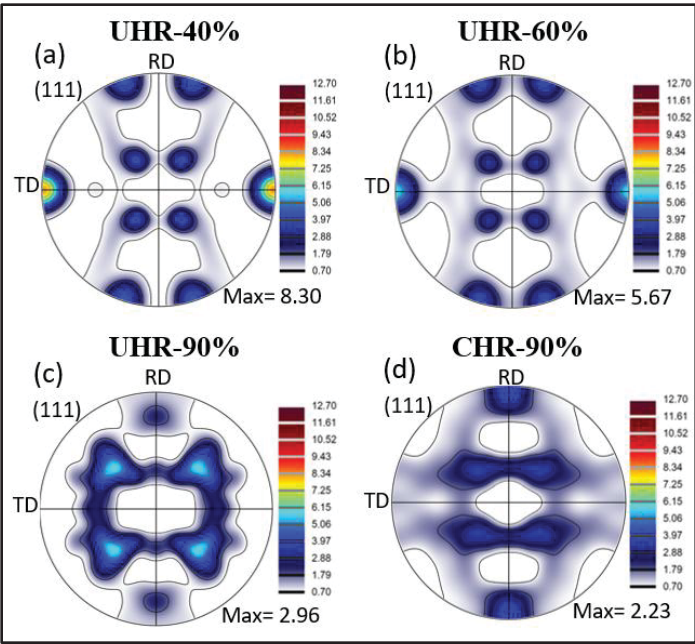


Figure 4-51 {111} pole figures of as-extruded AA6082 subjected to unidirectional pseudo-hot rolling (UHR) at various thickness reductions: (a) 40%, (b) 60%, (c) 90%, and for Cross pseudo-hot rolling (CHR) (d) 90%

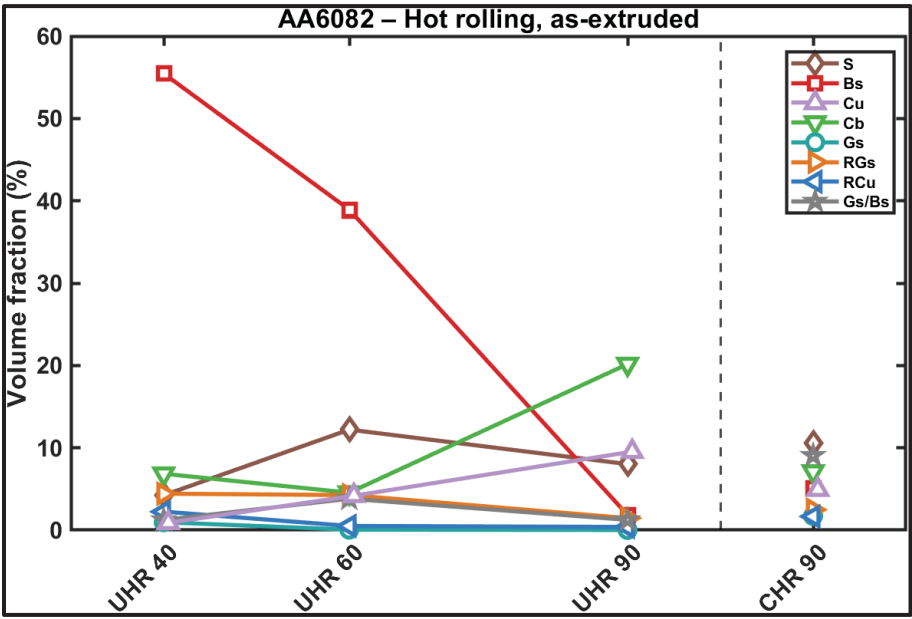


Figure 4-52 Volume fraction of selected texture components in as-extruded AA6082 as a function of thickness reduction under unidirectional and cross pseudo-hot rolling

4.4.2.3 Pseudo-hot Rolled As-cast AA6063

Microstructure Evolution during Pseudo-hot Rolling

The evolution of the UHR microstructure is characterized by persistent heterogeneity, where the initial stages of pseudo-hot rolling lead to a coarser area-weighted grain size compared to the as-cast material. The system gradually shifts toward refinement only at extreme reductions. At 40% reduction, a coarse and heterogeneous grain structure is observed, where a limited number of large grains dominate and orientation contrast persists between neighboring grains (Figure 4-53a). The corresponding band contrast map reveals an intragranular substructure in which LAGBs are partitioned within HAGB perimeters (Figure 4-53d). Grain-size statistics are consistent with this morphology. Compared with the initial as-cast state (maximum grain size $\sim 153\ \mu\text{m}$, 85% threshold $\sim 127\ \mu\text{m}$), the cumulative grain-size distribution shifts toward larger sizes and exhibits a broad spread with a pronounced coarse-grain tail. The maximum grain size increases to $\sim 316\ \mu\text{m}$, while 85% of the grains are smaller than $\sim 216\ \mu\text{m}$ (Figure 4-54).

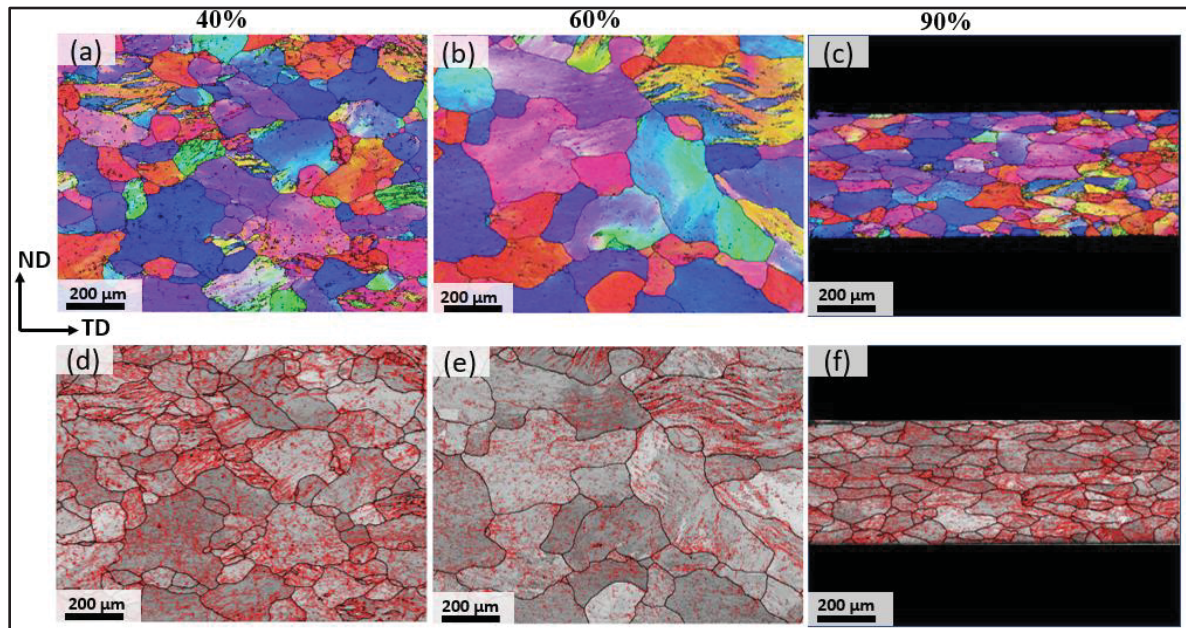


Figure 4-53 EBSD orientation maps (a–c) and corresponding band contrast maps (d–f) of rolled as-cast AA6063 processed by unidirectional pseudo-hot rolling (UHR) at thickness reductions of 40% (a,d), 60% (b,e), and 90% (c,f). All maps are acquired on sections perpendicular to the rolling direction in TD-ND plane

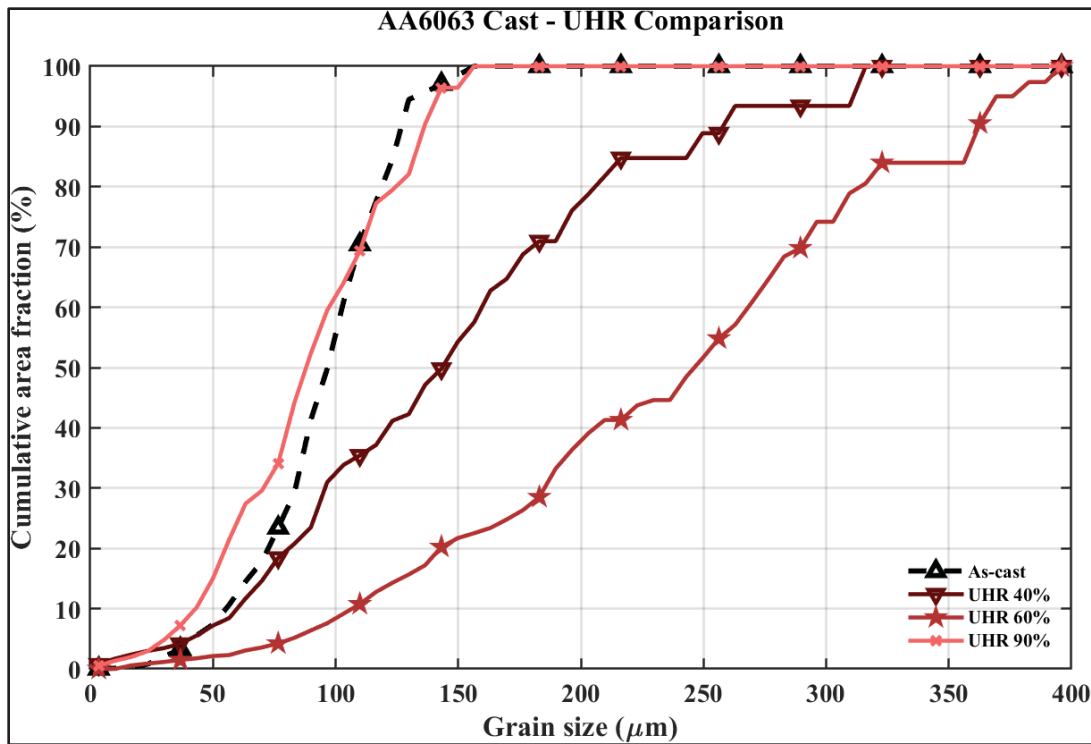


Figure 4-54 Cumulative grain size distribution curves for as-cast AA6063 under unidirectional pseudo-hot rolling (UHR) at various thickness reductions

At 60% reduction, the UHR microstructure remains coarse, with the EBSD map still showing large, continuous grains with limited internal orientation subdivision compared with 40% (Figure 4-53b). The area-carrying grains are larger than in the 40% condition, while the band contrast map shows a reduced density of intragranular LAGBs despite the higher reduction level (Figure 4-53e). Quantitatively, the cumulative grain-size distribution shifts further toward larger sizes, with the maximum grain size increasing to $\sim 399 \mu\text{m}$ and 85% of the grains remaining below $\sim 342 \mu\text{m}$ (Figure 4-54). It is only at 90% reduction that the UHR microstructure transitions to a markedly different morphology, with the full cross-section showing a strongly deformed, flattened (pancaked) grain structure in the TD–ND section with substantially increased subdivision (Figure 4-53c). The band contrast map reveals a dense boundary network, where HAGBs form continuous grain boundaries and LAGBs are distributed within grains (Figure 4-53f). This change is reflected in the grain-size statistics, where the cumulative distribution shifts toward smaller sizes compared with the 40% and 60%

conditions, returning to values comparable to the initial as-cast state. The maximum grain size decreases to $\sim 156 \mu\text{m}$, while 85% of the grains are smaller than $\sim 134 \mu\text{m}$ (Figure 4-54).

For the CHR condition, a more refined and uniformly distributed grain structure is observed at 40% reduction when compared to its UHR counterpart, though it still remains coarser than the initial as-cast state. The EBSD orientation map shows a greater number of smaller grains in the TD-ND section corresponding to the TD-ND plane of the first rolling pass (RD1) (Figure 4-55a). The band contrast map exhibits a denser and more uniformly distributed boundary network, where HAGBs and LAGBs are continuously distributed across the scanned area (Figure 4-55d). The cumulative grain-size distribution is shifted toward finer sizes compared with 40%-UHR, with a maximum grain size of approximately $\sim 238 \mu\text{m}$ and 85% of the grains smaller than $\sim 181 \mu\text{m}$ (Figure 4-56).

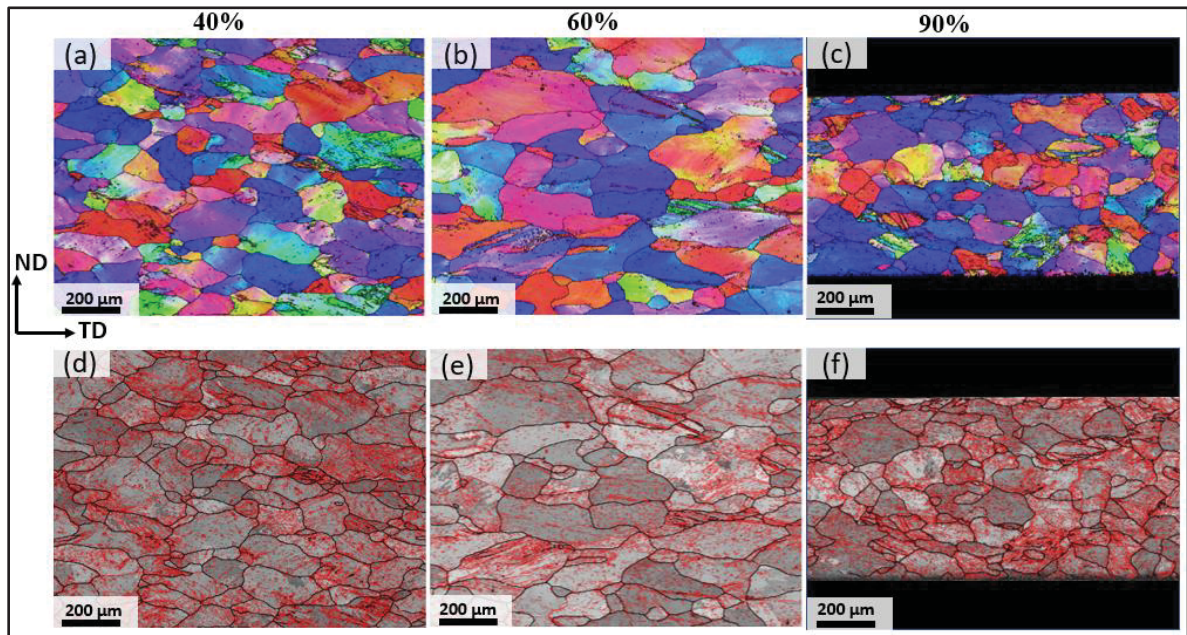


Figure 4-55 EBSD orientation maps (a–c) and corresponding band contrast maps (d–f) of rolled as-cast AA6063 processed by cross pseudo-hot rolling (CHR) at thickness reductions of 40% (a,d), 60% (b,e), and 90% (c,f). All maps are acquired on sections perpendicular to the primary rolling direction(R1)

At 60% CHR, the EBSD map indicates an increase in area-weighted grain size relative to 40%, with elongated grains becoming more visible (Figure 4-55b). The band contrast map shows a lower density of LAGBs, with many intragranular boundaries diminishing after subsequent

passes (Figure 4-55e). Quantitatively, the cumulative grain-size distribution shifts toward larger sizes, with a maximum grain size of approximately $\sim 312 \mu\text{m}$ and 85% of the grains below $\sim 249 \mu\text{m}$ (Figure 4-56).

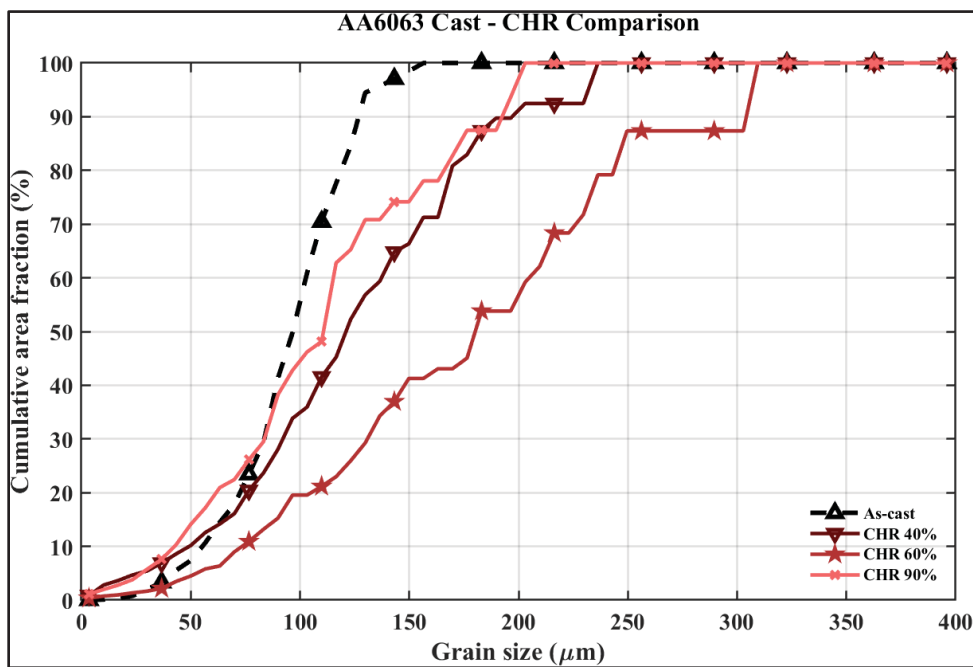


Figure 4-56 Cumulative grain size distribution curves for as-cast AA6063 under cross pseudo-hot rolling (CHR) at various thickness reductions

At 90% reduction, the CHR microstructure again exhibits refinement relative to the 60% condition, with flattened grains visible across the entire TD-ND cross-section and a well-developed boundary network (Figure 4-55c, Figure 4-55f). Compared with 90% UHR, the grain shape appears less elongated, which is consistent with the cross-rolling sequence: the repeated 90° rotation redistributes strain and prevents strong localization that would otherwise intensify elongation along one dominant strain path. However, while the grains are less elongated, the area-weighted grain size remains coarser than the 90% UHR condition. The grain-size statistics reflect this relative refinement from the 60% CHR state: the maximum grain size decreases to approximately $\sim 201 \mu\text{m}$, and 85% of the grains are smaller than $\sim 172 \mu\text{m}$ (Figure 4-56).

As shown by the EBSD results (Figure 4-53, Figure 4-55), the evolution of the deformation substructure during pseudo-hot rolling is non-monotonic. This behavior is evident between 40% and 60% reduction, where a lower density of LAGBs and locally coarser grains are observed despite the increase in total reduction. This trend differs from the expected progressive increase in grain subdivision with increasing strain. This non-monotonic behavior is consistent with the thermal exposure imposed during the pseudo-hot rolling procedure. To limit temperature loss, the sample was returned to the furnace and reheated to 400 °C between successive rolling passes. These inter-pass reheating steps provide time for recovery processes to occur during deformation processing, which can reduce the density of intragranular boundaries and contribute to local coarsening.

To further support this interpretation, an additional short heat treatment was applied to the 90% as-rolled UHR AA6063 condition (Figure 4-57). The specimen was reheated at 400 °C for 3 min and rescanned in the same EBSD field of view, allowing a direct grain-by-grain comparison before and after heat treatment. After reheating, the microstructure shows a clear reduction in LAGB density, together with a cleaner and more equiaxed grain structure. This observation indicates that the deformation substructure developed during rolling is thermally unstable at 400 °C and can be partially removed during short thermal exposure.

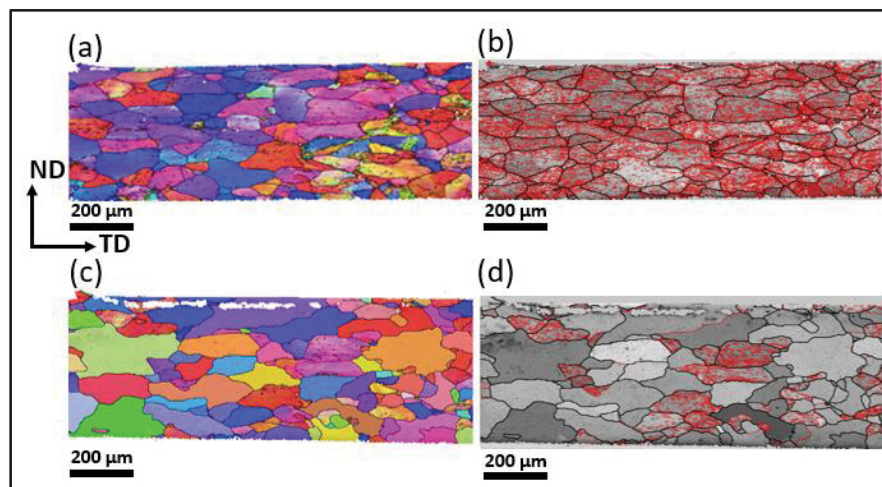


Figure 4-57 EBSD orientation maps (a, c) and corresponding band contrast maps (b, d) of as-cast AA6063 after UHR to 90% thickness reduction, before heat treatment (a, b) and after heat treatment at 400 °C for 3 min (c, d). The maps were acquired from the same EBSD field of view, enabling direct grain-by-grain comparison before and after heat treatment

Texture analysis during pseudo-hot rolling

The $\{111\}$ pole figures for rolled as-cast AA6063 (Figure 4-58) indicate that both UHR and CHR develop a typical rolling-type texture. At 40% reduction, the texture is weak in both routes (Max ≈ 2.2 for UHR and ≈ 2.3 for CHR). At 60% reduction, the UHR pole figure shows a clearer sharpening of the main features (Max ≈ 2.9), whereas CHR decreases slightly (Max ≈ 2.0), consistent with a less pronounced strengthening in the cross-rolling sequence. At 90% reduction, the pole figures keep its rolling-type texture, with maxima in the ~ 2.6 – 2.9 range (Max ≈ 2.6 for UHR and ≈ 2.8 for CHR), indicating a distributed rolling texture rather than dominance of a single strong component.

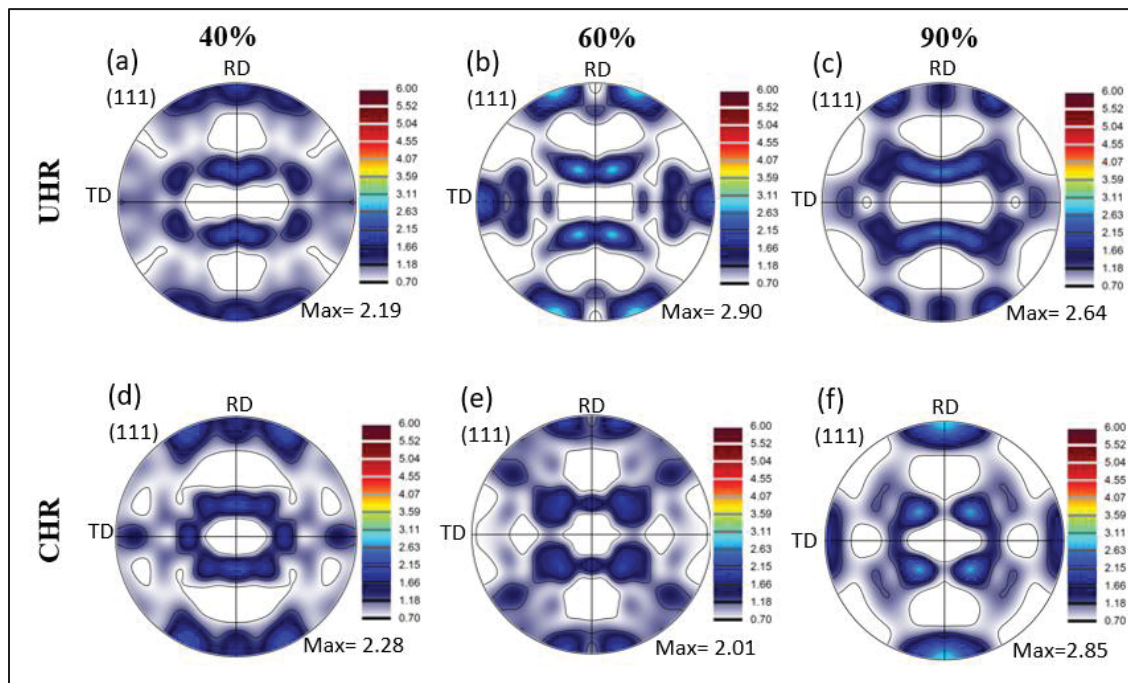


Figure 4-58 $\{111\}$ pole figures of rolled as-cast AA6063 subjected to unidirectional pseudo-hot rolling (UHR) at various thickness reductions: (a) 40%, (b) 60%, (c) 90%, and for Cross pseudo-hot rolling (CHR) (d) 40%, (e) 60%, and (f) 90%

The $\{111\}$ pole figure of the heat-treated UHR-90% condition (Figure 4-59) shows that the overall rolling-type texture is retained after heat treatment. The maximum intensity (Max ≈ 2.4) remains comparable to the as-rolled state, while the intensity distribution appears

less localized, with broader and less well-defined features. This indicates a slight increase in the randomness of the texture without the emergence of new dominant components.

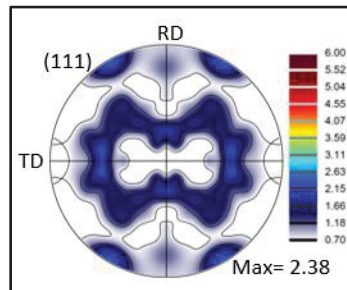


Figure 4-59 $\{111\}$ pole figure of as-cast AA6063 after UHR to 90% thickness reduction followed by heat treatment at 400 °C for 3 min

The texture-component volume fractions (Figure 4-60) support these observations. Across the 40%, 60%, and 90% reductions, S and Bs remain dominant in both rolling routes, typically within the ~10–15% range. In UHR, Cu is highest at 40% (~13%) and decreases sharply at 60% and 90%, while S remains relatively strong at 90% (~11%) and Bs rises with increasing reduction, reaching ~14% at 90%. The Cb component shows only a modest increase. In CHR, Bs also increases and reaches its highest fraction at 90% reduction (~14%), while S decreases from 40% to 60% and then remains around ~13% at 90%. A key feature of the CHR route is the higher Gs/Bs fraction at 60% and 90%, reaching ~13% at 90%, indicating a more mixed texture state. No component exhibits a strong monotonic build-up across all conditions, which is consistent with the moderate texture strength observed in the pole figures.

For the heat-treated UHR-90% condition, the major β -fiber components remain present but with markedly reduced volume fractions. In particular, S decreases from ~11% to ~4%, Bs from ~14% to ~5%, and Cu from ~6% to ~3%, while Cb shows a more moderate reduction. Minor components such as RGs and RCu become negligible after heat treatment. Despite this overall reduction in component fractions, the main deformation-texture components are retained. These results indicate that heat treatment weakens the deformation texture rather than replacing it with a new dominant texture type, consistent with the recovery-related microstructural changes observed in Figure 4-57.

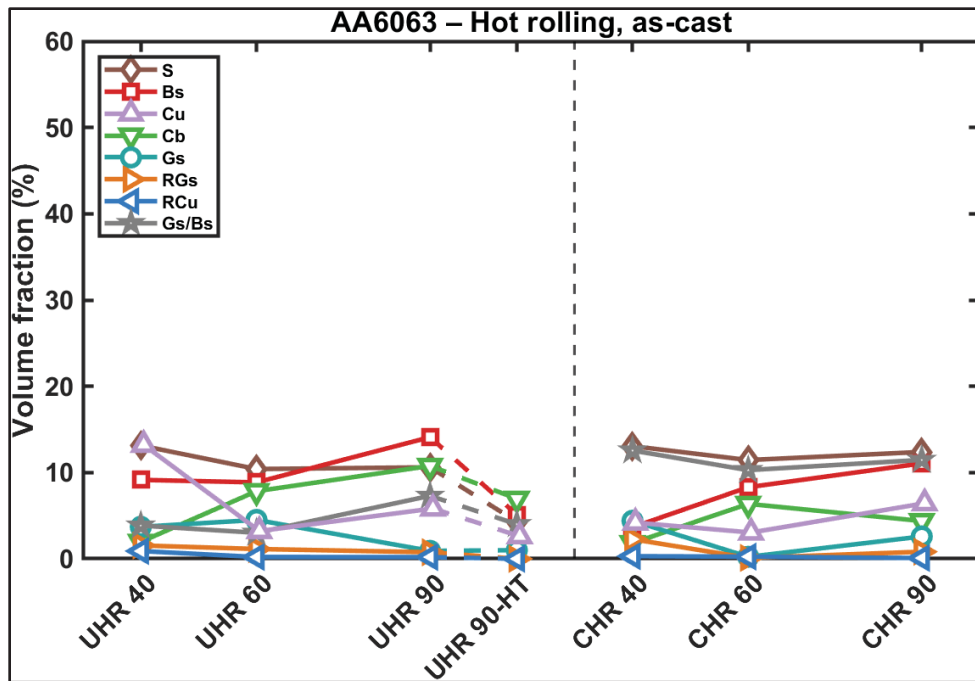


Figure 4-60 Volume fraction of selected texture components in as-cast AA6063 as a function of thickness reduction under unidirectional and cross pseudo-hot rolling

4.4.2.4 Second-phase Particle Distribution and Associated Grain Morphology during Pseudo-hot Rolling

The evolution of grain morphology in hot-rolled AA6082 and AA6063 is influenced by the presence and spatial distribution of second-phase particles, the imposed rolling path, and the associated microstructural evolution during processing. To avoid repetition, SEM-BSE images and corresponding EDS maps are presented only for selected representative conditions at 90% thickness reduction. At least three regions were examined for each selected condition, while one representative field is presented in the figures below.

In hot-rolled as-cast AA6082, pronounced particle-related features are observed at 90% reduction. The EDS maps show a higher apparent number of Mn-, Mg-, and Si-enriched features than those observed in rolled as-cast AA6063. Because the UHR and CHR conditions show comparable general elemental distributions, only the CHR condition is presented here as a representative example (Figure 4-61). The UHR condition showed the same overall elemental

characteristics across the examined regions, although the enriched features appeared less frequent than in the CHR condition.

At 90% reduction, EBSD orientation maps of rolled as-cast AA6082 show pronounced grain elongation in both rolling paths. In the UHR condition, the TD–ND section perpendicular to the rolling direction reveals strongly flattened and elongated grains (Figure 4-41c). In the CHR condition, the TD–ND section taken with respect to the primary rolling direction also shows elongated grains at 90% reduction (Figure 4-44c). This elongated grain morphology coincides with the higher apparent presence of Mn-, Mg-, and Si-enriched features observed in the SEM–EDS maps.

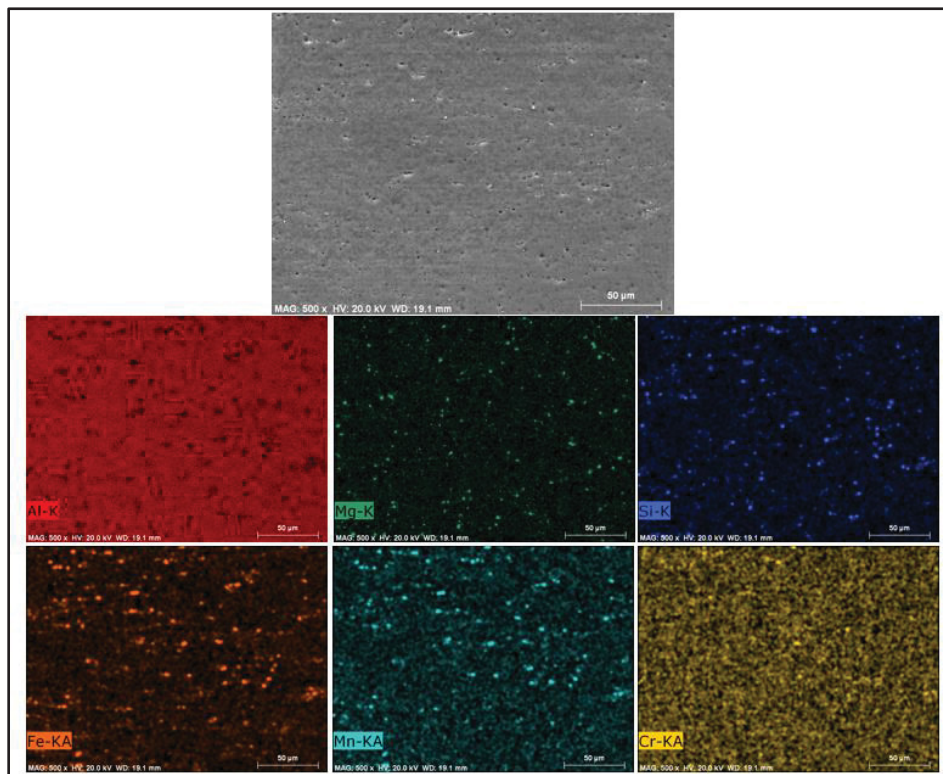


Figure 4-61 SEM backscattered electron (BSE) image and corresponding EDS elemental maps of AA6082 processed by cross pseudo-hot rolling (CHR) in the as-cast condition at 90% thickness reduction

In hot-rolled as-extruded AA6082, the SEM-BSE image and corresponding EDS maps for the UHR condition at 90% reduction show second-phase particles arranged in elongated,

directionally aligned bands (Figure 4-62). This alignment follows the prior extrusion direction and remains visible after pseudo-hot rolling, indicating that the particle distribution inherited from the strong extrusion deformation is largely retained. In the CHR condition at 90% reduction, fewer second-phase particles are visually apparent in the examined SEM fields, and the remaining features exhibit a weaker directional alignment than in UHR (Figure 4-63).

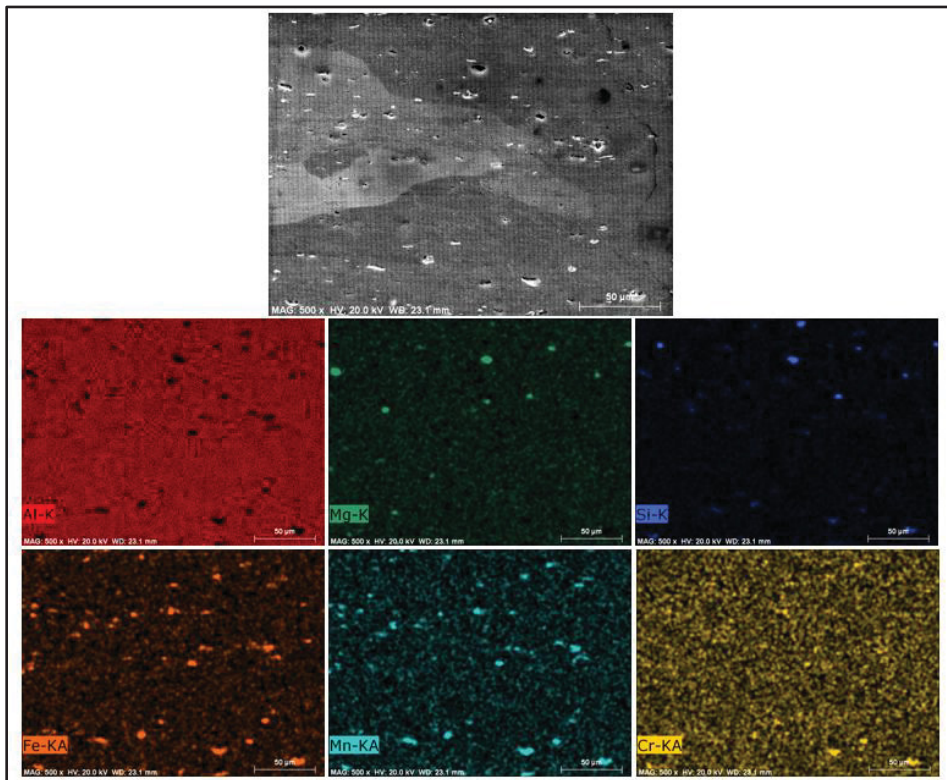


Figure 4-62 SEM backscattered electron (BSE) image and corresponding EDS elemental maps of the as-extruded AA6082 sample, unidirectionally pseudo-hot rolled along the transverse direction (TD) to 90% thickness reduction

The EBSD maps show a consistent trend in grain morphology (Figure 4-49). At 90% reduction, the UHR condition exhibits a pronounced directional microstructure with elongated grains aligned along the prior extrusion direction (Figure 4-49 a,c), consistent with the directionally banded particle distribution observed by SEM–EDS. In contrast, the CHR condition displays reduced grain elongation and a more fragmented, less directionally aligned structure at the same reduction (Figure 4-49 b,d). Following cross pseudo-hot rolling, the directional particle

banding becomes less pronounced, and fewer clearly aligned particles are observed in the examined SEM fields, coinciding with the reduced grain elongation observed in the CHR microstructure.

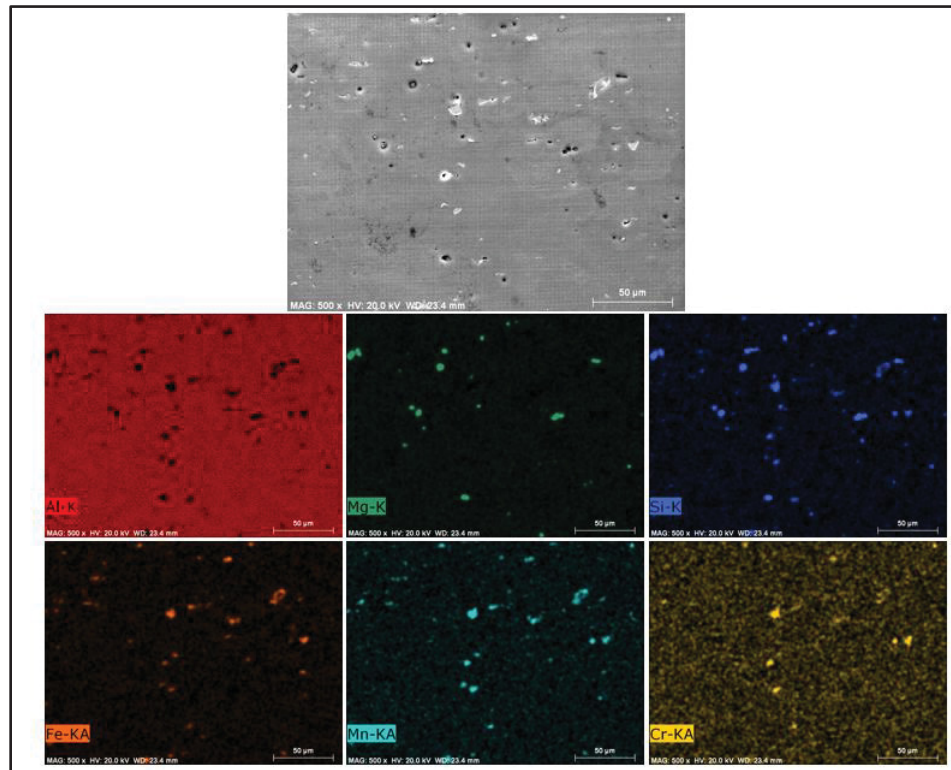


Figure 4-63 SEM backscattered electron (BSE) image and corresponding EDS elemental maps of as-extruded AA6082 processed by cross pseudo-hot rolling (CHR) to 90% thickness reduction, where the first rolling pass was applied parallel to the transverse direction (RD1 || TD)

In hot-rolled as-cast AA6063, the SEM-BSE image shows a relatively clean matrix with a limited number of bright-contrast features. The corresponding EDS maps indicate that Fe-containing particles are present but sparsely distributed across the scanned area. No clear alignment, banding, or directional arrangement of second-phase particles is observed. Since the UHR and CHR conditions showed the same general features, only the CHR condition is presented here as a representative example (Figure 4-64). The UHR condition exhibited a similar clean matrix, sparse Fe-containing particles, and no clear directional alignment or particle banding.

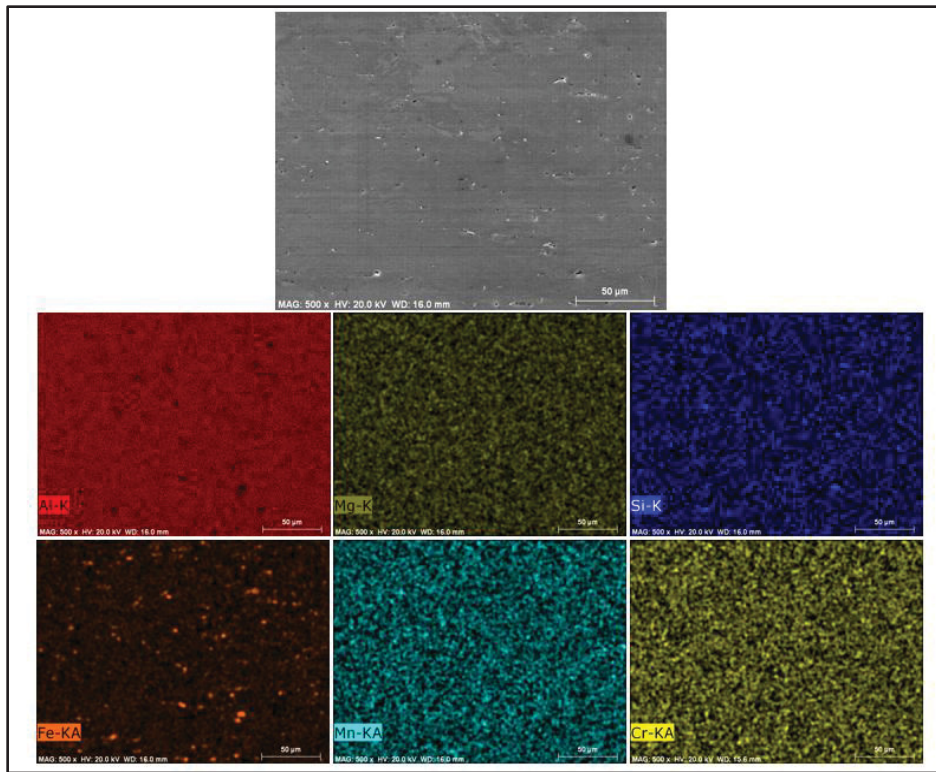


Figure 4-64 SEM backscattered electron (BSE) image and corresponding EDS elemental maps of AA6063 processed by cross pseudo-hot rolling (CHR) in the as-cast condition at 90% thickness reduction

This relatively uniform and dilute particle distribution is consistent with the grain morphologies observed by EBSD (Figure 4-53 and Figure 4-55). In both UHR and CHR conditions, rolled as-cast AA6063 does not exhibit strongly elongated grains, even at high reductions. Instead, the microstructure remains heterogeneous at lower reductions and evolves toward flattened, but not strongly elongated, grains at 90% reduction.

When the selected conditions are compared at 90% reduction, the apparent particle distribution differs with alloy composition and initial processing condition. Rolled as-cast AA6082 shows a greater apparent occurrence of Mn-, Mg-, and Si-enriched features in the examined SEM–EDS fields than AA6063. This difference reflects the greater apparent occurrence of Mn-, Mg-, and Si-enriched features in AA6082 within the examined SEM–EDS fields. This makes particle fragmentation and redistribution during severe deformation more visible in the two-dimensional SEM–EDS maps. In the as-extruded AA6082 condition, the particle distribution

is also influenced by prior thermomechanical processing. Under UHR, directionally aligned particle bands remain visible, while CHR reduces the persistence of this alignment. In contrast, rolled as-cast AA6063 shows a relatively low particle population. Therefore, the SEM–EDS maps reveal limited contrast and no clear directional particle arrangement. These observations indicate that the apparent particle distribution is affected by alloy chemistry, initial microstructural state, and imposed rolling path.

4.4.2.5 Orientation-based Slip Metrics under RD Loading during Pseudo-hot Rolling

To complement the crystallographic texture characterization presented previously, orientation-based slip metrics were extracted from the EBSD orientation dataset assuming uniaxial loading along the rolling direction (RD). Two quantitative indicators were evaluated: (i) the global average Taylor factor (TF_{avg}) and (ii) the Schmid-factor-based area fractions of grains classified as soft ($SF > 0.45$), medium ($0.35\text{--}0.45$), and hard ($SF < 0.35$). These metrics reorganize the measured grain orientations according to their slip favorability under a fixed loading direction and provide a statistical description of how the orientation population partitions into different slip-compatibility classes (Table 4-2).

Rolled as-cast AA6082 exhibits a different response. Under UHR, TF_{avg} decreases from 3.39 at 40% to 3.23 at 90%. The Schmid-factor partition indicates a strong decrease in the hard-grain fraction ($22\% \rightarrow 10\%$) together with an increase in the soft fraction ($14\% \rightarrow 20\%$), while the medium class remains dominant. In other words, increasing reduction in AA6082-UHR reduces the hard-grain subset under RD loading and increases the proportion of grains favorably oriented for slip. Under CHR, AA6082 shows the strongest shift in slip favorability: TF_{avg} decreases from 3.34 at 40% to 3.11 at 90%, while the soft fraction increases from 24% to 32% and the hard fraction decreases sharply from 18% to 6%. Among all investigated conditions, AA6082-CHR at 90% reduction exhibits the lowest hard-grain fraction and the highest soft-grain fraction, indicating the most pronounced redistribution toward slip-favorable orientations under RD loading.

Table 4-2 Average Taylor factor (TF_{avg}) and Schmid-factor-based area fractions of soft ($SF > 0.45$), medium ($0.35-0.45$), and hard ($SF < 0.35$) grains under RD loading at 40% and 90% reductions

No	Condition	Average Taylor Factor	Schmid factor		
			(> 0.45)%	(0.35–0.45)%	(< 0.35)%
1	AA6082-40% UHR as-cast	3.39	14	63	22
2	AA6082-90% UHR as-cast	3.23	20	70	10
3	AA6082-40% CHR as-cast	3.34	24	58	18
4	AA6082-90% CHR as-cast	3.11	32	62	6
5	AA6082-40% UHR as-extruded	3.35	20	66	15
6	AA6082-90% UHR as-extruded	3.14	20	65	15
7	AA6082-90% CHR as-extruded	3.20	30	56	14
8	AA6063-40% UHR as-cast	3.29	22	58	20
9	AA6063-90% UHR as-cast	3.21	14	71	15
10	AA6063-40% CHR as-cast	3.24	29	55	16
11	AA6063-90% CHR as-cast	3.31	16	67	16

For rolled as-extruded AA6082 processed by UHR, TF_{avg} decreases from 3.35 at 40% to 3.14 at 90%, but the Schmid-factor partitions remain unchanged across this interval (soft: 20%; hard: 15%). Thus, despite the increase in reduction and the decrease in TF_{avg} , the statistical partition of slip favorability remains comparatively stable in this condition. For the as-extruded CHR condition at 90% reduction, TF_{avg} equals 3.20, with a soft fraction of 30% and a hard fraction of approximately 14%.

For rolled as-cast AA6063 processed by UHR, TF_{avg} decreases slightly from 3.29 at 40% reduction to 3.21 at 90%. Over the same reduction range, both extreme Schmid-factor tails decrease: the soft fraction drops from 22% to 14% and the hard fraction drops from 20% to 15%, while the medium fraction increases from 58% to 71%. Thus, increasing reduction in AA6063-UHR is associated with a reduction of both very-soft and very-hard grains and a stronger dominance of the medium-slip class. In contrast, under CHR, AA6063 shows a slight increase in TF_{avg} from 3.24 at 40% to 3.31 at 90%. This is accompanied by a marked reduction in the soft fraction (29% $\rightarrow$ 16%), whereas the hard fraction remains approximately constant

(~16%). Therefore, AA6063 under CHR shifts away from very-soft orientations at high reduction without a comparable suppression of the hard-grain fraction.

4.4.3 Pseudo-hot Rolling Discussion

This chapter compared the pseudo-hot rolling responses of as-cast AA6082, as-extruded AA6082, and as-cast AA6063 under UHR and CHR strain paths. The discussion focuses on how alloy chemistry, initial microstructural state, and imposed strain path affect the balance between strain accumulation and restoration mechanisms during hot deformation. These mechanisms include dislocation storage, subgrain formation, dynamic recovery, boundary migration, and recrystallization-related processes. The following sections discuss these effects separately for AA6082 and AA6063, with particular attention to microstructural refinement, particle-related heterogeneity, texture evolution, and the influence of the rolling path.

4.4.3.1 Microstructural Evolution and Particle Effects in AA6082 during Pseudo-hot Rolling

As-cast AA6082 exhibits a relatively stable deformation substructure and a strong tendency toward microstructural refinement at high strain. Although some non-monotonic behavior is observed, such as the reduction in LAGB density from 40% to 60% reduction under UHR, the characteristic grain sizes remain substantially smaller than those of AA6063 at equivalent reductions. This indicates a greater resistance to recovery-induced coarsening. This higher stability is consistent with the greater solute content of AA6082 and the presence of Mn-containing dispersoids and second-phase particles, which can increase solute drag and particle-pinning effects. These effects restrict boundary migration and delay the removal of deformation substructure during pseudo-hot rolling.

At 90% reduction, the EBSD maps for both UHR and CHR show elongated and highly subdivided grains with dense boundary networks (Figure 4-41c and Figure 4-44c). The morphology observed at this stage is consistent with local grain-boundary bulging, as indicated by the localized curvature and protrusion of boundaries along the highly deformed elongated

grains. Such bulging is generally associated with differences in stored energy across neighboring regions, which provide a driving force for local boundary migration. Therefore, AA6082 appears to accumulate sufficient stored energy during severe pseudo-hot deformation to promote localized boundary migration and grain-boundary instability.

The presence of grain-boundary bulging in both UHR-90% (Figure 4-43a) and CHR-90% (Figure 4-43b) suggests that, under the imposed temperature and reduction conditions, the main controlling factor is the development of sufficiently large local stored-energy differences to promote boundary migration, rather than the strain path alone. However, the strain path still modifies the spatial distribution of deformation. In particular, CHR interrupts continuous directional strain accumulation and can therefore influence the distribution of local boundary instabilities and the final grain morphology.

The SEM–EDS observations further support the interpretation that second-phase particles affect restoration kinetics and grain morphology at high reductions. In AA6082, the representative CHR-90% map shows a higher apparent number of Mn-, Mg-, and Si-enriched features than AA6063, indicating a greater potential for particle-related heterogeneity and boundary-pinning effects. These features are consistent with the stronger substructure stability observed by EBSD and with the greater tendency of AA6082 to retain deformation substructure up to high reductions.

4.4.3.2 Effect of Initial State and Strain Path in As-extruded AA6082 during Pseudo-hot Rolling

The as-extruded AA6082 condition shows that the inherited microstructure strongly affects the early stages of pseudo-hot rolling. At UHR-40% and UHR-60%, the TD–ND sections remain dominated by elongated grains aligned with the pre-existing extrusion morphology (Figure 4-48). This indicates that the applied reductions are not sufficient to fully disrupt the extrusion-derived grain shape and associated orientation field. In this regime, deformation is mainly accommodated by intragranular subdivision within the elongated grains, while the overall grain morphology remains strongly influenced by the initial extruded state.

A transition is observed at UHR-90%, where the elongated morphology becomes less dominant and the microstructure shows increased subdivision and a greater tendency toward grain fragmentation (Figure 4-49a). This transition coincides with a marked redistribution of texture components (Figure 4-51), suggesting that the combined effect of high strain and restoration becomes sufficient to modify the inherited extrusion texture and microstructural state.

Cross pseudo-hot rolling is more effective in reducing extrusion-induced anisotropy at 90% reduction. The CHR-90% microstructure is more equiaxed and homogeneous than the corresponding UHR condition (Figure 4-49b). This behavior is consistent with the effect of strain-path changes, which interrupt the persistence of deformation alignment with the inherited elongated grain structure and particle arrangement. Therefore, CHR reduces the directional continuity of both grain morphology and orientation memory, leading to a more fragmented and less directionally aligned structure.

The SEM–EDS maps further support this interpretation. In the as-extruded AA6082 condition, UHR preserves directionally aligned particle bands, whereas CHR reduces the persistence of this alignment. This trend agrees with the EBSD observations, where UHR retains a more directional morphology, while CHR promotes a more homogeneous and less anisotropic microstructure.

4.4.3.3 Recovery-controlled Response in AA6063 during Pseudo-hot Rolling

A distinctive feature of as-cast AA6063 is the non-monotonic microstructural evolution between 40% and 60% reduction. Instead of showing a continuous increase in substructure density and progressive grain refinement with increasing strain, the EBSD results reveal a reduction in LAGB density and locally coarser area-carrying grains at 60% compared with 40% reduction (Figure 4-53). From a metallurgical perspective, this behavior suggests that restoration processes temporarily outpace strain accumulation at intermediate reductions.

This response is consistent with a recovery-dominated regime, where dislocation annihilation, subgrain rearrangement, and subgrain coarsening are relatively efficient at the rolling temperature. The representative SEM–EDS maps for AA6063 at 90% reduction reveal only a

sparse population of Fe-containing particles (Figure 4-64), indicating limited Zener pinning pressure on migrating boundaries and subgrain interfaces. Under these conditions, subgrain boundaries are less stable. LAGBs can be eliminated through recovery and boundary rearrangement, while selective growth of favorably oriented grains can proceed through strain-induced boundary migration. As a result, the microstructure may exhibit transient coarsening even though the macroscopic strain continues to increase.

The short annealing experiment applied to the UHR-90% condition provides direct evidence of this high restoration sensitivity. After reheating at 400 °C for 3 min, the rapid decrease in LAGB density and the development of a cleaner and more equiaxed morphology (Figure 4-57) indicate a substantial release of stored energy through recovery and boundary migration over a short time scale. This observation supports the interpretation that boundary mobility is relatively high in AA6063 and that the deformation substructure is not effectively stabilized. This explains why the expected monotonic refinement trend is interrupted between 40% and 60% reduction.

Compared with AA6082, the lower particle density in AA6063 is consistent with weaker pinning and higher boundary mobility. This promotes a recovery-dominated response and a stronger tendency for selective boundary migration, as observed in the EBSD results. Therefore, while AA6082 can retain deformation substructure more effectively and reach high-strain refinement conditions, AA6063 is more prone to dissipate stored energy at earlier stages through recovery and boundary migration.

4.4.3.4 Texture Evolution during Pseudo-hot Rolling

The pole figures and component analysis show that deformation textures dominate the as-cast conditions. These textures are mainly represented by rolling-type components such as S and Bs/Brass-related families. At high strain, however, restoration increasingly affects the orientation distribution. This effect is particularly clear in the as-extruded AA6082 condition. For as-cast AA6082, UHR produces stronger texture sharpening at intermediate reduction, with a maximum intensity of approximately 4 at 60%. This behavior is consistent with the

retention of deformation substructure during rolling. It also suggests that recovery-driven texture weakening is less pronounced in this condition. In contrast, CHR changes the imposed deformation path from pass to pass. As a result, the texture becomes more redistributed compared with UHR.

The evolution of the Brass component further shows the combined effect of strain path and initial state. In as-cast AA6082, Brass remains relatively stable but low under UHR, consistently below approximately 12%. Under CHR, the Brass fraction is higher at low reduction, reaching approximately 20% at CHR-40%, before decreasing at 90% reduction. This trend suggests that cross rolling can initially promote or retain Brass-related orientations, while severe deformation at 90% reduction reduces their stability. This behavior is consistent with reports that pass-to-pass 90° rotation can destabilize specific rolling components (Nayan et al., 2019). The pronounced decrease at 90% also suggests that severe deformation and competing restoration mechanisms, including recovery and localized grain-boundary migration, may contribute to the redistribution of orientations rather than allowing the Brass component to remain deformation-stable.

The most pronounced texture change occurs in the as-extruded AA6082 condition (Figure 4-52). At 40% reduction, the orientation distribution is dominated by Brass, with a fraction of approximately 55%. This confirms that the initial texture is strongly inherited from the extrusion process. By 90% reduction, however, the Brass component decreases to nearly negligible levels. At the same time, the Cube fraction increases substantially, reaching approximately 21% under UHR. This indicates that the inherited extrusion texture is strongly modified at high strain.

This change should not be interpreted as a direct transformation from Brass to Cube. The classical oriented-growth explanation for Cube development is often discussed in relation to Cube growth at the expense of S-oriented regions through favorable 40° $\langle 111 \rangle$ boundaries. Such a relationship is not generally expected for a Brass–Cube pair. Therefore, in the present work, the simultaneous decrease in Brass and increase in Cube are interpreted as evidence that the inherited extrusion texture is strongly modified during high-strain pseudo-hot rolling. The

EBSF texture results show that the inherited Brass-dominated state becomes unstable under severe hot deformation. At the same time, Cube-oriented volumes become more prominent during the subsequent restoration-influenced microstructural evolution.

In the CHR-90% condition, the Cube fraction remains much lower, at approximately 7%. This suggests that the pass-to-pass change in rolling direction limits the persistence of one dominant orientation path. Instead of promoting a Cube-dominant state, CHR leads to a more distributed texture. This agrees with the microstructural observations, where CHR produces a more fragmented and less directionally aligned structure than UHR in the as-extruded AA6082 condition.

For as-cast AA6063, the texture remains moderate and does not show strong monotonic sharpening. This behavior is consistent with the higher efficiency of recovery and boundary mobility in this alloy. During pseudo-hot rolling, these restoration processes can reduce orientation gradients and weaken the retention of deformation texture. Compared with AA6082, the weaker texture development in AA6063 supports the interpretation that deformation-induced orientation changes are less strongly retained.

4.4.3.5 Orientation-based Slip Metrics under RD Loading during Pseudo-hot Rolling

The orientation-based slip metrics presented in Section 4.4.2.5 provide a directional assessment of the measured orientation fields by quantifying their compatibility with uniaxial loading along the rolling direction (RD). In contrast to conventional texture analysis, which describes the distribution of orientations in crystallographic space, the Taylor factor and Schmid-factor statistics evaluate how these orientations respond to a specific loading direction. Therefore, this section discusses how strain path, reduction level, and initial material state affect the distribution of slip favorability under RD loading. These results are also related to the microstructural changes observed during pseudo-hot rolling.

The comparison between UHR and CHR shows that strain-path modification changes how orientations are distributed among different slip-favorability classes under RD loading. AA6082 exhibits a clear strain-path response. Under UHR, the fraction of hard grains

decreases strongly at high reduction. Under CHR, this effect becomes even more pronounced, with the hard-grain fraction decreasing to approximately 6%. This indicates that CHR is effective in reducing the fraction of orientations that are poorly oriented for slip under RD loading. In the as-extruded condition, the slip-favorability distribution is more resistant to change under UHR, suggesting that the inherited orientation structure limits the extent of redistribution during monotonic rolling.

An important result of this study is that large changes in texture component fractions do not necessarily lead to proportional changes in Taylor factor or Schmid-factor statistics. The as-extruded AA6082 condition provides a clear example. Between 40% and 90% UHR, the Bs component decreases sharply and the Cube component increases substantially. However, the soft, medium, and hard Schmid-factor fractions remain nearly unchanged. This shows that different crystallographic texture components can occupy similar ranges of slip favorability under RD loading. Therefore, component-based texture analysis and slip-metric analysis describe different aspects of orientation evolution. Texture components identify where orientations are concentrated in orientation space, whereas slip metrics indicate how those orientations interact with a specific loading direction.

In as-cast AA6063 under UHR, increasing reduction suppresses both very soft and very hard grains. As a result, the orientation population becomes concentrated in the medium-slip range. This behavior is consistent with progressive orientation redistribution toward states that are geometrically compatible with plane-strain rolling. In contrast, CHR in AA6063 reduces the soft fraction at high reduction and slightly increases the average Taylor factor. This indicates that the orthogonal strain-path change does not promote soft-orientation enrichment in this alloy. The different responses of AA6063 and AA6082 suggest that strain-path-induced redistribution of slip favorability is affected by alloy chemistry and the associated restoration behavior.

4.4.3.6 Conclusions on Pseudo-hot Rolling

The pseudo-hot rolling section investigated the microstructure and texture evolution of AA6xxx series alloys under pseudo-hot rolling conditions. The comparison considered unidirectional pseudo-hot rolling (UHR) and cross pseudo-hot rolling (CHR), as well as different initial states, including as-cast and as-extruded conditions. The results show that the hot-rolling response is controlled by the balance between strain accumulation and restoration processes. This balance changes with alloy chemistry, initial microstructure, and strain path.

A key result is the different substructure stability of AA6082 and AA6063. In AA6082, the higher solute content and greater particle-pinning potential contribute to stronger retention of deformation substructure. This allows the alloy to preserve a higher degree of stored deformation structure up to large reductions. At 90% reduction, the microstructure shows pronounced subdivision and localized grain-boundary protrusions consistent with grain-boundary bulging and boundary migration. In contrast, AA6063 shows a stronger recovery-controlled response. Its lower solute content and sparse second-phase particle population are consistent with higher boundary mobility and weaker stabilization of deformation substructure. This explains the non-monotonic behavior observed at intermediate reductions, where the LAGB density decreases and locally coarser area-carrying grains are observed. The short heat-treatment experiment further supports this interpretation, because reheating produces a rapid reduction in substructure and a cleaner, more equiaxed morphology.

The initial material state also affects the hot-rolling response. In as-extruded AA6082, the elongated grain morphology and inherited texture remain visible up to intermediate reductions. This indicates that the extrusion-derived structure is not fully removed during the early stages of rolling. At 90% reduction, this inherited structure becomes less dominant, especially under CHR. Compared with UHR, CHR interrupts the directional continuity of deformation and reduces the persistence of the elongated morphology. This produces a more fragmented and more homogeneous microstructure in the as-extruded condition.

Texture analysis shows that rolling-type deformation components, mainly S and Bs/Brass-related orientations, dominate many of the as-cast conditions. In AA6082, stronger texture

development at intermediate reduction is consistent with better retention of deformation substructure. In the as-extruded condition, the most important texture change is the strong decrease in Brass and the simultaneous increase in Cube under UHR at 90% reduction. This change should not be interpreted as a direct Brass-to-Cube transformation. Rather, it shows that the inherited extrusion texture is strongly modified during high-strain pseudo-hot rolling, allowing Cube-oriented volumes to become more prominent. Under CHR, the Cube fraction remains lower and the texture is more distributed, indicating that pass-to-pass rotation limits the persistence of a single dominant orientation component. In AA6063, the texture remains weaker and less monotonic, which is consistent with the stronger effect of recovery during pseudo-hot rolling.

The orientation-based slip metrics add a directional interpretation to the texture results. They show how the measured orientations interact with RD loading through Taylor factor and Schmid-factor distributions. The results indicate that changes in texture components do not necessarily produce proportional changes in slip-favorability classes. This is especially clear in the as-extruded AA6082 condition, where large changes in Bs and Cube fractions occur while the soft, medium, and hard Schmid-factor fractions remain relatively stable. Therefore, the slip metrics provide information that is not fully captured by component-based texture analysis alone.

CHAPTER 5

MICROSTRUCTURE AND TEXTURE EVOLUTION DURING HEAT TREATMENT OF ROLLED SAMPLES

5.1 Introduction

The microstructural state developed during rolling is inherently unstable due to the high density of stored deformation energy. Heat treatment following rolling restores the microstructure through recovery and recrystallization processes, which directly influence grain structure, boundary characteristics, and crystallographic texture. Understanding these transformations is essential for controlling the final properties of AA6xxx alloys, particularly AA6082, where both the initial condition (as-cast or as-extruded) and the imposed deformation path significantly affect the subsequent annealing response.

This chapter investigates the evolution of microstructure and texture in rolled AA6082 during post-deformation heat treatment under different thermal conditions. The objective is to establish a clear link between the initial deformation state, the annealing parameters, and the resulting restoration mechanisms. Special attention is given to the competition between recovery and recrystallization, as well as to the stability of deformation-induced texture components during thermal exposure.

Section 5.2 presents a detailed study of heat treatment after 60% cold rolling of as-extruded AA6082, conducted at intermediate temperatures of 350 °C and 400 °C using a muffle furnace. In addition, results for rolled as-cast AA6082 at 400 °C are included to highlight the influence of the initial microstructure. The annealing durations range from 5 min to 1600 min, allowing the investigation of both early-stage recovery and long-term recrystallization behavior. This section provides a comprehensive description of the slower, diffusion-controlled restoration processes occurring at moderate temperatures.

Section 0 focuses on heat treatment at a higher temperature (520 °C), applied to rolled as-extruded AA6082 at 60% and 90% reduction using a hot salt bath. In contrast to the previous section, the annealing durations are significantly shorter (1–10 s), enabling the study of rapid recrystallization kinetics under high thermal activation. This section examines the evolution of microstructure, grain size distribution, and texture, as well as the stability of the non-recrystallized (NR) subset during near-complete recrystallization.

Finally, a comparative analysis is performed to evaluate the effect of three key parameters on the heat-treatment response: temperature, annealing time, and prior deformation level. By integrating the results from both intermediate and high-temperature conditions, this chapter aims to provide a comprehensive understanding of how these parameters interact to control recrystallization kinetics, microstructural refinement, and texture evolution in AA6082.

5.2 Article - Annealing of Rolled As-extruded AA6082 at 350-400 °C

Texture Evolution and Recrystallization Behavior of Cold Rolled AA6082 Samples with Different Initial Textures Under Various Annealing Times and Temperatures

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5.2.1 Abstract

This study aims to investigate the recrystallization behavior of cold-rolled aluminum alloy AA6082 at annealing temperatures of 350 °C and 400 °C and the impact of initial microstructures. It focuses on microstructural evolutions during annealing and their

relationship to the spatial distribution of misorientations in the deformed state. A cast material with random texture is compared to an as-extruded material with a strong fiber texture rolled perpendicular to its extrusion direction. Electron backscatter diffraction analysis (EBSD) is used to describe the deformed microstructure and the formation and stability of recrystallization components. The impact of particles and orientation pinning on grain boundary mobility was documented. Thermodynamic calculations were performed using FactSage to predict phase stability and precipitation behavior during annealing. These results were used to identify the formation and dissolution of secondary phases that could influence recrystallization. The study further explores the heterogeneity of the recrystallization process and how pinning effects contribute to texture evolution by preserving some of the deformation-induced orientations even after long annealing times and by limiting the development of recrystallization components. Particular attention is given to identifying which texture components resist prolonged annealing and how these resultant components affect microstructural and texture evolution. These findings contribute to a better understanding of recrystallization mechanisms and provide valuable insight into optimizing processing parameters. Among the aluminum alloys used in automotive structures.

5.2.2 Introduction

Vehicle mass reduction is an important strategy for improving fuel efficiency and lowering emissions, which has increased the use of aluminum alloys in automotive structures (Joost, 2012). Processed aluminum profiles are frequently used in bumper and crash-box systems, where they absorb impact energy through controlled plastic deformation (G. Wang, Zhang, Zheng, Chen, & Yu, 2022). Their mechanical response may vary with the loading direction, a behavior known as mechanical anisotropy. This anisotropy can affect plastic flow and ductility, making its control important for reliable component performance (Fourmeau, Børvik, Benallal, Lademo, & Hopperstad, 2011). A major contributor to this directional response is crystallographic texture, defined as the preferred orientations of grains developed during thermomechanical processing (Engler & Hirsch, 2002). Therefore, understanding how

processing influences microstructure, and texture is essential for the design of lightweight automotive structures.

Among the aluminum alloys used in automotive structures, the AA6xxx series comprises heat-treatable wrought alloys based primarily on the Al–Mg–Si system. (Sheppard, 1999). In these alloys, magnesium and silicon form Mg_2Si -based precipitates during heat treatment, providing the principal precipitation-hardening mechanism (Sheikh et al., 2026). Within this series, AA6082 is an Al–Mg–Si–Mn alloy with relatively high Mg and Si contents and a significant Mn addition. This composition provides strong precipitation-hardening potential, while Mn-containing dispersoids can restrict grain-boundary motion and enhance microstructural stability during thermomechanical processing (Humphreys & Hatherly, 2004). Consequently, AA6082 is widely used for load-bearing automotive components, including bumpers, crash boxes, and structural beams (Sheppard, 1999). However, the final mechanical response is not controlled by precipitation alone; it also depends on microstructural evolution, recrystallization behavior, and the persistence of deformation-induced orientations during thermomechanical processing (Lypchanskyi et al., 2024). Therefore, the influence of initial texture condition, deformation structure on recrystallization behavior, and texture stability after cold-rolled AA6082 during annealing requires further clarification.

The microstructural evolution of AA6082 during cold rolling is sensitive to the initial material condition, resulting in different deformation morphologies. The as-cast material contains relatively coarse grains, whereas the as-extruded material retains elongated grains, a deformation substructure, and a distinct texture (Nienaber, Esterl, Ben Khalifa, & Bohlen, 2026). These differences in grain morphology, substructure, and texture can influence how plastic strain is accommodated during subsequent cold deformation (Poletti, Wójcik, & Sommitsch, 2013). With increasing reduction, both conditions undergo progressive grain subdivision through the formation of dislocation cells, dense boundary networks, and lamellar structures (Y. C. Huang, Liu, Li, Liu, & Yang, 2016). However, the resulting distributions of stored energy and boundary misorientation remain dependent on the initial microstructure, thereby influencing subsequent recovery and recrystallization (Buzolin, Krumphals, & Poletti, 2023).

Annealing, which consists at heating the material to a specific temperature and for a given time, transforms the cold-deformed microstructure through recovery and recrystallization (Imran & Walther, 2016). As annealing proceeds, the dislocation density and stored deformation energy decrease, while the elongated deformation structure is progressively replaced by recrystallized grains, usually equiaxed and smaller (Miszczyk et al., 2017). Temperature and holding time determine the rate and extent of recrystallization, as well as the resulting grain morphology and crystallographic texture (P. Du et al., 2024). The developed texture then influences the plastic flow, and annealing parameters must be selected carefully to achieve an appropriate balance between strength, formability, and anisotropic response in the final product and mechanical anisotropy .

Crystallographic texture is commonly characterized using pole figures and orientation distribution functions (S. Papadopoulou, Tsiolis, Gavalas, & Papaefthymiou, 2022). Texture intensity is expressed in multiples of a random distribution, where 1 m.r.d. corresponds to the intensity of a random distribution and high values indicate stronger orientation concentrations (Amer et al., 2025). The volume fraction of a specific texture component can be calculated by integrating the ODF intensity within a defined angular tolerance around its ideal orientation (W. C. Liu & Morris, 2002b). These quantitative parameters allow texture evolution to be compared and related to plastic anisotropy in aluminum sheets (Sidor, 2021).

The texture components in rolled aluminum alloys are described using specific components: Brass (Bs), S, Copper (Cu), Cube, Rotated Cube (r-Cube), $\{111\}\langle uvw \rangle$, and Goss (Gs). Among these, the Bs, S, and Cu textures are typical deformation orientations. These textures store high dislocation density during deformation (Li, Zhao, Liu, & Li, 2018b) and their stabilities depend on factors such as alloy composition, initial microstructure, deformation path, amount of deformation, and annealing conditions (W. C. Liu, Li, Yuan, & Yang, 2007). Recrystallization components in rolled aluminum alloys are generally described using four types of recrystallization textures: (1) Cube $\{001\}\langle 100 \rangle$ and R $\{124\}\langle 211 \rangle$ components, which are often associated with nucleation in relatively low-stored-energy deformation bands; (2) Gs $\{011\}\langle 100 \rangle$, Q $\{013\}\langle 231 \rangle$, and P $\{011\}\langle 122 \rangle$ orientations, associated with grains from deformation bands or shear regions; (3) Rolling texture rotated by $40^\circ\langle 111 \rangle$ produced

by oriented growth; and (4) Retained rolling textures orientation, indicating insufficient driving force or pinning effects that hinder grain boundary migration (Engler & Lücke, 1992).

Thermomechanical processing and severe plastic deformation introduce a high density of crystallographic discontinuities into aluminum alloys: mainly dislocations, subgrain boundaries, and local deformation heterogeneities (Dhal, Panigrahi, & Shunmugam, 2017). The accumulation of these defects results in stored deformation energy and makes the deformed microstructure thermodynamically unstable (Yan, Mao, Lin, Zhang, & Wang, 2017). During subsequent annealing, this stored energy is reduced through restoration processes, mainly static recovery and static recrystallization (Ding, Zhang, Khan, & Yanagimoto, 2022). Static recovery occurs first and is controlled by dislocation glide, cross-slip, and climb, allowing dislocations to annihilate and rearrange into more stable configurations (Dhal et al., 2017). This process leads to polygonization and the formation of organized subgrain structures separated by low-angle grain boundaries (LAGBs), without the formation of new grains (Hansen & Juul Jensen, 2011; Humphreys & Hatherly, 2004).

When sufficient stored energy remains after recovery, static recrystallization can occur through the nucleation and growth of new strain-free grains, accompanied by the migration of high-angle grain boundaries (HAGBs) into the deformed matrix (Ding et al., 2022). In some cases, recrystallization may also develop gradually through increasing misorientation across subgrain boundaries, where LAGBs progressively transform into HAGBs (Maric et al., 2026). The onset and progress of recrystallization are therefore influenced by local variations in stored energy, defect structure, and orientation gradients introduced during deformation. Regions with higher defect density or stronger misorientation gradients can provide greater local driving forces for nucleation and boundary migration, allowing recrystallized grains to form or grow preferentially during annealing (Maric et al., 2026).

Experimental observations in Al–Mg–Si alloys illustrate how deformation texture and substructure evolution influence subsequent recrystallization. Jin and Lloyd (Jin & Lloyd, 2005) found that Cube and Goss orientations in AA6111 were metastable during cold rolling and formed distinct bands along the rolling direction. During subsequent solution heat treatment above 500 °C for a few minutes, these bands acted as preferred recrystallization

nucleation sites, and their spatial arrangement remained evident in the final solution-treated sheet. Yamamoto et al. (Yamamoto et al., 2011) monitored microstructural changes during annealing at 400 °C in Al–Mg–Si alloys cold rolled by 30% and 50% and showed that progressive subgrain coarsening promoted the formation and migration of high-angle grain boundaries, eventually leading to recrystallization. For AA6082, Liu et al. (S. Liu, Xu, Jin, Li, & Peng, 2022b) investigated a homogenized ingot that was hot rolled, subjected to intermediate annealing at 400, 475, or 550 °C for 1 h, and then cold rolled from 5 to 1.5 mm before a final T6 treatment. They showed that increasing the intermediate annealing temperature promoted coarsening of Mg₂Si particles and weakened the subsequent rolling texture, particularly the β -fiber components. After the final solution treatment and aging, higher intermediate annealing temperatures promoted particle-stimulated nucleation, resulting in finer and more equiaxed recrystallized grains together with higher fractions of Cub_{END} and random orientations.

The texture change during recrystallization in aluminum alloys is typically described using two approaches: oriented nucleation and oriented growth (Hjelen, Ørsund, & Nes, 1991). Oriented nucleation favors specific grain orientations in the deformed microstructure that tend to nucleate first and therefore retain an early advantage during recrystallization. Oriented growth is based on the observation that some grain boundary misorientations grow the fastest (Hjelen et al., 1991; Ma et al., 2014; Mishin, Juul Jensen, & Hansen, 2010). These two mechanisms, combined with initial deformation process and annealing conditions, control the recrystallization textures observed in aluminum alloys.

Particle-stimulated nucleation (PSN) is another important mechanism affecting recrystallization texture. PSN occurs around coarse second-phase particles, typically larger than 1 μ m, that remain non-deformable during plastic deformation (Engler et al., 1996). During rolling, these particles create highly strained particle-deformation zones in the surrounding matrix, where stored energy and local orientation gradients are high (K. Adam & Field, 2023; Sidor et al., 2015). During subsequent annealing, subgrains within these zones can grow rapidly and act as preferred nucleation sites for recrystallized grains (Engler et al., 1996). Because the orientations in particle-deformation zones are usually complex and less ordered,

PSN can weaken strong recrystallization textures, such as Cube, and promote a more random overall texture in the final sheet (Kumar et al., 2020).

These deformation and recrystallization mechanisms can be characterized using EBSD, which enables microstructural and crystallographic texture analyses at the grain and subgrain scales. Because recovery and recrystallization modify internal orientation gradients and boundary structures during annealing, EBSD misorientation analysis provides a means of tracking these changes (Humphreys & Hatherly, 2004). Such approaches have been developed to distinguish recrystallized grains from deformed or recovered regions in partially recrystallized microstructures (A. Ayad et al., 2021). Grain Orientation Spread (GOS) and Grain Average Misorientation (GAM) provide grain-scale information, whereas Kernel Average Misorientation (KAM) and Geometrically Necessary Dislocation (GND) density describe more localized variations in orientation and deformation (Abdelhak Ayad et al., 2012). Traka (Traka, 2022) used KAM to evaluate local orientation gradients in the deformed microstructure and relate them to subsequent recrystallization behavior. Unlike KAM and GND density, which are calculated from misorientations between neighboring pixels, GOS represents the average misorientation of all measurement points within a reconstructed grain relative to its mean orientation. Ayad et al. (Abdelhak Ayad et al., 2012) showed that GOS is effective for assessing intragranular orientation gradients and distinguishing deformed grains from recrystallized grains. In the grain-reconstruction procedure reported by Ayad et al. (A. Ayad et al., 2021), a tolerance angle of 5° was used, such that neighboring pixels with misorientations below this value were assigned to the same reconstructed grain. GOS is therefore comparatively less sensitive to EBSD step size than local misorientation parameters (A. Ayad et al., 2021).

Recrystallized grains are commonly identified using a low GOS threshold. Values between approximately 1° and 3° based on the different materials have frequently been applied (A. Ayad et al., 2021). Ayad et al. (Abdelhak Ayad et al., 2012) reported that GOS increases during plastic deformation and decreases during recrystallization, demonstrating its sensitivity to changes in intragranular orientation spread. However, GOS alone can lead to classification errors. Small subgrains may exhibit low GOS values and therefore be incorrectly identified as

recrystallized during the deformed or early recrystallization stages. Conversely, similarly oriented recrystallized grains may impinge and be reconstructed as a single grain when the misorientation between them is below the selected grain-detection tolerance (A. Ayad et al., 2021). To reduce these limitations, Ayad et al. (A. Ayad et al., 2021) combined a minimum grain-size criterion with GOS and GOS normalized by grain diameter, GOS/D. This combination improved the separation of recrystallized and non-recrystallized regions. Wagner et al. (Wagner, Allain-Bonasso, Berbenni, & Field, 2012) also reported that GOS and GOS/D increase with deformation and decrease with increasing annealing time and temperature, indicating that both parameters can be used to assess microstructural evolution.

To comprehend the recrystallization behavior and texture evolution of AA6082 aluminum alloy, this study focuses on the influence of the initial material state and annealing temperature on the microstructure and crystallographic texture evolution. A randomly oriented cast microstructure and a highly textured extruded sample deformed perpendicularly to its extrusion direction were compared. The latter condition was selected to document how the orientations stabilized during the extrusion process may get destabilized by severe rolling normal to the extrusion direction. The competition between recovery and recrystallization mechanisms at 350 and 400 °C gave an opportunity to evaluate the characteristics of the regions that resist recrystallisation in such processing conditions.

5.2.3 Experimental

5.2.3.1 Experimental Procedure

The Al-Mg-Si-Mn alloy used in this study is an AA6082, with standard composition limits (wt%) listed in Table 5-1. The as-received materials were obtained in two initial conditions: as-cast and as-extruded. The as-cast material was a 100 mm diameter bar produced by horizontal direct chill casting (DC) and homogenized at 560 °C for 2 hours. The as-extruded material was a 58 mm diameter commercial extruded bar. Rectangular samples with the dimension of 70 × 40 × 5 mm were cut from the central region of both bars for further processing.

Table 5-1 Chemical compositions limits per EN 573-3 for AA6082 aluminum alloy (wt%)

Alloy	Mg	Si	Cu	Mn	Cr	Fe	Al
AA6082	0.6-1.2	0.7-1.3	max 0.1	0.4-1.0	max 0.25	max 0.5	Balance

Cold rolling was conducted using a unidirectional rolling process at room temperature for these samples. Rolling was performed at the Centre de métallurgie du Québec (CMQ) in Trois-Rivières using a Stanat two-high laboratory rolling mill equipped with rolls of 152.4 mm (6 inches) in diameter. The rolling speed was 40 mm/s, and the sample thickness was reduced from 5 mm to 2 mm, corresponding to a total rolling reduction of 60% and an effective plastic strain of 1.06. As the extruded sample was rolled perpendicular to the extrusion direction, the extrusion direction (ED) corresponds to the transverse direction (TD) of the rolling process.

5.2.3.2 Annealing Treatment

Multiple specimens were cut from rolled samples. The specimens were heated in a muffle furnace at either 350 °C or 400 °C, with soaking times ranging from 5 to 1600 min to document the onset of recrystallisation as well as the stable condition. Microstructural characterization was performed at the center of each specimen to minimum the potential contribution of shear deformation.

5.2.3.3 Microstructural and Texture Analysis

The texture measurements were conducted using EBSD with a Hitachi SU-70 field emission gun scanning electron microscope (SEM). The grain structure and texture of partially and fully recrystallized specimens were examined on the TD–ND plane in order to have more statistics for a texture evaluation. The band contrast (BC) of the pseudo-Kikuchi diffraction patterns can be used to quantify the local dislocation density. Bright areas indicate low dislocation density regions and dark regions indicate regions with high dislocation density. Subgrains were

identified by a misorientation of 2° – 15° , marked in red, and grains by misorientations $>15^{\circ}$, marked in black. A step size of $3\text{ }\mu\text{m}$ was used at $100\times$ magnification to get a global view of the microstructure, and $0.7\text{ }\mu\text{m}$ for higher magnifications.

ATEX version 4.14 (Beausir & Fundenberger, 2017) was used to plot maps, obtain the texture, and identify orientation of recrystallized grains. The crystallographic texture was visualized using $\{111\}$ pole figures and the main texture components were calculated with a 15° deviation from the ideal orientation. The spatial distribution of specific texture component maps was documented, with distinct colors assigned to each component. Table 5-2 lists the Miller indices, $\{hkl\}\langle uvw\rangle$, and Euler angles (φ_1 , θ , φ_2) of the most frequently observed orientations in rolled and recrystallized AA6XXX specimens. In this study, the summation of the volume fractions of these listed variants was reported for each texture component and the symmetry of the rolling process was not applied.

Table 5-2 Euler angles and Miller indices of common texture components in FCC metals and alloys

Texture Component	Euler Angles (φ_1 , φ , φ_2)	Miller Indices
Cube (Cb)	(0, 0, 0)	$\{001\}\langle 100\rangle$
Goss (Gs)	(0, 45, 0)	$\{011\}\langle 100\rangle$
Rotated-Goss (RG)	(90, 45, 0)	$\{011\}\langle 011\rangle$
Brass (Bs) I	(35, 45, 0)	$\{011\}\langle 2\bar{1}1\rangle$
Brass II	(145, 45, 0)	$\{011\}\langle 2\bar{1}1\rangle$
Copper (Cu) I	(90, 35, 45)	$\{112\}\langle \bar{1}\bar{1}1\rangle$
Copper II	(270, 35, 45)	$\{112\}\langle 11\bar{1}\rangle$
Rotated-Copper (RCu) I	(0, 35, 45)	$\{112\}\langle 1\bar{1}0\rangle$
Rotated-Copper II	(180, 35, 45)	$\{112\}\langle \bar{1}10\rangle$
S I	(61, 34, 64)	$\{2\bar{1}\bar{1}\}\langle 1\bar{2}4\rangle$
S II	(241, 34, 64)	$\{21\bar{1}\}\langle 1\bar{2}4\rangle$
S III	(299, 34, 26)	$\{11\bar{2}\}\langle 4\bar{2}1\rangle$
S IV	(119, 34, 26)	$\{\bar{2}11\}\langle 1\bar{2}4\rangle$

5.2.3.4 Grain Size Distribution

Grain sizes were extracted from EBSD maps as the equivalent-circle diameters (ECD) of the grain. The relationship used throughout is given in (5-1), which converts a grain's mapped area A to the equivalent diameter D of a circle having the same surface:

$$A = \pi \left(\frac{D}{2} \right)^2 \quad (5-1)$$

Grain morphology was quantified using the aspect ratio, defined as the minor-axis length divided by the major-axis length:

$$AR = \frac{\text{minor-axis length}}{\text{major-axis length}} \quad (5-2)$$

According to this definition, represents an equiaxed feature, whereas values approaching zero indicate increasingly elongated morphology. The same aspect-ratio definition was subsequently applied to both grains and second-phase particles.

Distributions are reported as normalized probability density functions (PDFs) together with the cumulative distribution function (CDF) on an area-fraction basis. Following Toth et al. (Toth et al., 2013), PDFs were used instead of raw histograms so that curves are comparable regardless of bin width; all datasets employed 42 equal-width bins (area under each PDF = 1). To avoid artifacts from undersized features, grains smaller than 3 pixels were excluded, corresponding to a minimum grain area of $27 \mu\text{m}^2$ and an equivalent-circle diameter threshold of $D_{\min} \approx 5.8 \mu\text{m}$.

5.2.3.5 Recrystallization Analysis

To investigate the onset and progression of recrystallization, grains were reconstructed using a grain-detection tolerance angle of 5° to better capture the evolution of the microstructure. Neighboring pixels with misorientations below this value were assigned to the same

reconstructed grain. Grains smaller than three pixels were excluded from the analysis. Recrystallized grains were identified using a combined GOS and GOS/D criterion. Following the approach proposed in (A. Ayad et al., 2021), grains with $GOS < 2.5^\circ$ and $GOS(c)/D < 0.1^\circ/\mu m$ were classified as recrystallized in the present study. Grains that did not satisfy both criteria were classified as non-recrystallized. The thresholds were calibrated so that less than 1% of the grains in the fully deformed state were identified as recrystallized (A. Ayad et al., 2021). This calibration provided a reliable distinction between recrystallized regions and grains containing deformation substructures.

5.2.3.6 Retention and resistance of texture components during annealing

To describe the microstructure evolution of specimens after near-complete recrystallization, two complementary measures were derived from the EBSD-based texture fractions and the distortion metrics defined in Section 5.2.3.5. The retention index (RI) quantifies texture persistence by comparing the weight fraction of each component in the non-recrystallized (NR) fraction with the rolled condition.

At each annealing condition, the NR fraction was calculated as:

$$f_{NR} = 1 - f_{REX} \quad (5-3)$$

where f_{REX} and f_{NR} are the recrystallized and non-recrystallized fractions, respectively.

For each texture component c , the measured fraction within the NR regions V_c^{NR} is weighted by the non-recrystallized fractions to quantify its retention:

$$V_{c,weighted}^{NR} = V_c^{NR} f_{NR} \quad (5-4)$$

The Retention Index (RI) of component c was then calculated as:

$$RI_c = \frac{V_{c,weighted}^{NR}}{V_c^{rolled}} \quad (5-5)$$

where V_c^{rolled} is the initial fraction of texture component c in the 60% cold-rolled reference before annealing. The rolled as-cast condition was used as the reference for the annealed as-cast specimens, while the rolled as-extruded condition was used for the annealed as-extruded specimens. Only components with an initial rolled fraction greater than 1.5% were considered to avoid unreliable RI values caused by very small initial fractions.

The NR substructure was described using a resistance score (RS) that combines the intragranular misorientation (GOS) and misorientation normalized by grain size $((GOS/D)_c)$. These values were normalized for a given annealing condition using the highest (max (x)) and the lowest (min (x)) GOS and $(GOS/D)_c$ values between S, RG, and Bs(I-II) using the following equation:

$$\tilde{x}_c = \frac{x_c - \min(x)}{\max(x) - \min(x)}, \quad x \in \{GOS, GOS/D\} \quad (5-6)$$

The resistance score (RS) of component c was then calculated as:

$$RS_c = \frac{1}{2} \widetilde{GOS}_c + \frac{1}{2} \left(\frac{\widetilde{GOS}}{D} \right)_c \quad (5-7)$$

RS_c lies in [0,1] and ranks components by the stability of their NR state. A high value of the RS indicates a relatively highly misoriented substructure compared to the other NR components, whereas a low RS indicates a lower intragranular misorientation and a more uniform internal orientation state.

5.2.3.7 Secondary Particle Quantification and Phase-Equilibria Modeling

Second-phase particles in the rolled as-extruded and annealed specimens were characterized using BSE-SEM and EDS. The images were processed in MATLAB R2021a to measure particle area fraction, equivalent circular diameter, aspect ratio, elongation ratio, and orientation. Area-weighted values were calculated using the projected particle area as the weighting factor.

Particle aspect ratio was calculated according to the definition introduced in (5-2). The elongation ratio was defined as its reciprocal:

$$ER = \frac{\text{major-axis length}}{\text{minor-axis length}} = \frac{l}{R} \quad (5-8)$$

Thus, an aspect ratio approaching unity corresponds to a more equiaxed particle, whereas a larger elongation ratio indicates a more elongated morphology.

Particle orientation was measured from the fitted major axis relative to the horizontal image axis, corresponding to TD and the prior extrusion direction. Only particles with $AR \leq 0.80$ were included in the orientation analysis because the orientation of nearly round particles is not reliably defined. This filter was applied only to the orientation analysis.

The accepted orientations were grouped into 10° intervals. The particle number fraction in each interval was calculated relative to the total number of analyzed particles as

$$f_j = \frac{n_j}{N_{\text{orientation}}} \times 100 \quad (5-9)$$

where n_j is the number of accepted particles in interval j , and N_{total} is the total number of analyzed particles. The radial axis of the polar plots therefore represents the percentage of all analyzed particles within each angular interval.

The concentration of the accepted particle orientations was quantified using the area-weighted axial mean resultant length, R . Following the treatment of axial data by Mardia and Jupp (2000) (Mardia & Jupp, 1999), the measured orientations were doubled because θ and $\theta + 180^\circ$ represent the same particle axis. The normalized projected particle area was incorporated as the weighting factor:

$$R = \sqrt{\left[\sum_{i=1}^{N_{orientation}} w_i \cos(2\theta_i) \right]^2 + \left[\sum_{i=1}^{N_{orientation}} w_i \sin(2\theta_i) \right]^2} \quad (5-10)$$

where $N_{orientation}$ is the number of particles accepted for orientation analysis, θ_i is the signed major-axis orientation of particle i relative to TD, and w_i is the normalized projected-area weight, calculated as:

$$w_i = \frac{A_i}{\sum_{i=1}^{N_{orientation}} A_i} \quad (5-11)$$

where A_i is the projected area of particle i . The mean resultant length, R , ranges from 0 to 1. A value approaching 1 indicates a highly concentrated orientation distribution, meaning that the particle major axes are preferentially aligned along a common direction. In contrast, a value approaching 0 indicates a highly dispersed orientation distribution with no dominant axial direction.

In addition, the chemical characteristics of the different particle populations were examined using EDS elemental maps. In parallel, the temperature-dependent equilibrium phase distribution in AA6082 and the corresponding theoretical precipitate volume fractions were predicted via thermodynamic calculations in FactSage 8.4, using the FTLite and FactPS databases for aluminum alloy systems.

5.2.4 Results

5.2.4.1 As-cast Specimens: Rolling and Annealing Behavior

The as-cast microstructure (Figure 5-1a, c) shows equiaxed grains with a weak random texture, typical of cast aluminum alloys produced by the DC casting technique. The microstructure after 60% cold rolling (Figure 5-1b) shows elongated grains, illustrating the significant plastic deformation caused by rolling. The EBSD orientation map is dominated by blue grains, corresponding to orientations near $\langle 111 \rangle \parallel \text{RD}$, together with magenta grains close to $\langle 112 \rangle$. Starting from an initially random texture, 60% reduction produces a strong β -fiber rolling texture with a maximum intensity of 4.2 (Figure 5-1d), characterized by measured fractions of the S (25%), Bs (13%), and Cu (10%) components (Figure 5-3b).

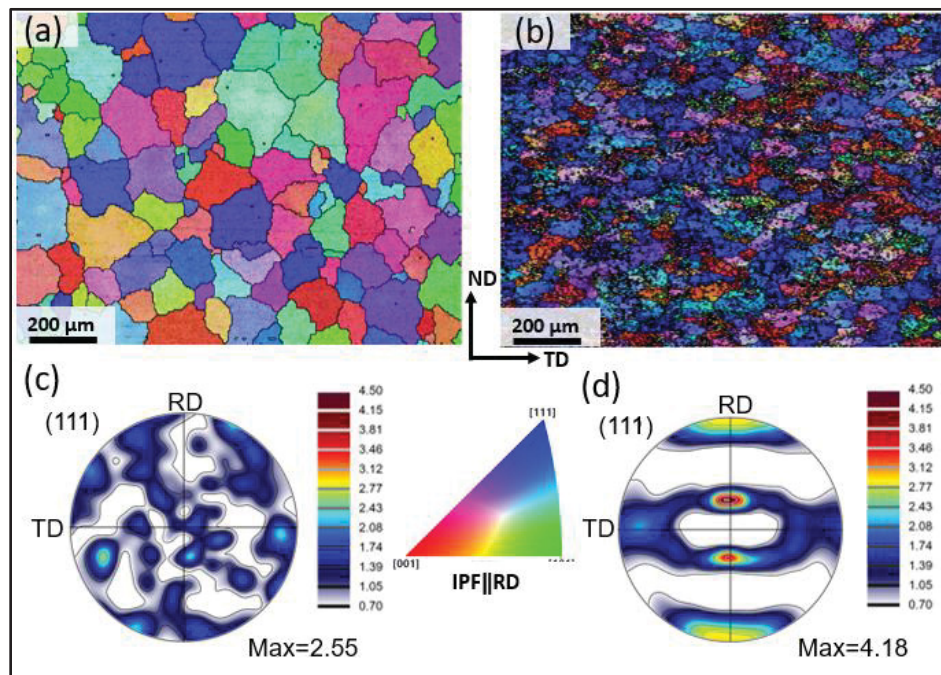


Figure 5-1 Orientation maps and texture characterization of the as-cast specimen: (a) initial as-cast microstructure, (b) microstructure after 60% cold rolling, (c) $\{111\}$ pole figure of the as-cast specimen, and (d) after 60% cold rolling

After annealing at 400 °C for various durations, the rolled as-cast specimens show a transition from a deformed to a recrystallized state. After 5 min of annealing, the band contrast map (Figure 5-2a) shows a microstructure full of dislocations, with only a few isolated regions exhibiting light gray levels. These isolated regions, bounded by HAGBs, suggest the onset of recrystallization. A high density of LAGBs, calculated from the EBSD maps, subdivides the microstructure into narrow regions. The EBSD orientation map (Figure 5-2d) shows that the microstructure still resembles the rolled as-cast condition, characterized by dominant deformation-type orientations. The $\{111\}$ pole figure (Figure 5-2g) displays a peak intensity of ~ 3.9 , which is lower than that of the as-deformed condition, while the texture still remains deformation-type. The texture component analysis (Figure 5-3b) also confirms that deformation components such as S (25%), Bs (12%), and Cu (11%) are still present, along with a Gs component of 5%. These observations are supported quantitatively by Figure 5-3a, where the recrystallized fraction is only $\sim 8\%$, while the HAGB fraction remains limited to 29%.

Significant microstructural changes occur between 5 and 10 min. By 10 min, the band contrast map (Figure 5-2b) shows a clear grain structure with many grains bounded by well-defined HAGBs. Some residual deformation is still present between the clear grain structure. The EBSD map (Figure 5-2e) shows many newly formed grains with clear boundaries. The recrystallization fraction increases sharply to $\sim 85\%$, and the HAGB fraction rises to 68%, resulting in a clear transition between the deformed regions to the recrystallized state (Figure 5-3a). The $\{111\}$ pole figure (Figure 5-2h) shows a maximum intensity of ~ 2.2 , indicating a strong weakening of the deformation texture and a broader distribution of orientations. The texture component analysis (Figure 5-3b) also shows a reduction in deformation texture components, with S, Cu, and Bs decreasing to 7%, 2%, and 8%, respectively. The Gs component also decreases from 5% after 5 min to 0.6% after 10 min, remaining below 1% throughout the rest of the annealing process.

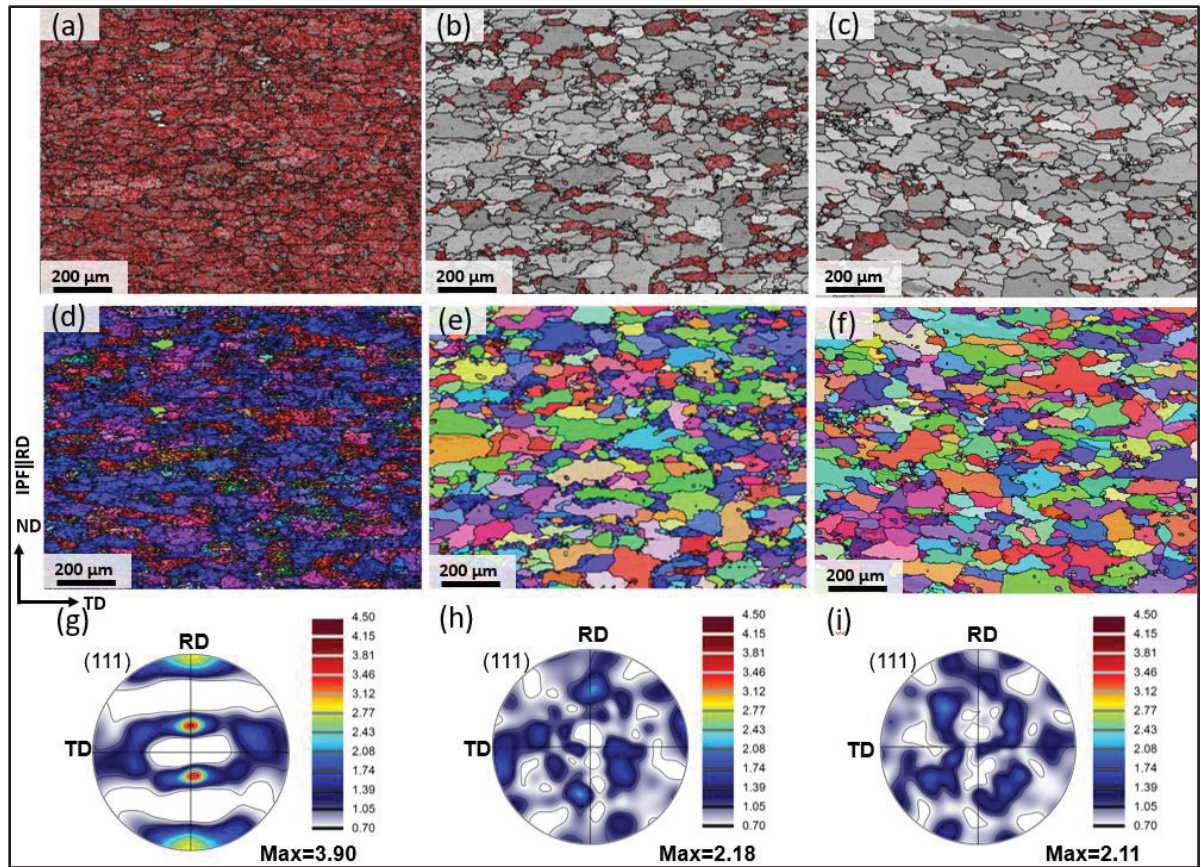


Figure 5-2 60% cold-worked as-cast annealed at 400 °C: band contrast maps (a, b, c), orientation maps (d, e, f), and $\{111\}$ pole figures (g, h, i) for 5 min (a, d, g), 10 min (b, e, h), and 30 min (c, f, i). Misorientation angles are shown in red for 2° – 15° and in black for $>15^\circ$

By 30 min, the band contrast map (Figure 5-2c) shows a more developed recrystallized grain structure than at 10 min, with well-defined HAGBs and only limited remaining LAGB content, indicating a reduction in dislocation substructure. Some internal contrasts are still visible in a few grains. The EBSD map (Figure 5-2f) shows a uniform distribution of newly formed grains, although some local orientation gradients remain within certain grains. The recrystallization fraction rises further to $\sim 94\%$, while the HAGB fraction reaches 76%, confirming that recrystallization is nearly complete at this stage (Figure 5-3a). The $\{111\}$ pole figure (Figure 5-2i) shows further weakening of the texture (Max: ~ 2.1), consistent with the development of a nearly random orientation distribution. The texture component analysis (Figure 5-3b) remains similar to the 10 min condition, although the Cube component becomes more prominent after 30 min, reaching 8%.

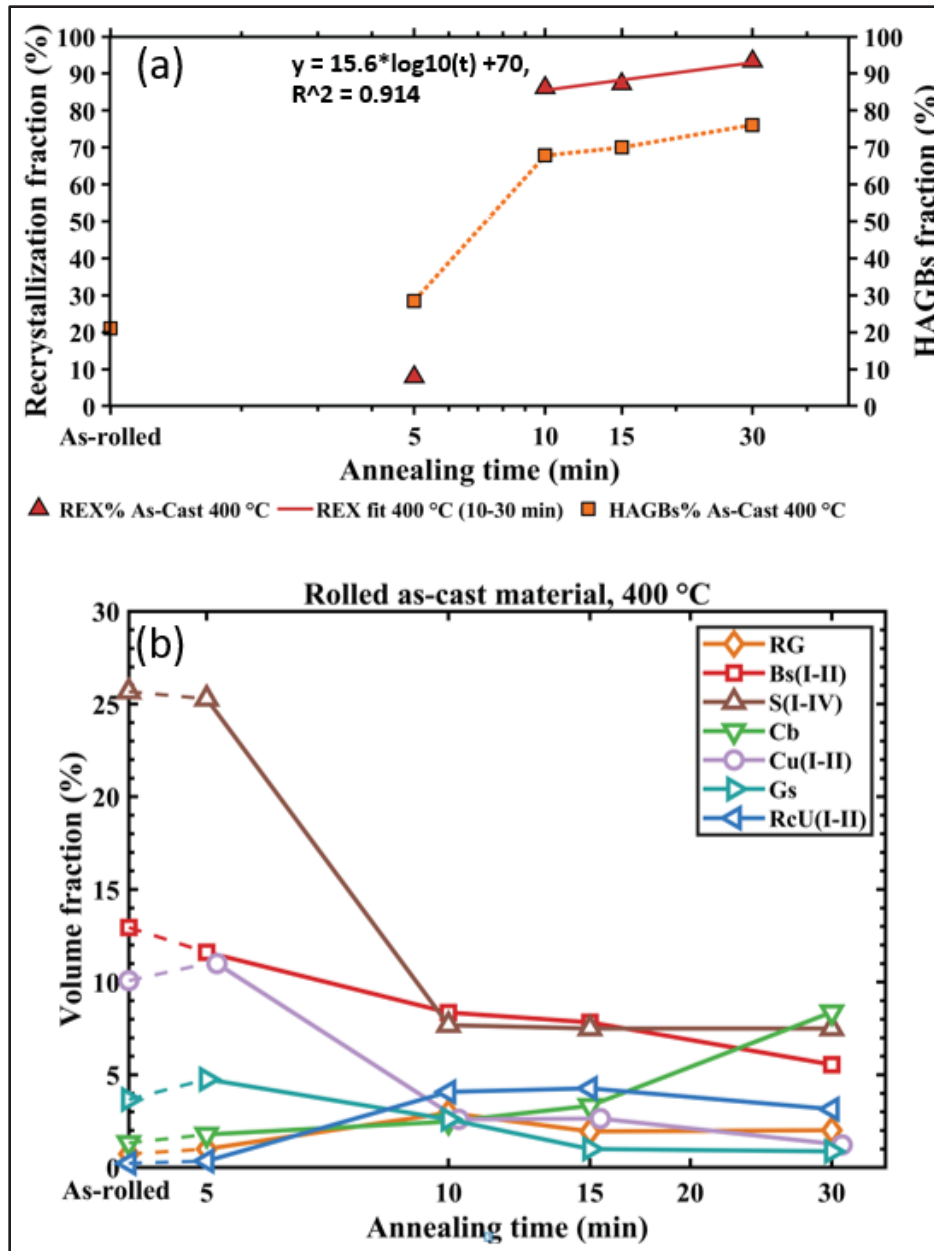


Figure 5-3 (a) Recrystallization and HAGBs fractions plots for rolled as-cast specimens annealed at 400 °C, and (b) volume fractions of texture components as a function of annealing time. Note: the x-axis is plotted on a logarithmic scale

Cumulative grain-size distributions for all conditions are shown in Figure 5-4. In the rolled as-cast state, the distribution is steep and shifted toward smaller grain sizes, with a maximum grain size of approximately 84 μm , indicating a relatively narrow grain-size range. After 5 min at 400 °C, the curve shifts only slightly toward larger grain sizes, with the maximum grain size

increasing to approximately 92 μm . The major change occurs between 5 and 10 min, where the cumulative curve shifts markedly to larger diameters and the upper grain-size tail extends to approximately 153 μm . Only limited changes occurred in the upper tail of the distribution from 10 to 15 and then 30 min: the curves follow a similar trend which stabilizes near 160 – 170 μm . This indicates that most grain growth occurs between 5 and 10 min, while the grain-size distribution becomes comparatively stable at longer annealing times.

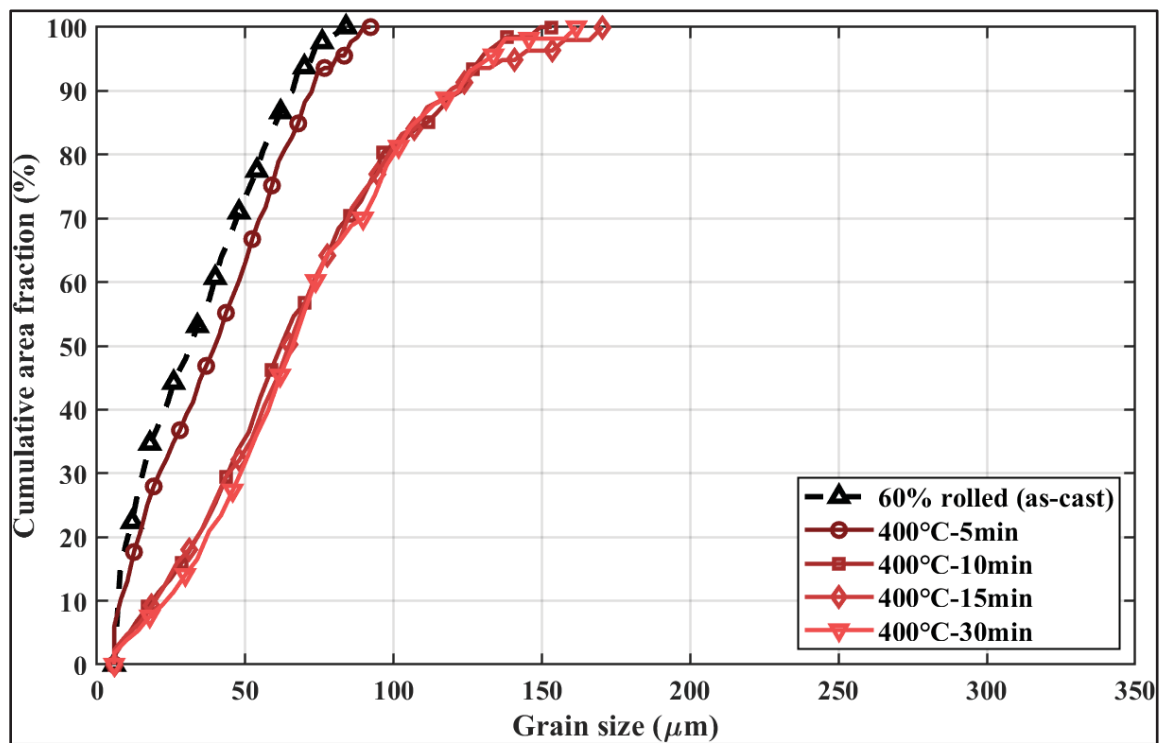


Figure 5-4 Cumulative grain-size distributions for 60% rolled as-cast and specimens annealed at 400 °C for 5–30 min. The 10, 15, and 30 min curves show similar trends, indicating limited additional grain growth at longer annealing times

5.2.4.2 As-Extruded Specimens: Rolling and Annealing Behavior

The band-contrast map of the as-received extruded specimen (Figure 5-5a) shows relatively dark contrast levels, indicating that some stored energy and dislocation networks remain from the hot-extrusion process. The map also reveals a heterogeneous substructure with a high density of LAGBs and dispersed HAGBs. Long grains, aligned with the extrusion direction,

are subdivided into narrow bands by LAGBs. Using the IPF map referenced to the RD, which is here perpendicular to the extrusion direction, the microstructure contains dominant $\langle 001 \rangle$ (red) and $\langle 101 \rangle$ (green) fiber alignments, together with contributions from $\langle 111 \rangle$ (blue) and $\langle 112 \rangle / \langle 113 \rangle$ (purple/magenta) orientations (Figure 5-6a). The orientation is relatively uniform within individual bands, while distinct color changes occur across boundaries.

After 60% rolling perpendicular to the extrusion direction, the microstructure retains elongated grains aligned with the extrusion direction (Figure 5-5b) even if the microstructure is now deformed perpendicularly. The band-contrast map shows darker gray levels associated with a subdivision of the grains. The EBSD map (Figure 5-6b) displays deformation-induced substructures and a reduction in grain thickness compared with the as-extruded condition. A color shift toward purple and violet is also seen in the EBSD map, indicating changes in orientation distribution.

Annealing at 400 °C results in a faster recrystallization response than in the random cast structure. After 5 min, the band-contrast map (Figure 5-5c) still shows a dense LAGB network and dark contrast in many regions, but some lighter areas and HAGB-bounded regions are visible, indicating the beginning of recrystallization. The EBSD map (Figure 5-6c) still contains elongated deformation features, although some local regions with clearer boundaries appear. At this stage, the recrystallized fraction reaches ~30%, and the HAGB fraction is ~47% (Figure 5-6, Figure 5-7). After 60 min, the band-contrast map (Figure 5-5d) shows that the grain structure has not changed markedly; elongated regions and residual internal contrast remain. However, the EBSD map (Figure 5-6d) shows more newly formed grains within the retained deformed structure. The recrystallized fraction increases only slightly to ~40%, while the HAGB fraction remains close to ~49% (Figure 5-6, Figure 5-7). From 5 to 800 min, the recrystallized fraction shows a good linear relationship with the logarithm of annealing time ($R^2 = 0.91$), indicating gradual recrystallization before the sharp increase observed at 1600 min. By 1600 min, the band-contrast map (Figure 5-5e) shows a much more recrystallized structure, with a high fraction of HAGBs and strongly reduced LAGB content. The EBSD map (Figure 5-6e) shows that the elongated deformation morphology is largely replaced by finer

recrystallized grains. At this stage, recrystallization increases sharply to $\sim 87\%$, and the HAGB fraction rises to $\sim 83\%$ (Figure 5-6, Figure 5-7).

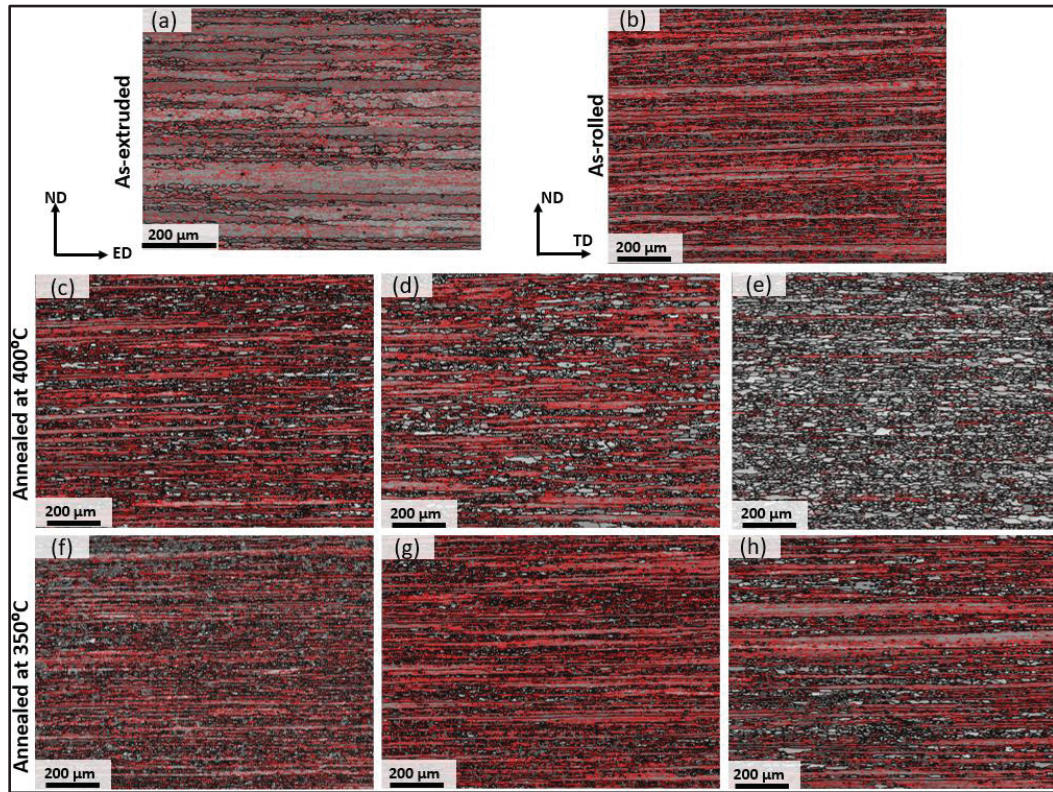


Figure 5-5 Band contrast maps of (a) as-extruded specimen; (b) rolled as-extruded specimen; annealed at 400 °C for (c) 5 min, (d) 60 min, (e) 1600 min; and annealed at 350 °C for (f) 30 min, (g) 60 min, (h) 1600 min. Grain boundaries: (red) 2° – 15° , (black) $>15^{\circ}$

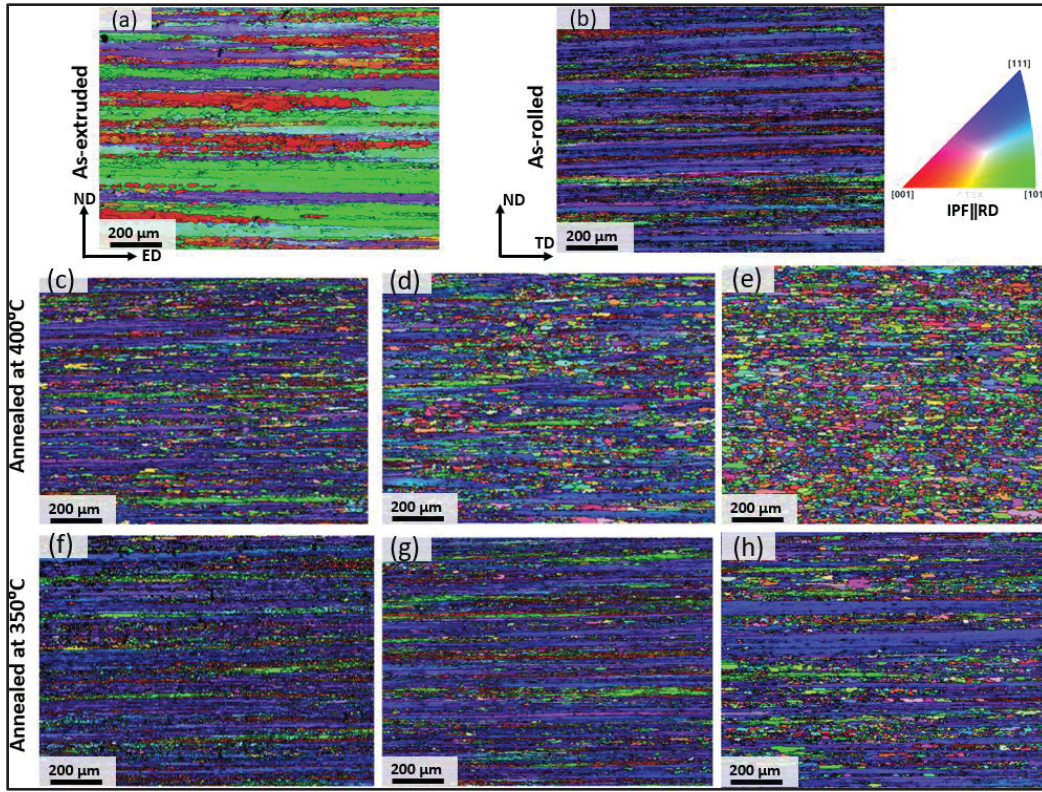


Figure 5-6 Orientation maps of (a) as-extruded specimen, (b) rolled as-extruded specimen; annealed at 400 °C for (c) 5 min, (d) 60 min, (e) 1600 min; and annealed at 350 °C for (f) 30 min, (g) 60 min, (h) 1600 min

Compared with the 400 °C condition, annealing at 350 °C produces a much weaker recrystallization response, and the rolled as-extruded specimen retains much more of its deformation substructure. After 30 min, the band-contrast map (Figure 5-5f) displays dark contrast and a dense LAGB network, indicating that the deformation substructure is still dominant. The EBSD map (Figure 5-6f) also shows elongated grains and retained deformation-type features. At this stage, the recrystallized fraction is ~16%, while the HAGB fraction increases to ~44% (Figure 5-6, Figure 5-7). After 60 min, the band-contrast map (Figure 5-5g) remains dominated by elongated regions subdivided by LAGBs, and the EBSD map (Figure 5-6g) shows only limited microstructural change. The recrystallized fraction remains nearly unchanged at ~17%, and the HAGB fraction stays close to ~44% (Figure 5-6, Figure 5-7). Even after 1600 min, the band-contrast map (Figure 5-5h) still shows retained elongated features and residual substructure, while the EBSD map (Figure 5-6h) confirms that the

deformation morphology is not fully removed. The recrystallized fraction increases only slightly to ~26%, and the HAGB fraction reaches ~46% (Figure 5-6, Figure 5-7). Over the investigated range of 30–1600 min, the recrystallized fraction exhibits a strong linear relationship with the logarithm of annealing time ($R^2 = 0.98$), indicating a slow recrystallization response at 350 °C. Therefore, annealing at 350 °C cannot complete the recrystallisation process as some regions resist.

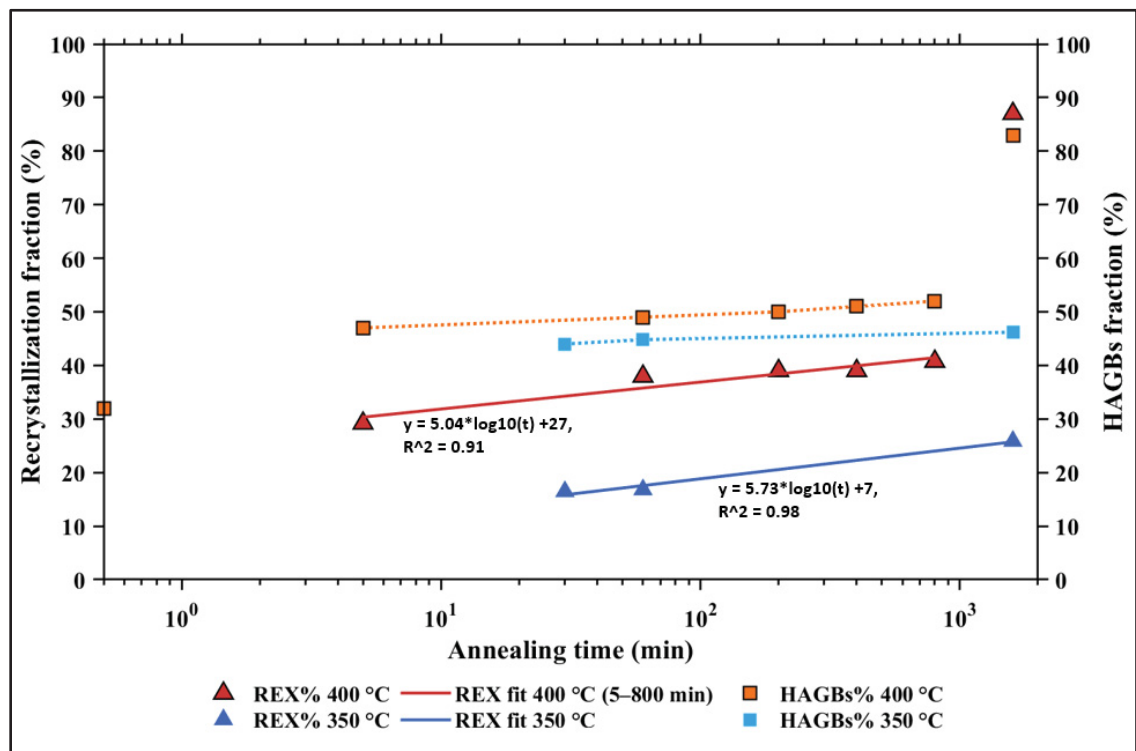


Figure 5-7 Recrystallized fraction versus annealing time for rolled as-extruded specimens annealed at 400 °C and 350 °C and the associated linear fits. The 400 °C–1600 min point is plotted but excluded from the 400 °C fit due to late-time acceleration. HAGB% is plotted on the secondary (right) y-axis

The cumulative grain-size distributions for the 60% rolled and annealed as-extruded conditions are shown in Figure 5-8. In the rolled as-extruded state on the black lines, the distribution is broad with 85% of grains smaller than approximately 170 μm and some extends beyond 200 μm . The cumulative curve rises in discrete steps at larger grain sizes, indicating that a limited number of coarse grains contribute substantial area fractions; these coarse grains likely

correspond to extrusion-aligned bands in Figure 5-6b that remain only weakly subdivided after rolling. After annealing at 400 °C, the curves shift rapidly toward smaller grain sizes. After 5 min, 85% of grains are smaller than approximately 80 μm , and the 60 min condition shows a similar distribution. After 800 min, the distribution changes only slightly as 85% of grains are smaller than approximately 65 μm , but some large region described as coherent grains are found with size even greater than in the as deformed conditions. This may reflect the spatial heterogeneity in the rolled–extruded microstructure as some regions may contain large, elongated regions with similar orientation that can be analyzed as single grain. A pronounced refinement occurs by 1600 min, where 85% of grains are now smaller than approximately 30 μm and all grains are below approximately 65 μm , suggesting that the regions still resisting recrystallization after 800 min are now recrystallized. In contrast, annealing at 350 °C produces a limited refinement response. Although the curves shift slightly toward smaller grain sizes after 30 min, they remain broad, and coarse grains are detected for longer exposure time. A region as large as 250 μm persist even after 1600 min. This indicates a significant grain-size heterogeneity after annealing at 350 °C.

The texture evolutions during annealing at 400 °C and 350 °C are presented in Figure 5-9 as $\{111\}$ pole figures and Figure 5-10 as the evolution of the components with temperature and annealing time. A strong continuous $\langle 111 \rangle$ fibre is found in the as-received material, is aligned with the extrusion direction (ED). The maximum intensity is approximately 13 m.r.d., which is characteristic of hot-extruded aluminum alloys (Figure 5-9a) (Chrominski & Lewandowska, 2022). Considering that the rolling deformation will be done perpendicularly to the ED, the initial texture can be described using typical rolling texture components. In this condition, the main texture component is a Rotated Copper (RCu, ~34%), then Brass (Bs, ~31%), and Cube (Cb, ~16%) accounting for approximately 80% of the measured volume fraction (Figure 5-10). The 60% rolling process (Figure 5-9b) lowers the maximum intensity to approximately 8 m.r.d., weakening of the extrusion-induced texture. The continuous $\langle 111 \rangle$ fibre is no longer visible, and the pole figure resolves into six discrete $\{111\}$ poles arranged symmetrically around TD, with intensities around 5 m.r.d., corresponding mainly to the Brass component. The component analysis confirms this transition: Bs increases to ~63%, RCu decreases to ~1%,

the S component appears at ~11%, and Cube decreases to ~5%, while RG remains relatively low and Cu remains negligible (Figure 5-10).

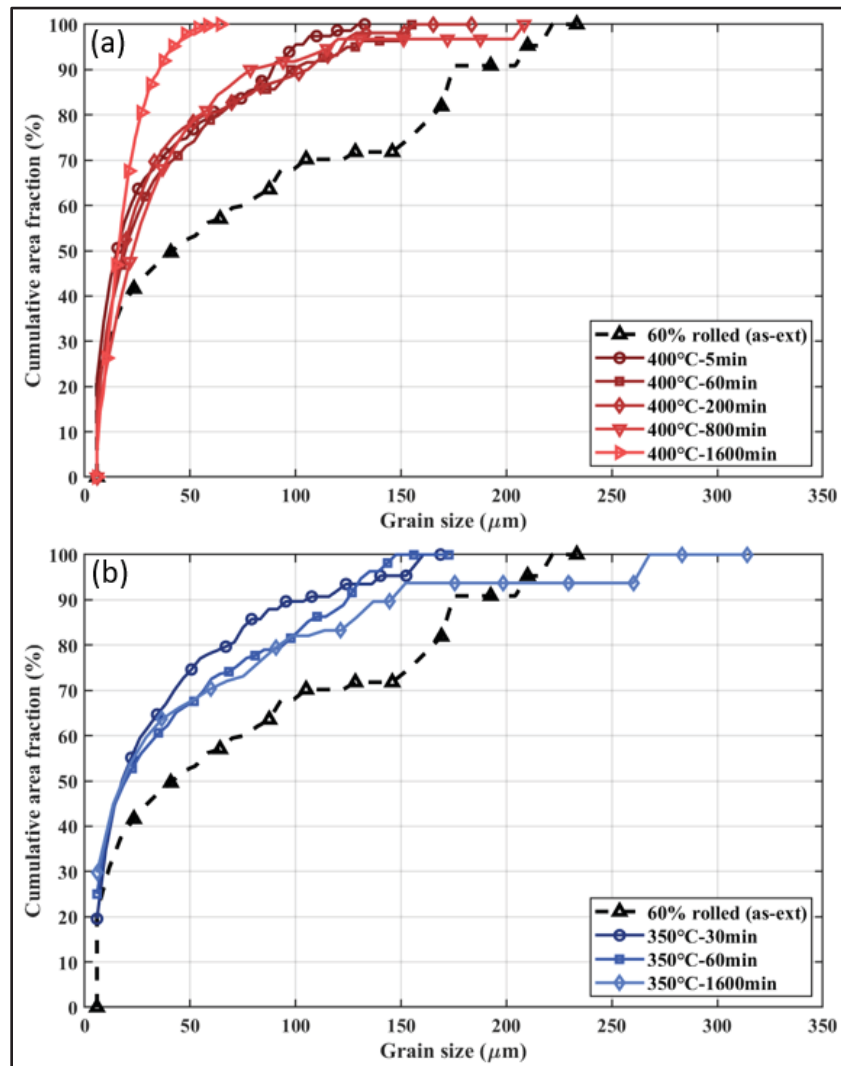


Figure 5-8 Cumulative grain-size distributions for rolled as-extruded after annealing at (a) 400 °C and (b) 350 °C for various times. Each color gradient represents increasing annealing duration, the darker the shorter

After annealing at 400 °C, the deformation texture begins to weaken rapidly and remains relatively stable up to 800 min (Figure 5-9c). Only after 1600 min, the deformation-induced texture transforms into a low-intensity, nearly random texture (Figure 5-9e). The maximum intensity decreases from approximately 6 m.r.d. after 5 min to approximately 4 m.r.d. after 800

min and approximately 2 m.r.d. after 1600 min. The component analysis at 400 °C (Figure 5-10a) shows that the Bs fraction decreases sharply from ~63% after rolling to ~39% after 5 min, to ~28% after 800 min, and only ~6% after 1600 min. The Cube component decreases from ~5% at 5 min to ~2–3% at longer annealing times. RG shows moderate fluctuations during annealing treatment but remains a minor component, while Cu remains negligible.

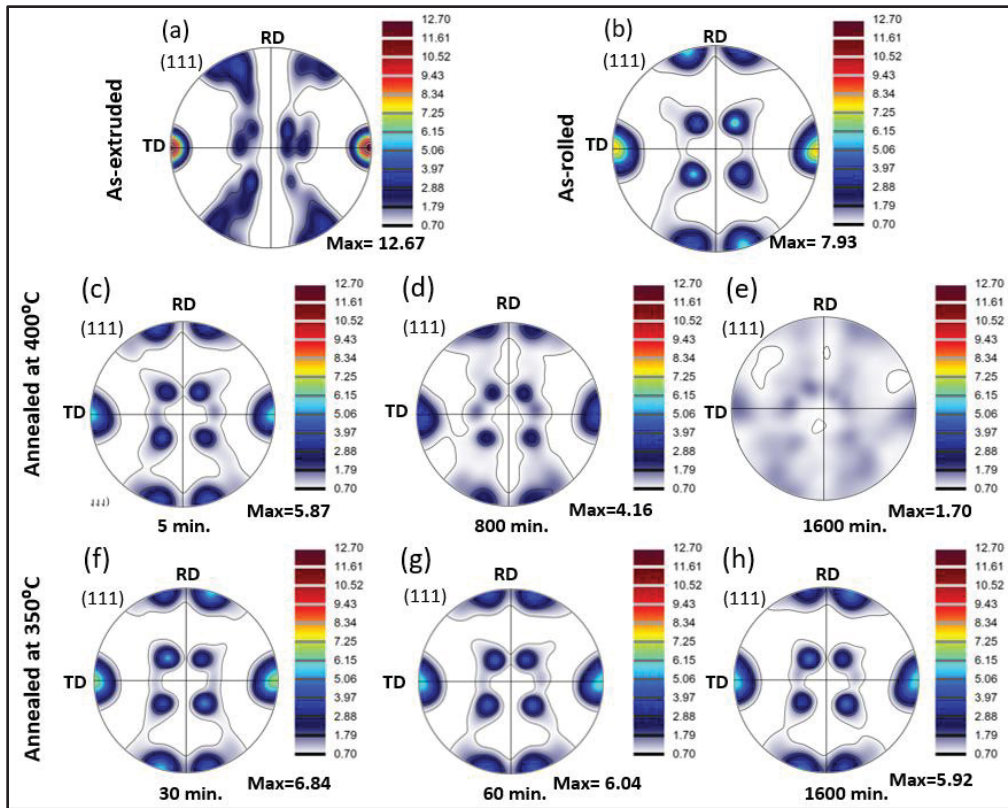


Figure 5-9 {111} Pole figures of (a) as-extruded specimen, (b) rolled as-extruded specimen; annealed at 400 °C for (c) 5 min, (d) 800 min, (e) 1600 min; and annealed at 350 °C for (f) 30 min, (g) 60 min, (h) 1600 min

At 350 °C, the peak positions remain largely unchanged, and only slight weakening occurs over time (Figure 5-9f-h). The maximum intensity changes only slightly from approximately 7 m.r.d. after 30 min to approximately 6 m.r.d. after 60 min remains close to this value after 1600 min. The component analysis at 350 °C (Figure 5-10b) confirms this slow evolution. The S component remains nearly constant at ~14% after 30 and 60 min, then decreases slightly to ~13% after 1600 min. The Bs fraction decreases from its rolled value of ~63% to ~54% after

30 min and ~48% after 1600 min. RCu increases slightly from ~1% in the rolled condition to ~3% after 1600 min, while RG and Cube show only minor changes throughout this temperature range.

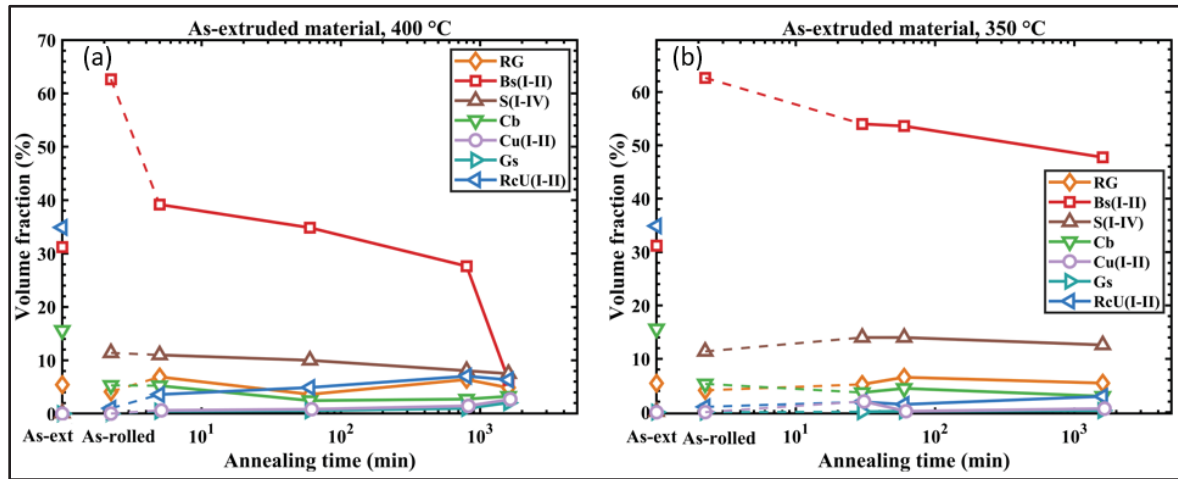


Figure 5-10 Texture of the main components volume fractions as a function of annealing time for rolled as-extruded specimens at a) 400 °C and b) 350 °C

5.2.4.3 Texture Persistence and Stability Across Annealing Conditions

The persistence of deformation textures and their stability against recrystallization were analyzed by comparing the recrystallized and NR fractions under different annealing conditions. For each initial material, the analysis first describes the microstructural and morphological features, followed by the quantitative evaluation of texture components using metrics describing (i) the retention index (RI), calculated from the NR-weighted texture-component fraction normalized by the corresponding rolled reference, and (ii) resistance score, (RS) representing substructure stability through the equal-weight composite of GOS and GOS/*Di*.

I. Microstructure and Texture Evolution under Annealing

The microstructure and texture evolution of both the rolled as-cast and rolled as-extruded materials during annealing were evaluated by separating EBSD data into recrystallized and NR subsets using the GOS and GOS/*Di* thresholds defined in Section 5.2.3.5. The analysis first considers the rolled as-cast material annealed at 400 °C for 5 and 10 min, as these conditions capture the main transition from deformation to recrystallization. The corresponding filtered EBSD maps, grain-size distributions, and {111} pole figures are presented in Figure 5-11 to Figure 5-13. The 30-minute condition shows morphologies similar to 10 min and is not repeated.

At 5 min (Figure 5-11a, b), the microstructure is dominated by NR grains; the recrystallized subset consists of a few small, isolated grains dispersed within the matrix. The grain-size distributions (Figure 5-12a, b) show that the NR fraction covers a wide range, with sizes up to ~88 μm , whereas the recrystallized fraction is concentrated below ~50 μm , with ~90% of grains below 23 μm . The recrystallized CDF rises steeply at small sizes, consistent with a fine-grained subset. The {111} pole figures (Figure 5-13a, b) show a strong deformation-type texture for NR (max ≈ 4.6) and a weak, low intensity texture for recrystallized (max ≈ 2.6). The comparison between the initial rolled texture and the NR-weighted texture-component fractions (Figure 5-13c) shows that Bs(I–II), S(I–IV), and Cu(I–II) remain the dominant deformation components. Their NR-weighted fractions remain close to their initial values, consistent with the high NR fraction after 5 min. The S component remains the most prominent, followed by Bs and Cu.

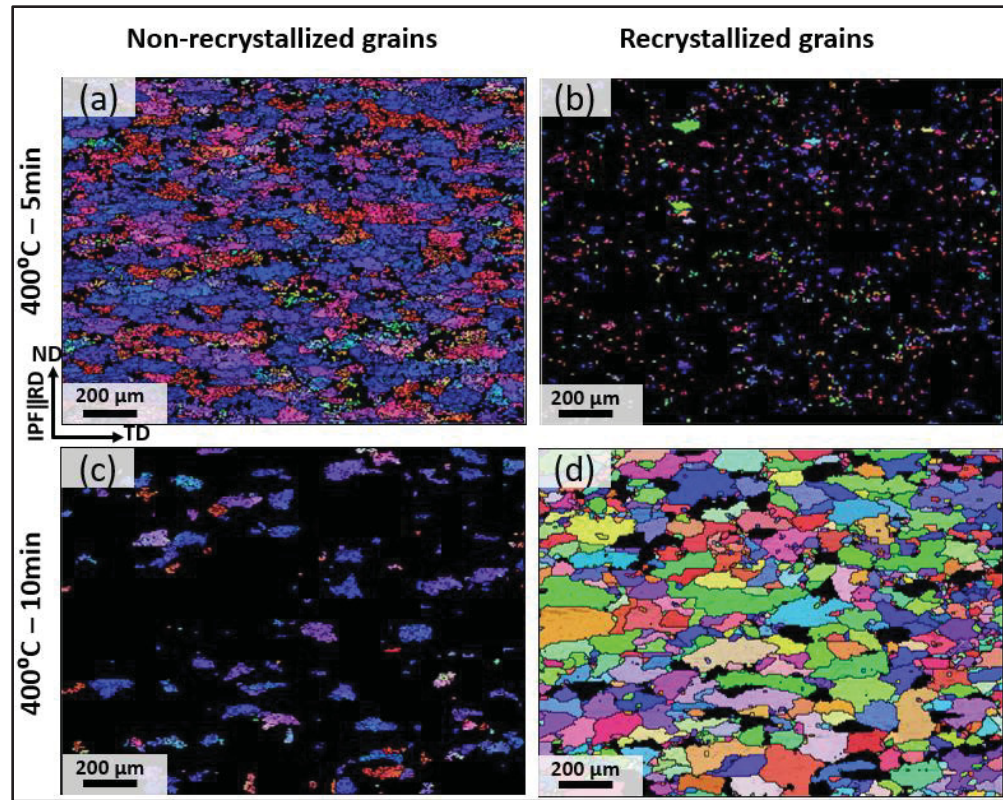


Figure 5-11 Orientation maps of the rolled as-cast AA6082 alloy annealed at 400 °C for 5 and 10 min, separated into non-recrystallized (NR) and recrystallized subsets based on the GOS and GOS/D criteria. (a, b) NR and REX at 5 min; (c, d) NR and REX at 10 min, respectively

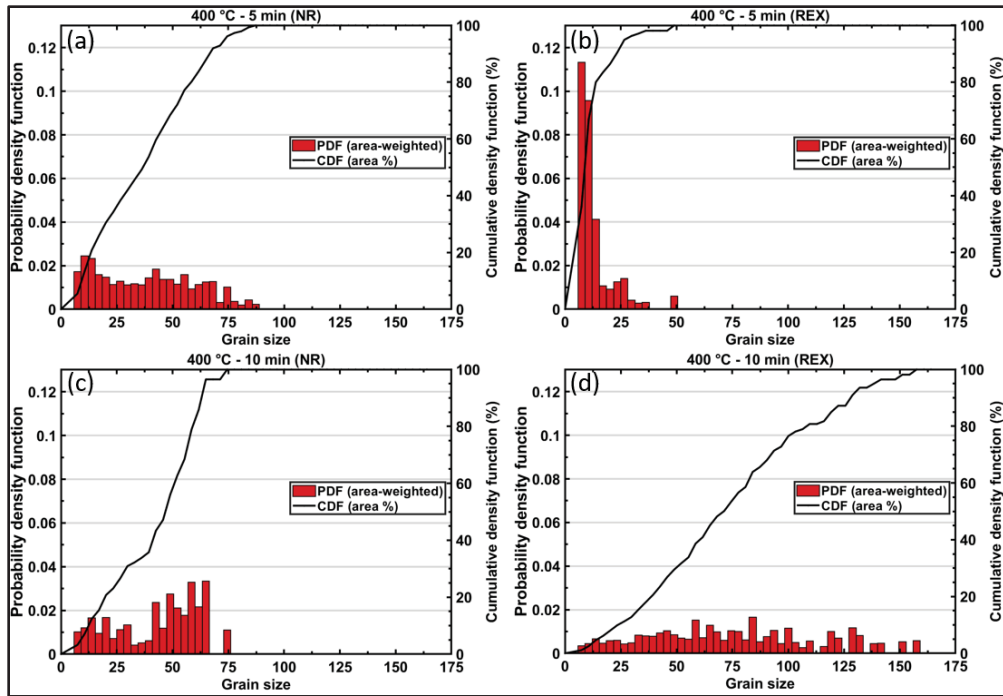


Figure 5-12 Grain-size distributions of the rolled as-cast condition at 400 °C. For each annealing time, the probability density function (PDF, red bars) and cumulative density function (CDF, black line) are plotted separately for non-recrystallized (NR) and recrystallized (REX) regions. (a, b) NR and REX at 5 min, respectively; (c, d) NR and REX at 10 min, respectively

After 10 min (Figure 5-11c, d), the recrystallized fraction increases significantly. The NR area is reduced and appears as isolated patches between recrystallized regions composed of larger grains with a large variety of orientations. The grain-size plots (Figure 5-12c, d) show that the NR histogram narrows with a main peak near 50–70 μm. The recrystallized subset shifts to a broader range than at 5 min and extends to ~157 μm; its CDF reaches saturation at larger sizes than at 5 min. In the pole figures (Figure 5-13d, e), NR displays a sharper texture with a higher maximum intensity (≈ 6.1), while recrystallized remains weak and has a random texture (max ≈ 1.6). The global texture (Figure 5-13f) shows decreased Bs(I–II) and S(I–IV) intensities, but these two components remain dominant in the NR regions with Bs(I–II) slightly exceeding S(I–IV) and a reduced amount of Cu compared with 5 min.

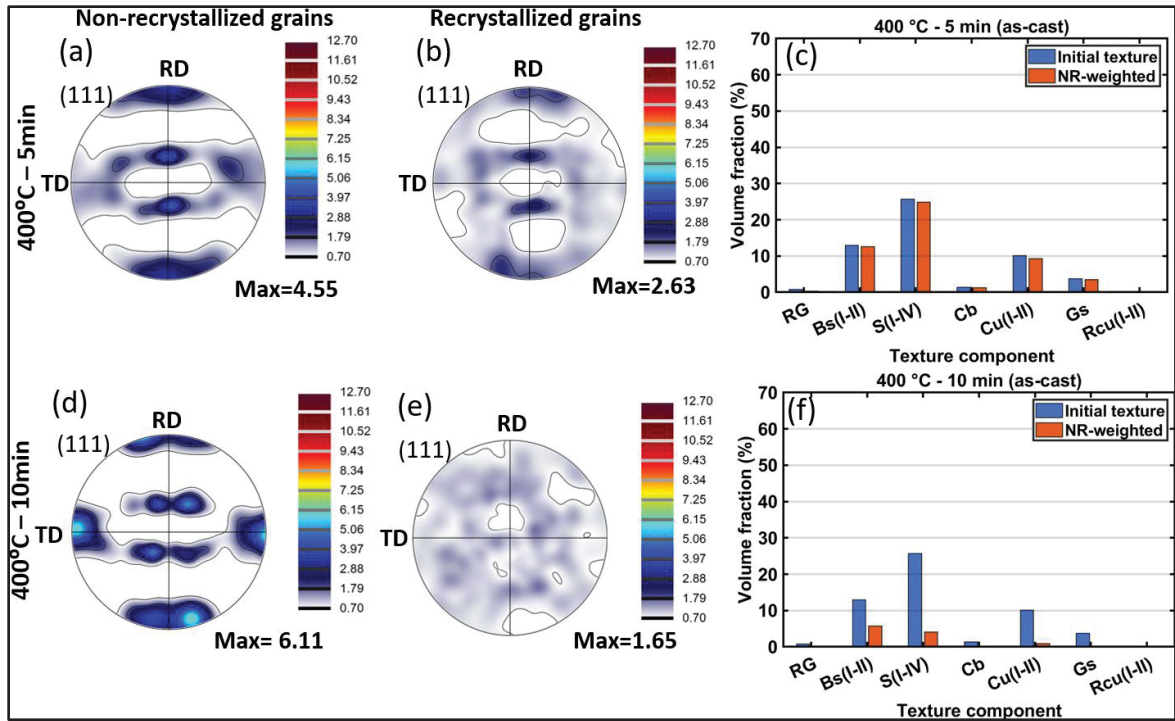


Figure 5-13 Texture analysis of the rolled as-cast material annealed at 400 °C. (a, b) NR and REX pole figures after 5 min; (c) texture-component fractions in the rolled reference and NR-weighted fractions after 5 min. (d, e) NR and REX pole figures after 10 min; (f) texture-component fractions in the rolled reference and NR-weighted fractions after 10 min

For the rolled as-extruded material, recrystallization proceeded much more slowly than in the rolled as-cast condition. Therefore, the 1600 min condition was selected at both temperatures to isolate the temperature effect under an identical holding time. This comparison reveals a pronounced difference, with recrystallized fractions of approximately 87% at 400 °C and 26% at 350 °C. At 400 °C (Figure 5-14a, b), only thin NR bands remain; however, the recrystallized microstructure is not fully equiaxed, as elongated grains are still present. The recrystallized grains have an area-weighted mean equivalent diameter of $20.7 \pm 11.5 \mu\text{m}$. At 350 °C (Figure 5-14c, d), most NR grains form regions of various width elongated along the transverse direction (prior extrusion direction) and display preferential orientations with $\langle 111 \rangle$ and sometimes $\langle 101 \rangle$ directions aligned with the rolling direction. The recrystallized grains have an area-weighted mean equivalent diameter of $16.5 \pm 10.0 \mu\text{m}$ and are distributed within elongated bands and exhibit mixed morphologies, ranging from small, relatively equiaxed grains to larger elongated grains.

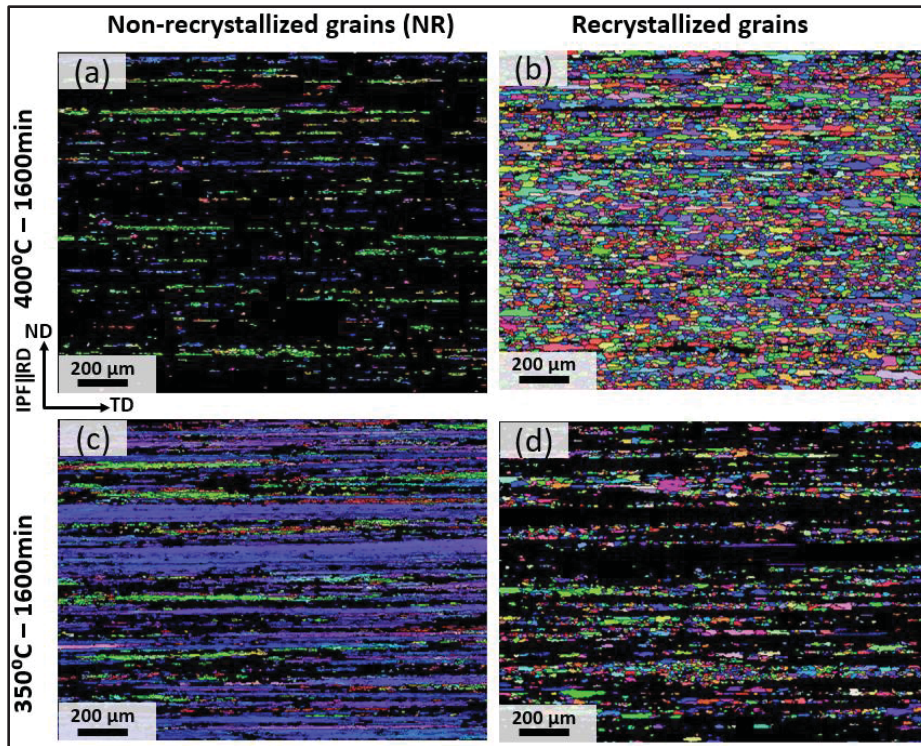


Figure 5-14 Orientation maps of the 60% cold-rolled as-extruded after annealing for 1600 min, separated into non-recrystallized (NR) and recrystallized (REX) grains. (a, b) NR and REX at 350 °C, respectively; (c, d) NR and REX at 400 °C, respectively

The grain-size distributions in Figure 5-15 further illustrate the difference in behavior between the two temperatures. At 400 °C, the remaining NR regions have an area-weighted mean equivalent diameter of $17.4 \pm 8.9 \mu\text{m}$ consistent with the limited and thin NR regions observed in the microstructure. The recrystallized grains have a larger area-weighted mean equivalent diameter of $20.7 \pm 11.5 \mu\text{m}$, representing an increase of $3.3 \mu\text{m}$, or approximately 19%, relative to the NR regions. At 350 °C, the NR grain-size distribution is considerably broader and strongly right-skewed, with an area-weighted median diameter of $64 \mu\text{m}$ and an interquartile interval of 18.5 to $143 \mu\text{m}$. Its long tail extends to approximately $320 \mu\text{m}$, indicating the coexistence of relatively small and very large NR grains. At this temperature, the recrystallized grains are concentrated at smaller sizes ($16.5 \pm 10.0 \mu\text{m}$) than the NR grains.

The aspect-ratio distributions shown in blue in the same figure reveal a marked difference between the NR grains at 400 °C and 350 °C. At 400 °C, approximately half of the NR grains exhibit aspect ratios above 0.4, whereas at 350 °C, nearly half have aspect ratios below 0.2,

indicating a much more elongated morphology. In contrast, the REX subsets show similar aspect-ratio distributions, mainly characterized by intermediate to high values, although a fraction of elongated grains remains present.

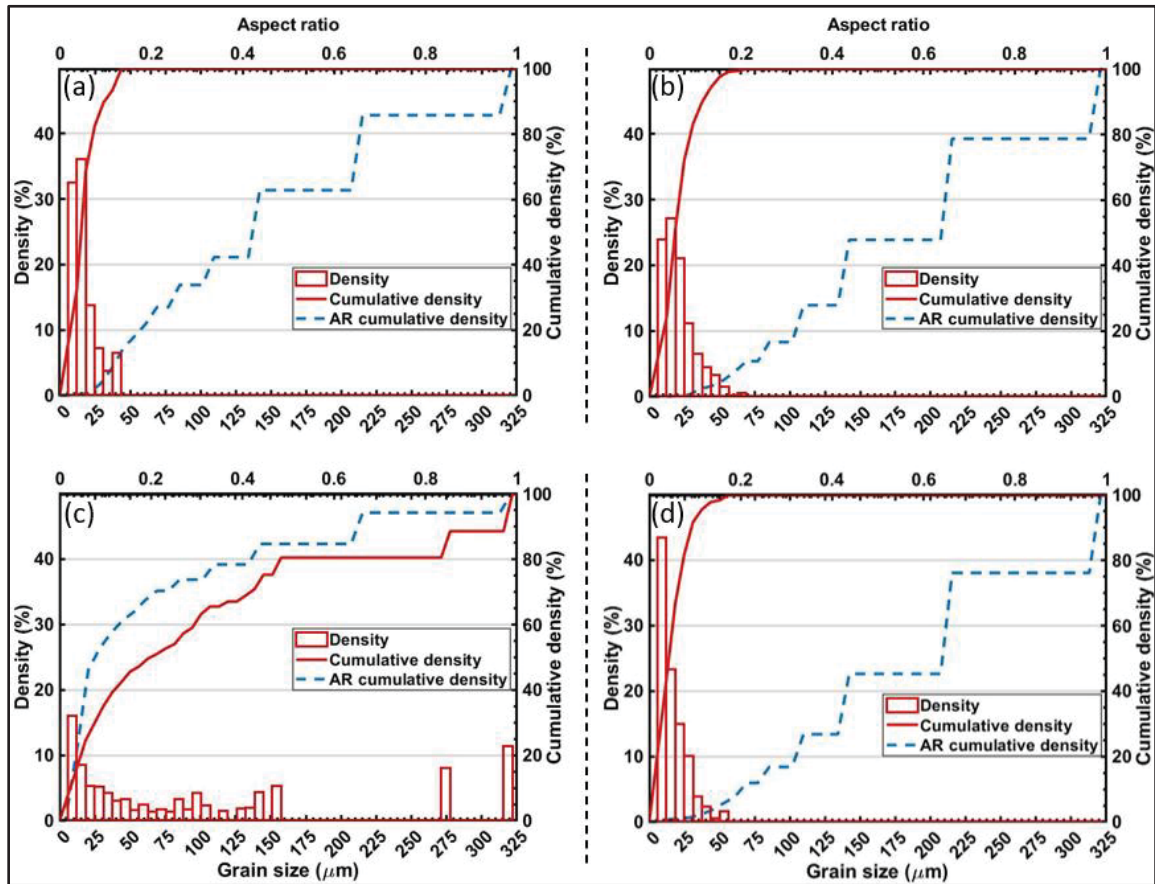


Figure 5-15 Grain-size and aspect-ratio distributions of the rolled as-extruded material annealed for 1600 min at (a, b) 400 and (c, d) 350 °C for (a, c) NR and (b, d) REX grains. Grain-size probability density and cumulative density are shown by the red bars and solid red lines, respectively, while the aspect ratio cumulative density is represented by the dashed blue lines

The corresponding pole figures (Figure 5-16a, b, d, e) show the texture changes in the NR and recrystallized subsets at 400 and 350 °C after 1600 min. At 400 °C, both subsets exhibit reduced intensity (NR ≈ 5.3 , REX ≈ 1.6). The recrystallized subset at 400 °C presents a random and low texture intensity, indicating that recrystallized grains do not share preferred orientations coming from the deformed state. After annealing at 350 °C, both subsets show a

deformation texture, the NR subset with a high maximum intensity ≈ 7.8 similar to the texture measured on the all sample before annealing, whereas the recrystallized subset with a lower intensity ≈ 2.7 .

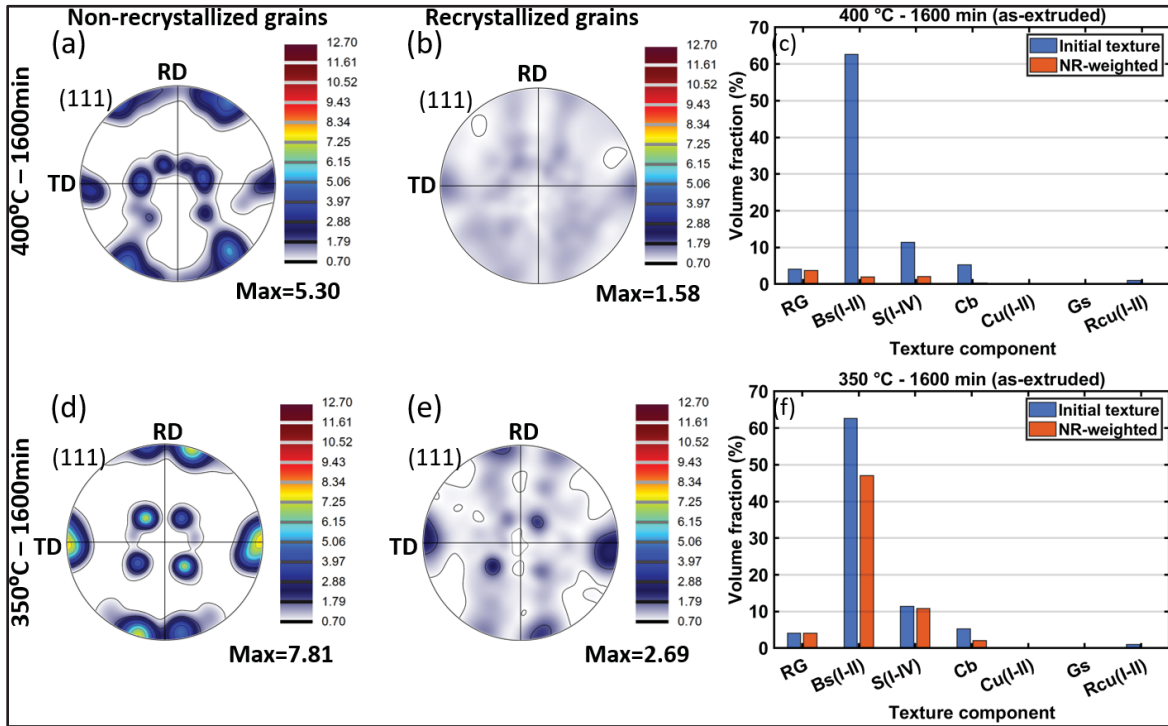


Figure 5-16 Texture analysis of the rolled as-extruded material after annealing for 1600 min.

(a, b) NR and REX pole figures at 400 °C; (c) texture-component fractions in the rolled reference and NR-weighted fractions at 400 °C. (d, e) NR and REX pole figures at 350 °C; (f) texture-component fractions in the rolled reference and NR-weighted fractions at 350 °C

The texture-component volume fractions (Figure 5-16c, f) further demonstrate these differences. At 400 °C, all NR-weighted fractions are substantially lower than the initial rolled fractions because only a small NR fraction remains after annealing. RG becomes the largest NR-weighted component, followed by S(I-IV) and Bs(I-II), while the initially dominant Bs(I-II) component shows a pronounced decrease. At 350 °C, Bs(I-II) remains the dominant NR-weighted component, followed by S(I-IV), whereas RG and Cb remain minor. The NR-weighted fractions at 350 °C are much closer to the initial rolled texture, indicating stronger retention of the deformation components than at 400 °C.

II. Integrated Assessment of Texture Persistence and Stability

Quantitative assessments of texture persistence were performed to document the retention and substructure stability of specific crystallographic components using Retention Index (RI) and Resistance Score (RS) defined in equations (5-5) and (5-7). Results are displayed using color mapping in Figure 5-17 and Figure 5-18. An *RI* value close to 1 indicates that the NR-weighted fraction remains comparable to that of the corresponding 60% cold-rolled reference, whereas a value approaching zero indicates limited persistence in the non-recrystallized regions. The retention behavior varies with texture component, annealing temperature, and exposure time. In the as-cast material, Bs(I-II), S(I-IV), and Cu(I-II) exhibit similar high retention after 5 min at 400 °C, with *RI* values close to 1. With increasing exposure time, the retention of all three components decreases, reflecting the progressive consumption of the rolled deformation texture. However, Bs remains the most persistent component after 10 and 30 min, with *RI* values of 0.45 and 0.24, respectively, as shown in Figure 5-17.

A different trend is observed for the as-extruded specimens. At 400 °C for 60 min, RG and S show relatively high retention, with *RI* values of 1.00 and 0.67, respectively, while Bs reaches 0.50. At longer exposure times, RG remains the dominant retained component, reaching 0.98 after 800 min and remaining high at 0.90 after 1600 min. In contrast, Bs and S decrease to 0.03 and 0.19, respectively, after 1600 min. This indicates that the remaining non-recrystallized regions become increasingly enriched in RG orientation during prolonged annealing. At 350 °C, the rolled texture is retained more effectively, with RG and S remaining close to *RI* = 1 even after 1600 min, while Bs also maintains a relatively high value of 0.75. These results demonstrate that texture persistence is substantially greater at 350 °C than at 400 °C.

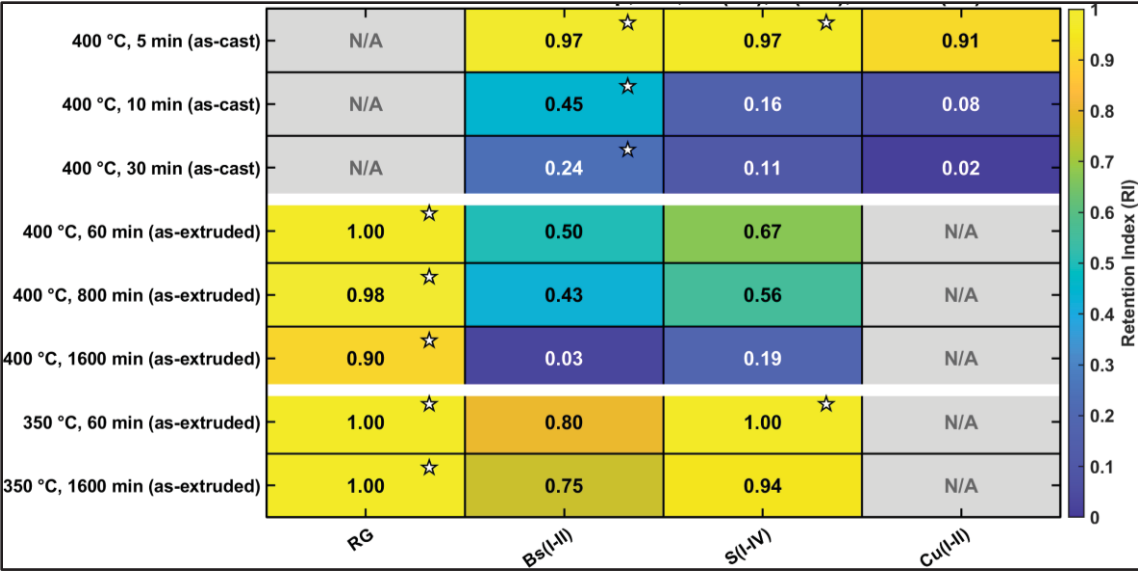


Figure 5-17 Retention Index (RI) for each condition and texture component; ☆marks the highest-RI component in each condition. White numbers give RI values

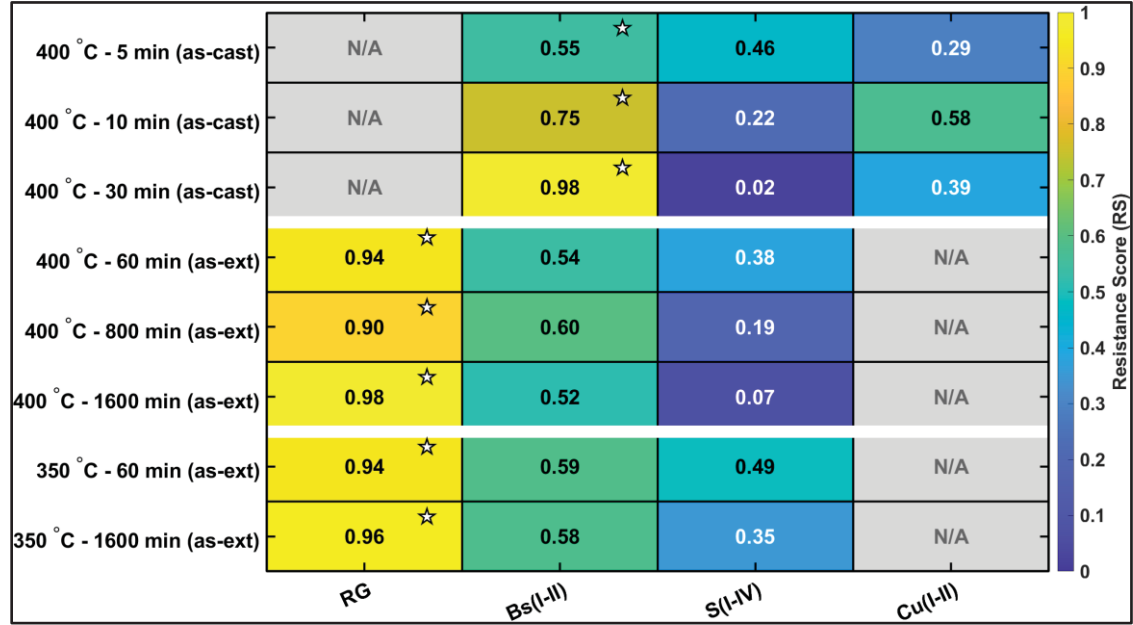


Figure 5-18 Resistance Score (RS), calculated from the equal-weight normalized combination of GOS and GOS/D, for each condition and texture component; ☆ indicate all components exhibiting the maximum RS within each condition

The RS shows the same general dependence on texture component, annealing temperature, and exposure time (Figure 5-18). In this representation, a higher RS value indicates that the corresponding non-recrystallized component retains a more stable deformation substructure, as reflected by the combined contribution of GOS and GOS/D. In the as-cast series, Bs(I-II) exhibits the highest RS values at all annealing times, increasing from 0.55 after 5 min to 0.75 after 10 min and reaching 0.98 after 30 min, as shown in Figure 5-18. In contrast, S(I-IV) decreases from 0.46 to 0.22 and then to 0.02, while Cu(I-II) remains lower, with values of 0.29, 0.58, and 0.39. These results indicate that, although the retention of all components decreases with recrystallization, the remaining Bs-oriented non-recrystallized regions preserve the most stable deformation substructure in the as-cast condition.

A comparable trend is observed in the as-extruded specimens, although the ranking differs because of the pre-existing extrusion texture. At 400 °C, RG exhibits the highest RS values throughout the annealing sequence, with values of 0.94, 0.90, and 0.98 after 60, 800, and 1600 min, respectively. Bs(I-II) remains at an intermediate level, between 0.52 and 0.60, whereas S(I-IV) decreases progressively from 0.38 to 0.19 and then to 0.07. At 350 °C, RG remains the dominant component, with values of 0.94 and 0.96 after 60 and 1600 min, respectively, while Bs(I-II) remains nearly constant at 0.59 and 0.58. S(I-IV) is better preserved at 350 °C than at 400 °C, with values of 0.49 and 0.35.

The RI and RS results identify the components resisting the recrystallization process during annealing for the two starting conditions. In the rolled as-cast material, it is the Bs orientation that is the main stable component as it exhibits the highest RI and RS values. In the rolled as-extruded material, RG maximizes both RI and RS, particularly after prolonged exposure, such as 1600 min at 400 °C. This difference illustrates clearly how the texture present before annealing strongly influences which deformation-texture component remains most prominent in the non - recrystallized regions.

5.2.4.4 Second-phase particles Evolution During Annealing

To identify the second-phase particles that may interact with migrating grain boundaries during recrystallization, equilibrium phase stability was evaluated at temperatures relevant to hot extrusion and annealing. At approximately 520 °C, the FactSage calculation predicts the presence of $\text{Al}_8\text{FeMg}_3\text{Si}_6$, $\alpha\text{-AlMnSi}$, and $\beta\text{-AlFeSi}$, while the calculated fractions of Mg_2Si and $\text{Al}_{13}\text{Cr}_4\text{Si}_4$ are nearly zero. At 400 and 350 °C, $\text{Al}_8\text{FeMg}_3\text{Si}_6$, $\alpha\text{-AlMnSi}$, Mg_2Si , and $\text{Al}_{13}\text{Cr}_4\text{Si}_4$ are predicted to be stable. The calculated fractions of the Fe-, Mn-, and Cr-bearing phases change only slightly between 400 and 350 °C, whereas the Mg_2Si fraction increases as the temperature decreases to 350 °C, as shown in Figure 5-19. These predictions establish which particle populations are expected to persist during annealing and therefore remain available to influence recrystallization through nucleation near large particles or restriction of grain-boundary migration.

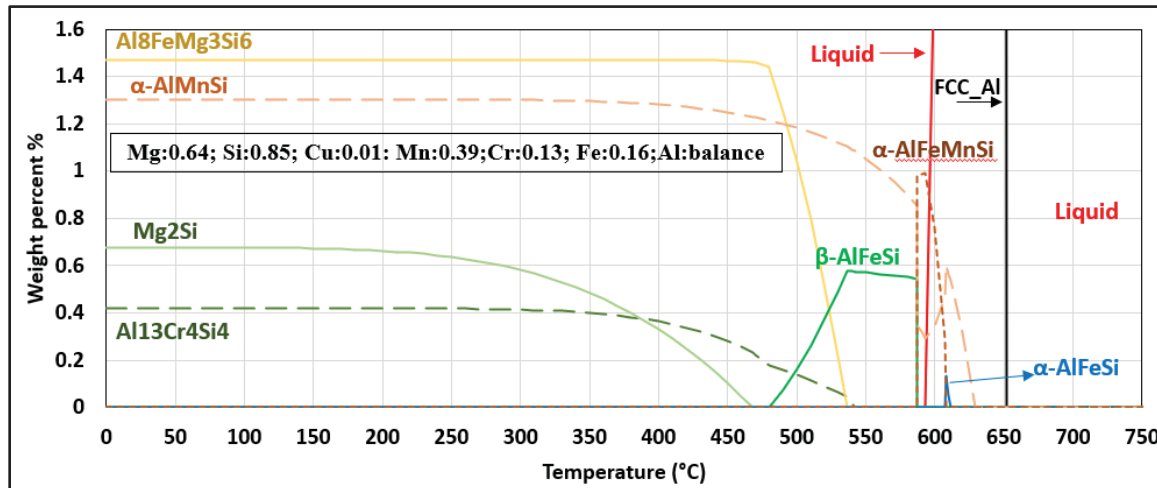


Figure 5-19 Equilibrium phase percentage for the AA6082 alloy as a function of temperature

The chemical composition of the particles found in the as-extruded rolled samples was documented using EDS elemental mapping and are shown in Figure 5-20. Fe and Mn show a more continuous band-like distribution, whereas Mg- and Si-rich particles do not exhibit the same continuous distribution. Localized Cr enrichment overlaps with some of the Fe- and Mn-rich particles, indicating the presence of Fe-Mn-Cr-bearing constituents. Only limited Mg-Si

enrichment is observed, which is consistent with the equilibrium prediction in Figure 5-19 showing that most Mg_2Si is dissolved at the extrusion die-exit temperature of approximately 520 °C.

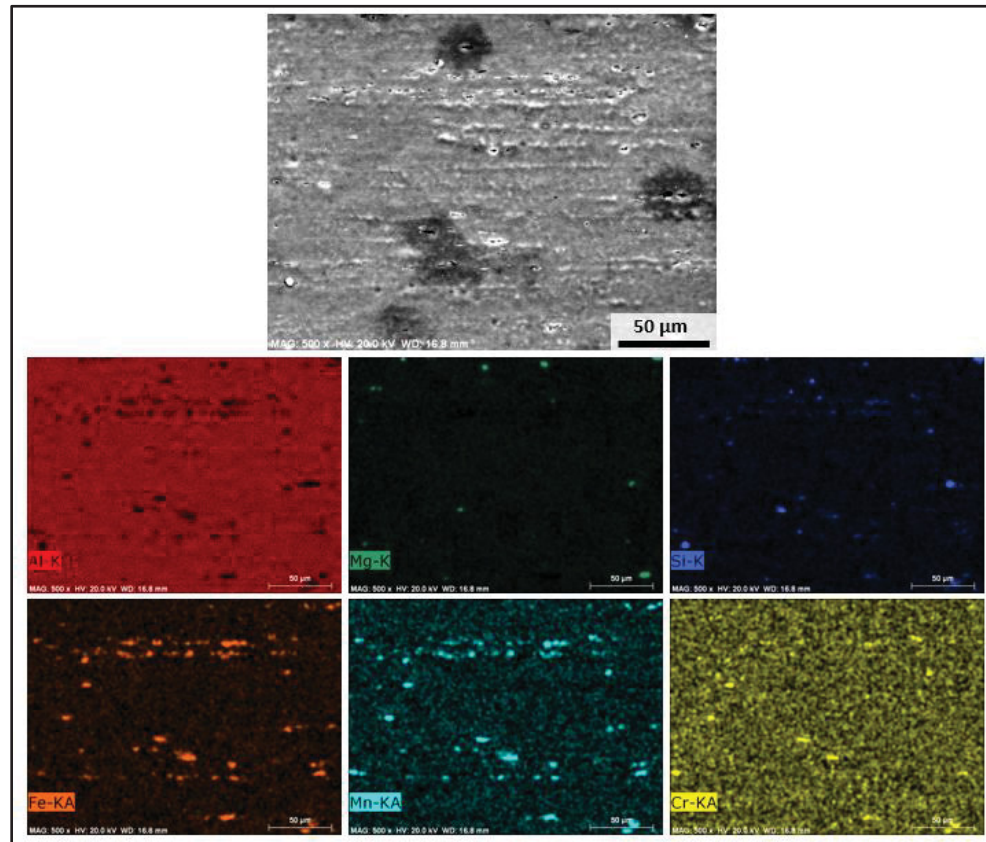


Figure 5-20 EDS elemental maps of the rolled as-extruded specimen showing the distribution of major alloying elements (Al, Si, Mg, Fe, Mn, and Cr)

The direct relationship between second-phase particles and recrystallization was examined in the rolled as-extruded specimen annealed at 400 °C for 800 min. This condition was selected because its partially recrystallized microstructure contains both recrystallized and non-recrystallized regions, allowing the particle locations to be examined relative to recrystallized grain boundaries. The BSE image in Figure 5-21a reveals elongated particle bands aligned with TD, which corresponds to the prior extrusion direction.

The interaction between second-phase particles in both, recrystallized and NR regions, is illustrated in Figure 5-21b where only the orientations of the recrystallized grains are shown on top of the BSE image. Several large particles, marked by red circles, are located directly on or adjacent to the recrystallized grain boundaries. These relatively large particles, with an average equivalent diameter of $3.8 \pm 1.8 \mu\text{m}$, together with the presence of particles extending to considerably larger sizes, support the possibility that part of the particle population contributed to PSN during annealing. The EDS maps in Figure 5-21c reveal particle populations similar in composition to those observed in the as-extruded rolled condition in Figure 5-20, including Fe-Mn-rich and Mg-Si-rich particles. After annealing at 400 °C for 800 min, the Mg-Si-rich particles remain clearly visible, which is consistent with the equilibrium prediction in Figure 5-19, where Mg_2Si is predicted to be stable at this temperature. Fe- and Mn-rich particles are also retained after annealing, while localized Cr enrichment is observed in association with some of these particles.

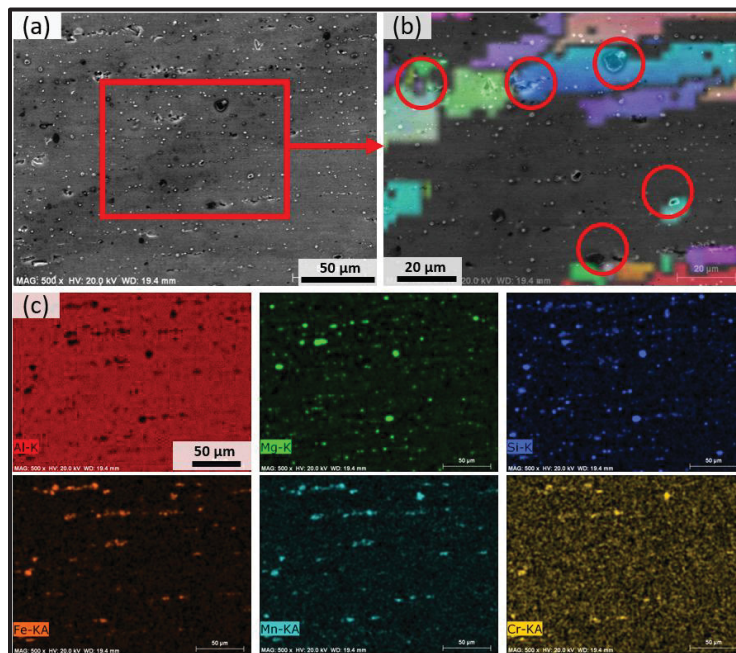


Figure 5-21 Rolled as-extruded sample after annealing at 400 °C for 800 min: (a) SEM-BSE image; (b) higher-magnification view of the boxed area in (a) with EBSD overlay, circled particles lie on/near grain boundaries; (c) corresponding EDS maps of (a) for Al, Mg, Si, Fe, Mn, and Cr

The overall effect of annealing on particle size and morphology was then evaluated by comparing the rolled as-extruded reference condition with the specimen annealed at 400 °C for 1600 min (Figure 5-22). The BSE image was processed using filtering to isolate only the second phase particles. The rolled as-extruded condition in Figure 5-22a shows a banded distribution of second-phase particles aligned with TD, corresponding to the prior extrusion direction. Many particles within these bands exhibit a distinctly elongated morphology. The particle area fraction was 5.3%, and the area-weighted mean equivalent diameter was $5.4 \pm 3.8 \mu\text{m}$. The quantitative results for the rolled condition are presented in Figure 5-23. The broad diameter distribution and its extension toward larger values indicate that coarse particles account for a substantial portion of the total particle area. The area-weighted elongation ratio was 2.2 ± 1.1 , but some particles are elongated as much as 8. The polar inset in Figure 5-23b shows a preferred particle elongation along TD (prior extrusion direction), with a mean resultant length of $R=0.5$, calculated using (5-10), indicating that most of the particle were relatively concentrated around this direction. The narrow concentration around 0° confirms the pronounced directional arrangement of the particle bands.

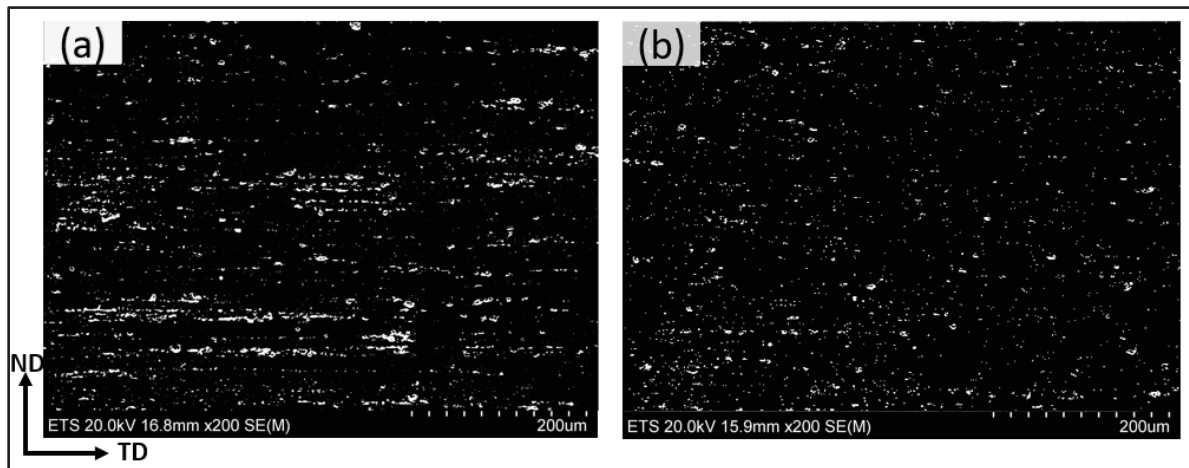


Figure 5-22 Second-phase particles in (a) rolled as-extruded AA6082 specimen, (b) after annealing at 400 °C for 1600 min

After annealing at 400 °C for 1600 min, the particle distribution appears less strongly banded and contains fewer coarse and highly elongated particles, as shown in Figure 5-22b.

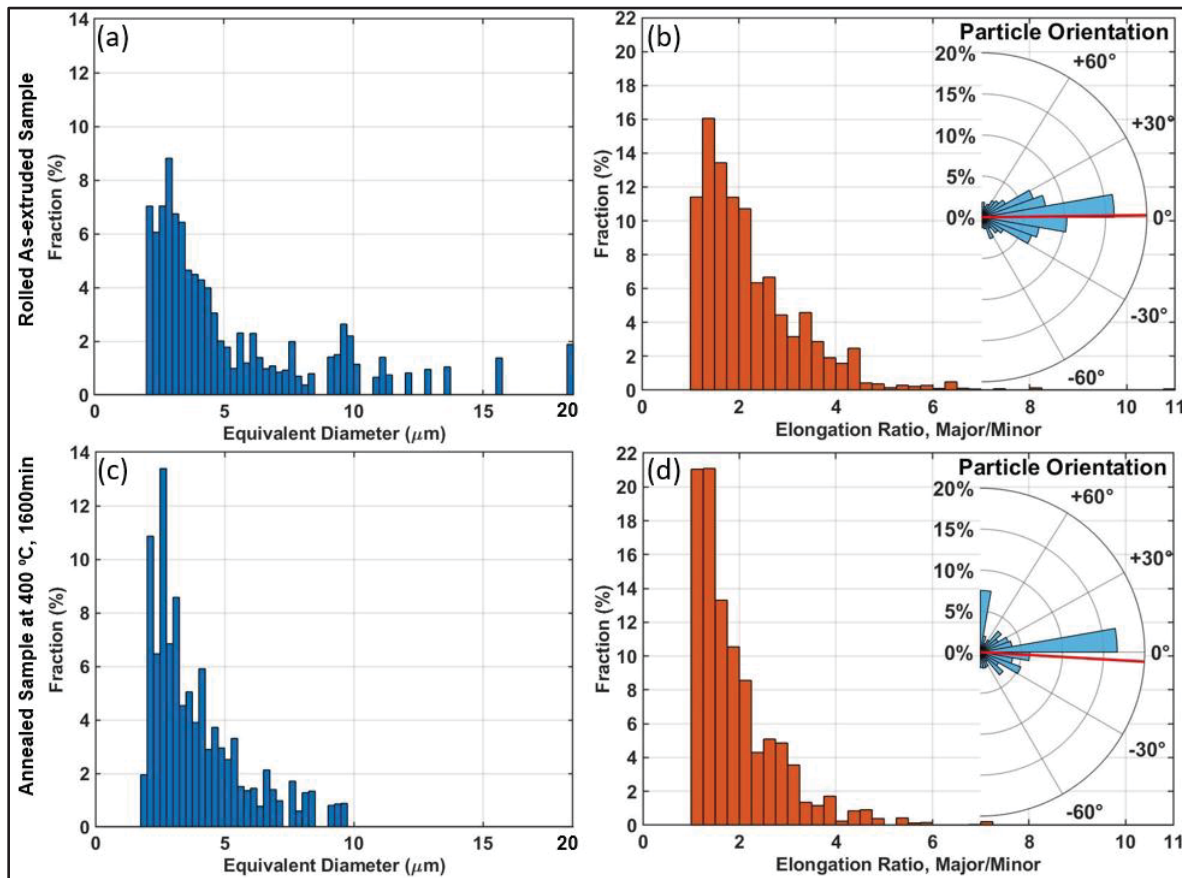


Figure 5-23 Area-weighted distributions of equivalent particle diameter and elongation ratio for the rolled as-extruded AA6082 specimen and the specimen annealed at 400 °C for 1600 min in the rolled (a,b) and annealed (c,d) conditions: (a,c) equivalent diameter, (b,d) elongation ratio in the rolled condition

Quantitative analysis confirms these morphological changes (Figure 5-23). The particle area fraction decreases from 5.3% to 3%. This measured value is slightly lower than the FactSage equilibrium phase fraction of approximately 3.4% at 400 °C, with a relative difference of about 12%. These values are not directly equivalent, as the FactSage prediction (Figure 5-19) assumes equilibrium conditions, whereas the experimental sample may not have reached full equilibrium during annealing and was subsequently quenched, limiting further phase transformations during cooling. The area-weighted mean equivalent diameter and related

standard deviation decrease from 5.4 ± 3.8 to 4.0 ± 1.8 μm . The reduction in standard deviation indicates a narrower particle-size distribution and the disappearance of the larger ones during the annealing process. Since the Zener pinning pressure is proportional to the particle fraction divided by the particle radius ($P_Z \propto f/r$), the measured values indicate a decrease in f/r of about 24%. This indicates a lower relative particle-pinning contribution after prolonged annealing. Furthermore, the average elongation ratio decreases but less significantly from 2.2 ± 1.1 to 1.9 ± 0.9 . A relatively large standard deviation is still present. The preferred orientation near TD (prior extrusion direction) remains visible in the image and in the polar inset (Figure 5-23d), however it seems to be composed of particles strictly aligned in the TD direction and others randomly aligned in the others. Consistent with this observation, the mean resultant length, (R), decreases from 0.5 in the rolled condition to 0.3 after annealing, indicating a lower concentration of particle orientations after heat treatment, as quantified using (5-10).

5.2.5 Discussion

The two materials exhibited substantially different recrystallization responses after 60% rolling reduction and annealing at 400 °C. In the rolled as-cast material, the recrystallized fraction increased from approximately 8% after 5 min to 85% after 10 min, indicating that recrystallization occurred abruptly within a narrow time interval (Figure 5-3). By contrast, the rolled as-extruded material reached approximately 30% recrystallization after 5 min, progressed only gradually up to 800 min, and increased to approximately 87% after 1600 min (Figure 5-7). Thus, substantial regions of the rolled as-extruded material remained non-recrystallized during prolonged annealing.

5.2.5.1 Recrystallization of rolled as-cast material

After 5 min, the band-contrast map shows a dense LAGB network and only a few isolated regions display HAGBs (Figure 5-2a), indicating that recovery and subgrain formation were still the dominant active mechanism. During the following 5 min, some subgrain boundaries developed into higher misorientations and became mobile thanks to the driving force provided

by the available stored energy in the recovered region. As a result, the HAGB fraction increased from approximately 29% to 68% (Figure 5-3a). At the same moment, the maximum texture intensity decreased from about 3.9 to 2.2 (Figure 5-2h) as new orientations appear and the deformed texture fade. These changes indicate that mobile boundaries and enough stored energy are available in most of the material. Limited changes occurred after 10 min, confirming that the main recrystallization stage occurred between 5 and 10 min, consistent with the well known incubation-and-growth behavior reported for cold-deformed aluminum alloys (S. Zhang et al., 2022). However, the recrystallization is not completed after 30 min, as some regions seem unable to generate HAGB or store enough energy to trigger recrystallization.

The recrystallized population develops a weak crystallographic texture from the early stage of recrystallization and does not show any preferred orientation as recrystallization proceeds. After 5 min, when only approximately 8% of the material is classified as recrystallized, the recrystallized subset consists mainly of small grains. The associated area-weighted mean equivalent diameter is $13 \pm 7.7 \mu\text{m}$. This relatively small grain size is consistent with the early stage of recrystallization, where newly nucleated recrystallized grains have formed but have undergone only limited subsequent growth. and the recrystallized grains show a weak texture with a maximum intensity of approximately 2.6 (Figure 5-13b). As the recrystallized fraction increases to approximately 85% after 10 min, the recrystallized grain-size distribution broadens considerably and the area-weighted mean equivalent diameter increases to $75 \pm 37 \mu\text{m}$, documenting a substantial growth of the recrystallized population (Figure 5-12). This increase in grain size and recrystallized fraction is accompanied by decrease in texture intensity to approximately 1.6 (Figure 5-13e), showing that recrystallization proceeds through grains with a broad range of orientations rather than through the preferential development of a single recrystallization component. Figure 5-2 With increasing annealing time to 30 min, the recrystallized fraction increases approximately to 93% and the area-weighted mean equivalent diameter changes only slightly to $77 \pm 37 \mu\text{m}$. The limited change compared with 10 min indicates that the recrystallized grain size becomes relatively stable at this stage, as recrystallization approaches completion and further growth is increasingly restricted by impingement between neighboring recrystallized grains. The recrystallized grains exhibit a weak texture. Among the recrystallized grains, the Cube orientation shows the clearest

development, increasing from approximately 1.3% of the recrystallized population after 5 min to 7.8% after 30 min. On a topological point of view, keeping in mind that recrystallization process is a 3D process and that 2D views are only partial explanations, recrystallized Cube-oriented grains are frequently located adjacent to or between deformed S-oriented regions, as depicted Figure 5-24. This suggests the possibility for the preferential migration of Cube boundaries into the surrounding S-deformed rich matrix, as a $40^\circ\langle 111 \rangle$ orientation relationship exist between Cube and S which can provide a mobility advantage to boundaries separating these orientations.

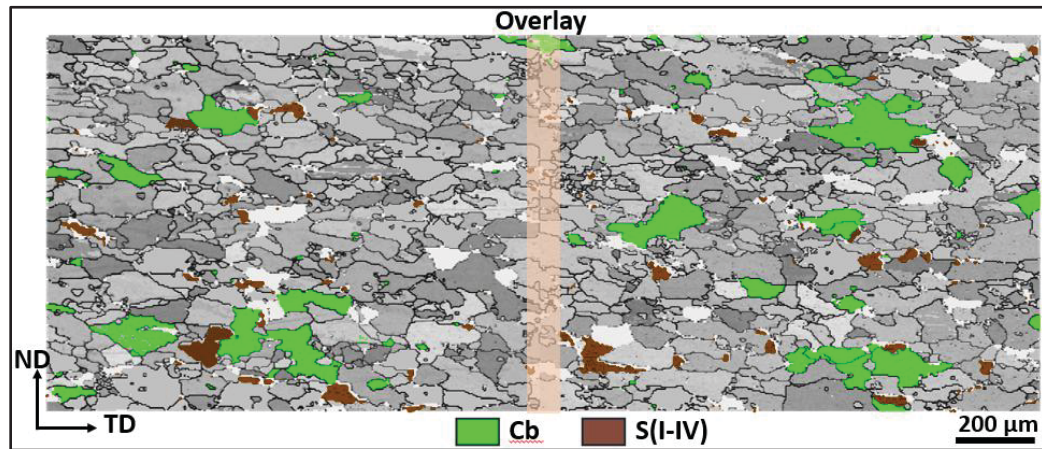


Figure 5-24 Zoom of the microstructure of Texture-component map of the annealed rolled as-cast AA6082 specimen at annealed 400 °C for 30 min, showing texture-component map the spatial distribution of recrystallized Cube (Cb, green) and deformed S(I–IV) (brown) orientations

The evolution of the non-recrystallized texture indicates a selective persistence of the β -fibre components during annealing. In the rolled as-cast material, Brass showed the greatest resistance among the deformation components. After both 5 and 10 min, Brass remains prominent within the non-recrystallized subset, while S and Cu are preferentially replaced by recrystallized grains (Figure 5-13). This behavior is consistent with the expected low stored energy associated with Brass and its lower tendency to participate in recrystallization relative to other β -fibre orientations (Li et al., 2018b). The high RI reported for the Brass-rich regions show that they resist recrystallization (Figure 5-17), while their high resistance score shows

that their internal substructure does not change significantly after the onset of recrystallization in the material (Figure 5-18).

5.2.5.2 Recrystallization of rolled as-extruded material

I. Material Heterogeneities

The prolonged annealing is required to increase the recrystallization percentage in the rolled extruded material. The heterogeneous banded structure retaining from the extrusion process led to a heterogeneous recrystallization process. Rolling perpendicular to the extrusion direction has modified the stability of the orientations generated by severe plastic deformation process and some regions have different boundary arrangements and deformation substructures than others, providing varied conditions for the onset of recrystallization. This is clearly seen when annealing is performed at 350 °C as some regions are fully recrystallized when some only present highly recovered structures (Figure 5-5f-h).

Once recrystallization is initiated within favorable regions, boundary migration proceeds readily in these regions; however, its progression seems to stop near neighboring bands that remain resistant to recrystallization. These regions may have lower stored energy or stronger particle-induced boundary pinning, limiting grain-boundary migration. Such local heterogeneity between adjacent bands may therefore account for the incomplete recrystallization at 350 °C and the slower progression observed at 400 °C.

At 400 °C, some regions remain non-recrystallized for considerably longer annealing times than others. This indicates that the incubation time for recrystallization differs between adjacent regions. It is relevant to note that the time exponent associated with the first stage of recrystallization is rather high at values around 5 (Figure 5-7). Slopes higher than 4 can be associated with an increasing nucleation rate with time, i.e., new nuclei keep forming throughout the entire recrystallization process, rather than all appearing at the exact start, corroborating the present observed behavior (Sohrabi & Gholamipour, 2021). The sharp

increase in recrystallized fraction from 41% to 87% between 800 and 1600 min annealing at 400 °C indicates that recrystallization nuclei can eventually form and grow within the regions that had remained resistant at shorter annealing times (Figure 5-7). The incubation stage is significantly longer in this region and seems to be triggered by a different mechanism, as suggested by the large deviation from the linear behavior reported for lower exposure time in Figure 5-7. At 350 °C, the recrystallization kinetic of the rolled as-extruded material is associated with a higher slope at 5.73, resulting in a predominantly non-recrystallized microstructure, even after 1600 min (Figure 5-15). The recrystallized fraction reached only approximately 26%, although a large HAGB fraction is available (about 46%). The thermal conditions at 300 °C allow the formation of HAGBs, but prevent their movement, even though the stored energy conditions are a priori sufficient to trigger the grain boundary movement as recovery is less active at this temperature than at 400 °C.

II. Texture-component stability during annealing

The evolution of the texture during annealing changes significantly between the rolled as-cast and the rolled as-extruded material, with a new orientation to consider (RG) and the need for longer annealing times to see any significant changes. The deformed as-extruded material can generate, stabilize and prevent the recrystallization of the RG orientation. At the shorter annealing time of 60 min at 400 °C, RG shows an RI of 1.00, indicating that it is fully retained. Its high RS of 0.94 also shows that the RG regions maintain relatively high intragranular misorientation and orientation gradients compare to the other NR components. In comparison, S is substantially retained, with an RI of 0.67, but its lower RS of 0.38 indicates considerably lower intragranular misorientation and orientation gradients in the surviving S regions. Brass shows lower retention, with an RI of 0.50, together with an intermediate RS of 0.54. The results indicate that high (RG) and low (S) intragranular orientation gradients may hinder the conditions necessary for recrystallization to take place. When the annealing time is increased to 800 min, the RI and RS values of RG do not change significantly, but the RI values of the Brass and S decrease only by 14 and 16 %, respectively (Figure 5-17 and Figure 5-18); while the recrystallization percent does not change significantly, going from 38% to 41%.

Figure 5-17A zoom into the microstructure after 800 min at 400 °C shown in Figure 5-25 provides insight into the microstructure configuration differences between the RG and Brass. In the texture-component map (Figure 5-25a), RG (orange) is preferentially retained within a largely non-recrystallized band, whereas several Brass-oriented regions (red) are located adjacent to recrystallized regions (black). The corresponding band-contrast map (Figure 5-25b) shows a denser network of LAGBs in the Brass-rich regions than in the RG-rich regions. Looking at the GND-density in the microstructure, one can quantify the localized lattice curvature associated with geometrically necessary dislocations and the associated map in Figure 26c and Figure 26d reveals that the distribution of local lattice curvature is different from one region to the next. The mean GND density is approximately 25% larger in the Brass-oriented regions than in the RG-oriented regions, showing that a significant difference in localized lattice curvature between the two orientations. Some RG grains have regions with very low GND density, around $4 \times 10^{13} \text{ m}^{-2}$; but these regions do not qualify as recrystallized according to the proposed criteria in section 5.2.3.5, as some open substructures which do not form subgrains are present (see red arrows Figure 5-25 b and c).

The difference in stored energy and recrystallization behavior between the RG and the Brass can be documented as early as the deformation state of the material, i.e., before annealing. In the as-rolled 60% condition, RG already exhibits the highest RS with a lower score than after annealing with a value of 0.84, closely followed by Brass at 0.64 and 0.32 for S. RG begins annealing with a relatively continuous substructure with largest grain-scale orientation spread but not much local accumulation of deformation. The deformation within RG therefore appears less strongly organized into subgrains preventing both, the formation of moving boundaries and the availability of stored energy for grain boundary migration. Even after long annealing time, no favorable sites for the development of recrystallization nuclei may be found and thereby increasing the incubation time for these regions. Even after 1600 min at 400 °C, RG maintains a high retention level, indicating that it is not readily consumed by neighboring recrystallized grains. Its prolonged retention can therefore be associated with both limited development of favorable nucleation sites within RG and restricted boundary migration into RG-rich regions.

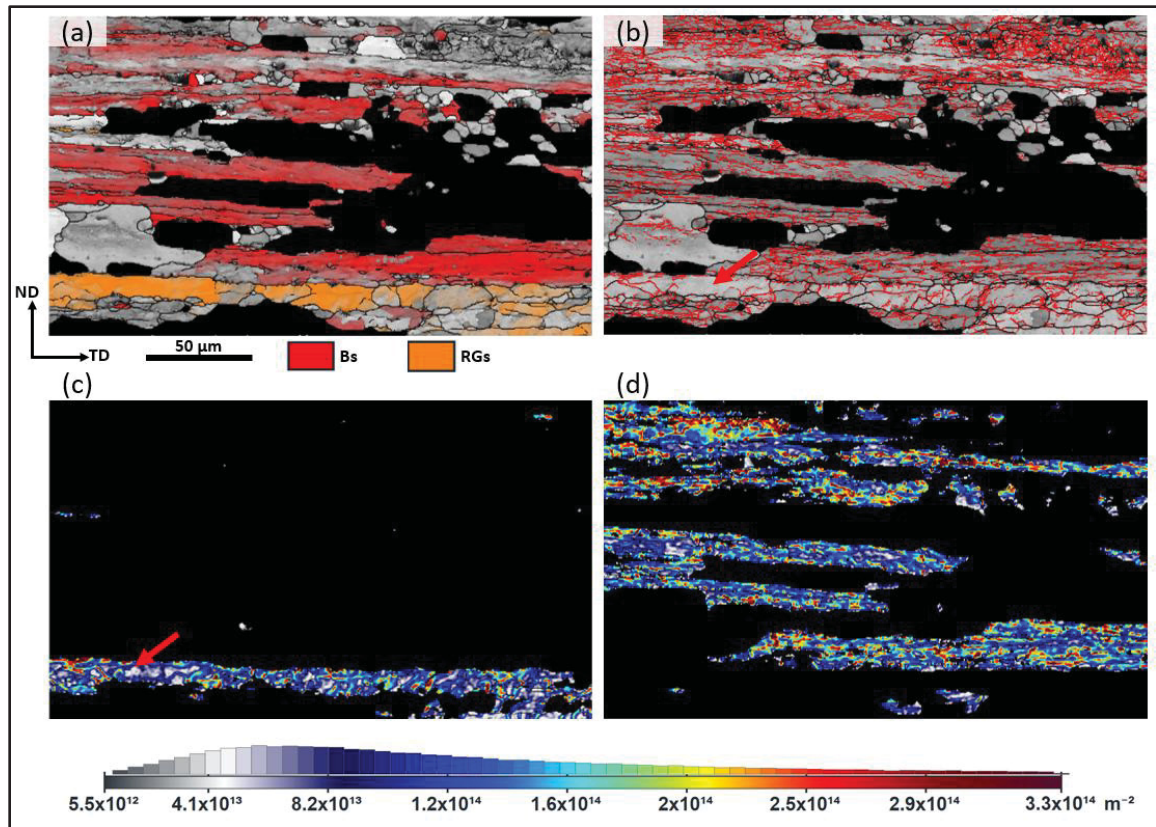


Figure 5-25 Zoomed view of the rolled as-extruded AA6082 microstructure annealed at 400 °C for 800 min: (a) texture-component map showing Brass (Bs, red) and Rotated Goss (RG, orange) orientations within the non-recrystallized regions; (b) corresponding band-contrast map showing LAGBs of 2°–15° (red) and HAGBs >15° (black); and orientation-filtered GND-density maps of the RG-oriented and Brass-oriented non-recrystallized regions in (c) and (d), respectively. Black regions correspond to recrystallized regions and red arrows indicate low-GND RG regions with open LAGB substructures

At 350 °C after 1600 min, the same behaviors are observed: RG and S remain with RI values close to 1, while Brass has a lower RI at 0.75 (Figure 5-17) and the RS values is high for RG at 0.96, medium for Brass at 0.58, and low for S at 0.35 (Figure 5-18). Even at lower temperature, S is able to recover rapidly but its low internal misorientation content present nucleation to take place. On the other hand, the Brass-oriented regions maintain a relatively high stored energy enabling nucleation and boundary migration; however, the limited recrystallized fraction after 1600 min, despite the presence of a substantial HAGB fraction (Figure 5-15), supports this limited grain mobility at this temperature. The larger amount of Mg_2Si at 350 °C compared with 400 °C predicted in Figure 5-19, may contribute to the restrict

boundary motion through particle-induced pinning and contribute to the confinement of the recrystallized region to small regions, helping the retention of the deformation texture.

The as-extruded rolled material requires longer annealing time to achieve complete recrystallization compared with the rolled as-cast material that is used here as a reference. One possible explanation is the persistence of extrusion-derived bands, within which the incubation time for recrystallization may vary significantly from one band to another, thereby limiting the spatial extent of nucleation. Furthermore, the presence of numerous small secondary particles may prevent boundary migration from neighboring recrystallized regions is restricted.

III. Influence of second phases on Recrystallization and Grain Boundary Migration

The heterogeneous particle distribution of the secondary particles may contribute to the coexistence of recrystallized and persistent non-recrystallized bands in the rolled as-extruded material. Only limited Mg-Si particles are observed as they are not stable in the range considered in the extrusion process (approximately 520-540 °C) and Fe-Mn-rich particles exhibit a banded distribution in some regions of the material. Coarse second-phase particles inherited from the extrusion condition and retained during subsequent rolling can locally change the imposed deformation, producing regions with enhanced dislocation density and orientation gradients that provide favorable conditions for PSN. Nucleation within these locally heterogeneous deformation zones can generate a broad range of recrystallized orientations, thereby contributing to the weak recrystallization texture if no preferential growth is happening.

At 400 °C, the compositions of the stable phases differ from the ones present at the extrusion temperature and some dissolution and precipitation process are expected to happen, but they may take some time to take place. Approximately 30% of the material is already recrystallized after 5 min, indicating that favorable nucleation sites are activated early in the process. The recrystallized population is weakly textured, suggesting that PSN is the nucleation mechanism and the particles present during the deformation process could be at the origin of the nucleation process. However, some regions do not undergo static recrystallization despite the presence of

large particles, even after 800 min of annealing. Coarse particles located close to recrystallized grain boundaries are seen in Figure 5-21b, but similar particles are not acting as a nucleation site in other regions of the material. The analyzed particle population has a mean equivalent diameter of approximately $3.8 \pm 1.8 \mu\text{m}$, which is well above the $\sim 1 \mu\text{m}$ size generally associated with PSN, indicating that a substantial fraction of these coarse particles is large enough to promote particle-stimulated nucleation. The precipitation of new particles may consume part of the stored energy in the alloy, particularly since precipitation can preferentially nucleate on dislocation structures. The formation of particles with new compositions may consequently reduce the local stored energy in regions not yet recrystallized, and this reduction would decrease the local driving force for recrystallization, potentially inhibiting or delaying recrystallization in these regions.

The substantial increase in recrystallization after prolonged annealing at 400 °C (87% after 1600 min) could be associated with the sudden decrease of the particle-related constraint on boundary migration, both as Zener-pinning effect and the reduction of the anisotropic distribution of Fe-Mn rich particles. Between the as-rolled condition and 1600 min, the relative Zener-pinning parameter, defined as the ratio between their volume fraction and the particle radius, decreases by approximately 24%, indicating a lower particle-pinning contribution. At the same time, the degree of directional alignment of the particle distribution, quantified by the mean resultant length decreases, goes from $R = 0.50$ to $R = 0.30$, showing that the initially TD-aligned particle population becomes less directionally concentrated within the bands. The lower pinning contribution, together with the reduced directionality of the particle, provides less restriction to migrating recrystallization boundaries and is consistent with the rapid increase in recrystallized fraction after a long low progression.

The stronger retardation of the recrystallization observed at 350 °C may also be associated with a greater particle-pinning contribution from Mg_2Si . Between 400 and 350 °C, only the percentage of stable Mg_2Si slightly increase, however, the dissolution and precipitation kinetics can be significantly lower at 350 °C. A higher Mg_2Si fraction may increase the Zener-pinning contribution and thereby further restrict grain-boundary migration at the lower temperature; however, the effectiveness of this pinning also depends on particle size and spatial

distribution. The present BSE analysis mainly focuses on particles larger than approximately 1 μm , which are relevant to the observed texture evolution through their possible contribution to PSN, but does not characterize the finer particle population. These finer particles may contribute more strongly to Zener pinning and therefore to the retardation of grain-boundary migration. Consequently, a more complete assessment of the particle-induced retardation would require the characterization of the fine Mg_2Si particle distribution at both 350 and 400 $^{\circ}\text{C}$.

5.2.6 Conclusions

This study investigated the effect of the initial microstructure and texture on the recrystallization behavior and texture evolution of 60% cold-rolled AA6082. Rolled as-cast and as-extruded materials were annealed under various time–temperature conditions and characterized using EBSD to distinguish recrystallized / non-recrystallized regions, evaluate grain structure, and document crystallographic texture. The persistence and stability of deformation-texture components were further assessed using the proposed Retention Index (RI) and Resistance Score (RS), while second-phase particle characterization and thermodynamic calculations were used to examine their potential contribution to the recrystallization behavior.

- The rolled as-cast and rolled as-extruded materials have very different recrystallization behavior. The as-cast material undergoes rapid consumption of the deformed matrix at 400 $^{\circ}\text{C}$ as expected from the literature, but the as-extruded material has much slower recrystallization kinetics and retains extrusion-derived bands that remain resistant over prolonged annealing. It is only after a long annealing time of 1600 min that recrystallization can process to almost completion, suggesting that a complex interaction between various mechanisms is taking place.
- The resistance to recrystallization depends strongly on the thermomechanical history of the initial material. Brass component is the most resistant component in the rolled

as-cast material, while it is the less stable orientation in the rolled as-extruded material whereas Rotated Goss exhibits the greatest stability in that condition. The persistence of Rotated Goss is associated with large gradients of orientation changes but a subgrain-boundary with lower localized lattice curvature. This seems to limit favorable nucleation conditions and the subsequent consumption of these regions of the material.

- The heterogeneous particle distribution in the rolled as-extruded material contributes to the spatially heterogeneous recrystallization. Coarse particles can promote PSN and contribute to the weak recrystallization texture, while particles concentrated at the border of the extrusion-derived bands can restrict recrystallization-boundary migration.
- The contribution of the secondary particles to the recrystallization process, both on the nucleation step and the boundary migration has been discussed but may require further analyses. At 400 °C, a reduction in particle pinning and in the directionality along the deformation banded structure is document as the annealing time increases, which is consistent with the eventual progression of recrystallization. At 350 °C, the higher predicted Mg₂Si fraction may increase the particle-pinning contribution and reinforce the retardation of boundary migration preventing the progression of the recrystallization front.

Declaration of generative AI and AI-assisted technologies

During the preparation of this work, the authors used ChatGPT (OpenAI) to improve the language, clarity, and readability of the manuscript. The authors subsequently reviewed and edited the content as needed and take full responsibility for the content of the publication.

Data availability

The data supporting the findings of this study are available from the corresponding author upon reasonable request.

Acknowledgments

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5.3 Annealing of Rolled As-extruded AA6082 at 520 °C

5.3.1 Results of Annealing at 520 °C

5.3.1.1 Annealing of Rolled As-extruded AA6082 at 60% Reduction

I. Microstructural and Texture Evolution

The rolled as-extruded AA6082 at 60% reduction under UCR (4.3.1.2, Figure 4-16f and Figure 4-17f) is characterized in the TD–ND section by a strongly elongated grain structure aligned along TD, resulting from rolling perpendicular to the extrusion direction. The microstructure consists of continuous deformation bands, within which a dense network of LAGBs is distributed, while HAGBs appear as discontinuous and partially connected boundaries. The EBSD map shows pronounced orientation continuity along TD, with limited grain subdivision and a clear banded structure.

During the early stages of annealing at 520 °C, the microstructure visually retains much of its deformed character. At 1 s (Figure 5-26a and Figure 5-27a), the elongated morphology is largely preserved, although the band contrast improves slightly. The boundary network remains dominated by LAGBs aligned with the deformation bands, while HAGBs are still not fully continuous. However, quantitative analysis (Figure 5-28) indicates that transformation

has already initiated, with the HAGBs fraction increasing to 44% and the recrystallized area fraction (REX%) reaching approximately 21%. At 2 s (Figure 5-26b and Figure 5-27b), the internal substructure appears less dense, consistent with a further increase in HAGBs fraction to 57% and REX% approaching 49%. The EBSD map shows the emergence of some regions with orientation deviation from the deformation bands. By 4 s (Figure 5-26c and Figure 5-27c), the reduction of LAGBs becomes more pronounced, corresponding to a HAGBs fraction of 68% and a REX% of approximately 64%. The elongated morphology is still partially preserved, while the EBSD map reveals a more heterogeneous orientation distribution with the appearance of new small grains within the deformed matrix.

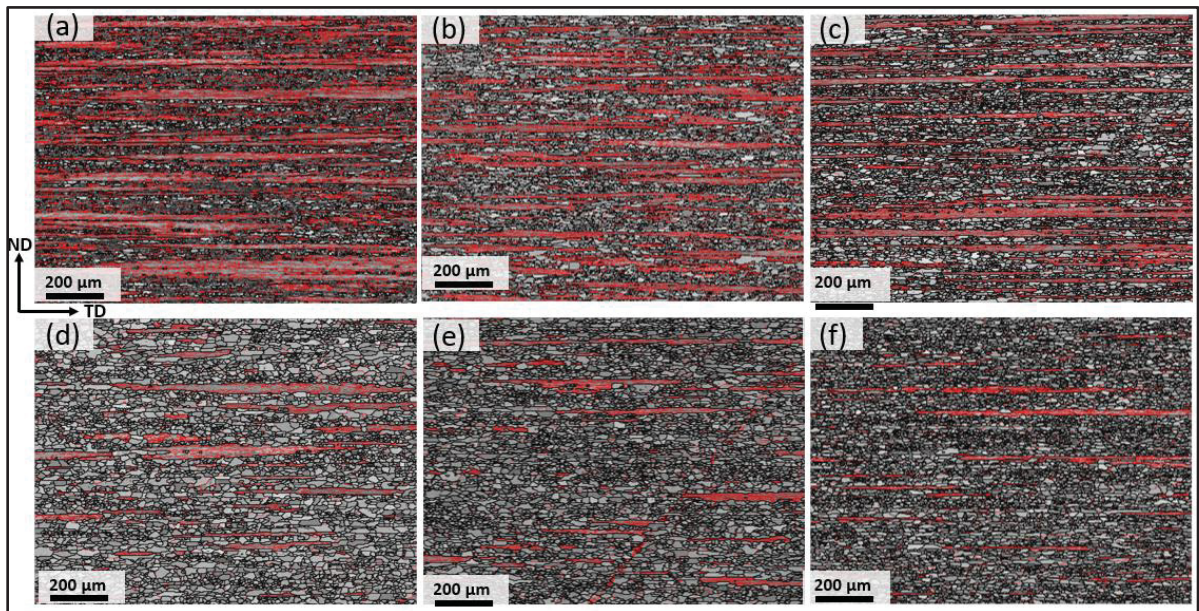


Figure 5-26 Band contrast maps of rolled as-extruded AA6082 at 60% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

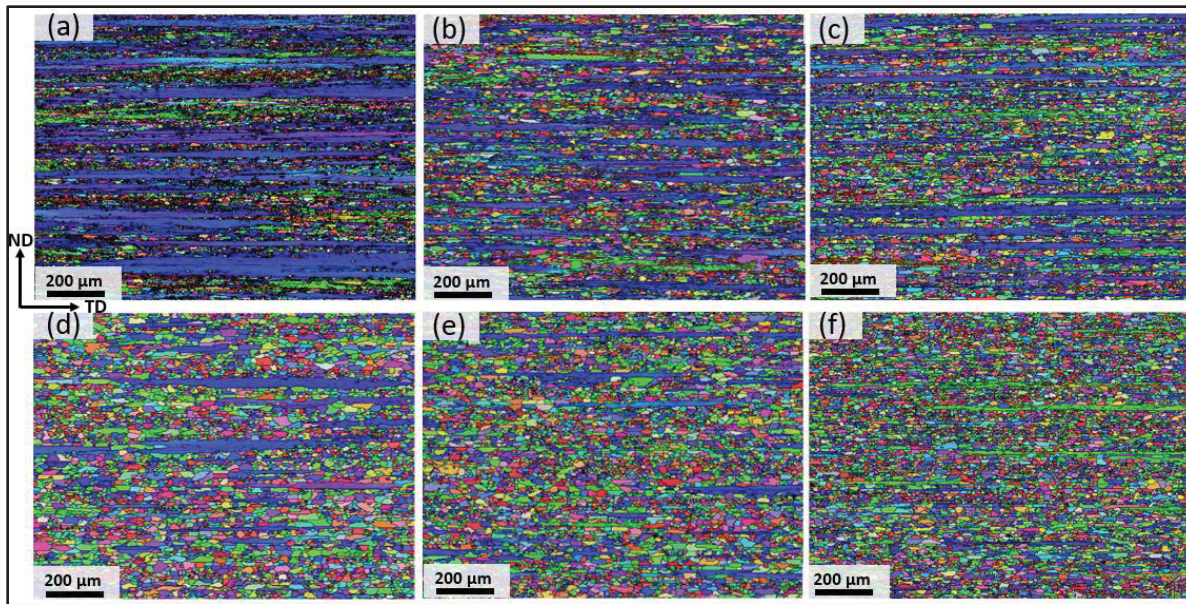


Figure 5-27 EBSD orientation map of rolled as-extruded AA6082 at 60% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

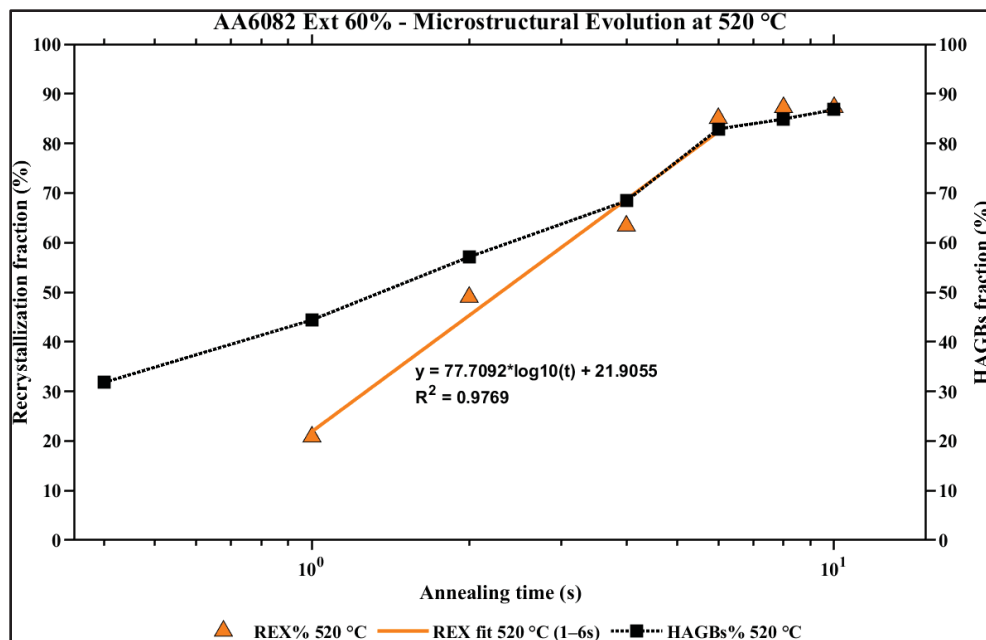


Figure 5-28 Microstructural evolution of AA6082 extruded and 60% cold-rolled specimens during annealing at 520 °C. The plot shows the recrystallized area fraction (REX%) on the primary y-axis and the high-angle grain boundary fraction (HAGB%) on the secondary y-axis as a function of annealing time. A linear regression fit for the recrystallization kinetics is provided for the 1–6 s interval

At longer durations, a clear microstructural transformation is observed. At 6 s (Figure 5-26d and Figure 5-27d), the microstructure exhibits a mixed morphology composed of remaining elongated regions and a significant fraction of newly formed grains. The boundary network becomes increasingly dominated by HAGBs, while LAGBs are further reduced, consistent with the HAGBs fraction reaching 83% and the REX% rising sharply to about 85% (Figure 5-28). The EBSD map confirms this transition through a broader orientation distribution. At 8 s (Figure 5-26e and Figure 5-27e), the microstructure becomes predominantly equiaxed, with a boundary network mainly defined by HAGBs and only a limited presence of LAGBs; the HAGBs fraction reaches 85% and the REX% approximately 87%. At 10 s (Figure 5-26f and Figure 5-27f), the microstructure becomes more homogeneous, and the elongated deformation bands are no longer comparable to those observed at shorter durations such as 4 s. The boundary network is dominated by continuous HAGBs, with minimal LAGBs, while the HAGBs fraction reaches 87% and the REX% remains at approximately 87%.

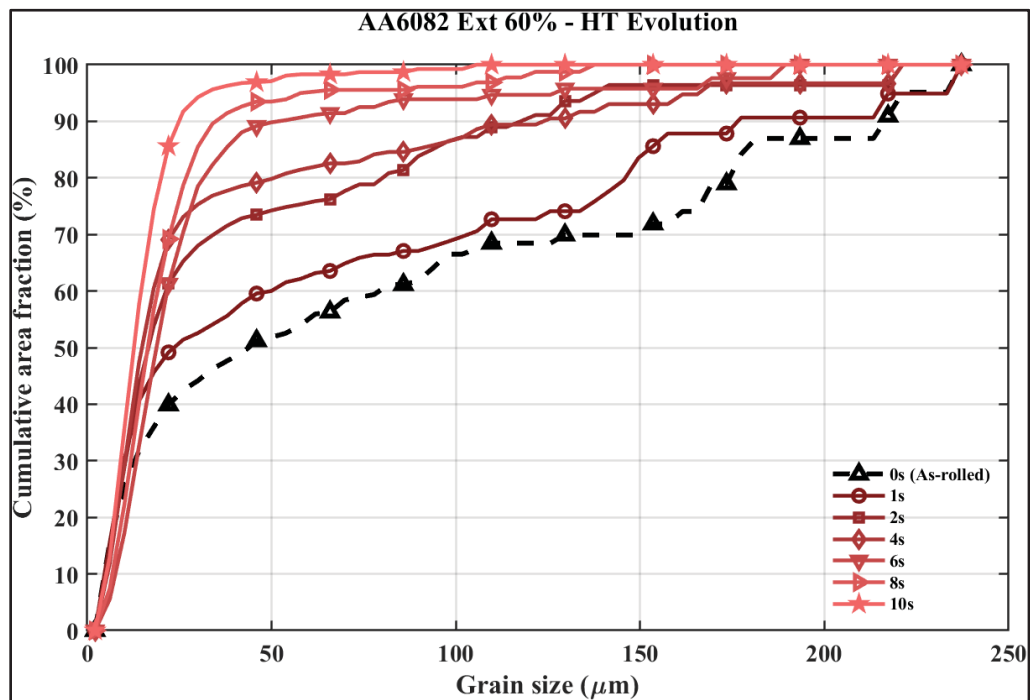


Figure 5-29 Evolution of grain size distribution for rolled as-extruded AA6082 (60% UCR) during heat treatment at 520 °C. The cumulative area-fraction curves are shown for the as-rolled condition (0 s) and annealing durations of 1 s, 2 s, 4 s, 6 s, 8 s, and 10 s

The grain size analysis further confirms the progressive refinement of the rolled as-extruded AA6082 (60% UCR) during heat treatment at 520 °C (Figure 5-29). As shown by the cumulative area-fraction curves, the distributions shift steadily toward smaller grain sizes with increasing annealing time and become progressively steeper, indicating the development of a finer and more homogeneous microstructure. In the as-rolled condition, the grain structure is still relatively coarse, with a maximum grain size of about 239 μm and an 85% cumulative grain size of about 177 μm . During the first 4 s of annealing, the maximum grain size changes only slightly and remains above 220 μm , whereas the 85% cumulative grain size decreases much more markedly to about 89 μm , showing that refinement has already advanced in a large fraction of the scanned area. A more pronounced transition occurs between 4 s and 6 s, where the cumulative curves shift sharply to the left; the maximum grain size decreases more noticeably to about 190 μm , while the 85% grain size drops from about 89 μm to about 39 μm . With further annealing, the refinement continues, and by 10 s the microstructure reaches a much finer state, with a maximum grain size of about 106 μm and an 85% cumulative grain size of about 24 μm . Thereby, these results indicate that the most significant grain refinement occurs after the first few seconds of annealing, consistent with the replacement of the coarse deformed structure by a finer recrystallized grain population.

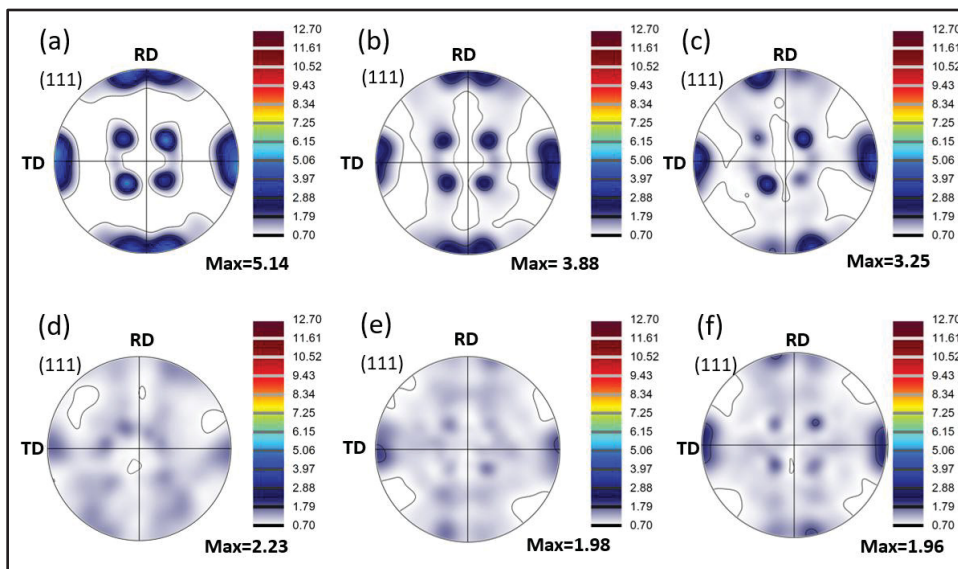


Figure 5-30 $\{111\}$ pole figures of rolled as-extruded AA6082 at 60% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

The evolution of crystallographic texture during heat treatment at 520 °C is illustrated by the {111} pole figures in Figure 5-30. Relative to the rolled as-extruded 60% condition (Figure 4-19f), which exhibits a maximum intensity of 7.9, the annealed sample at 1 s already shows a lower intensity (Max = 5.1), indicating an early weakening of the deformation texture developed during UCR. At 2 s and 4 s, the maximum intensity decreases further to 3.9 and 3.2, respectively, while the pole distribution becomes broader and less sharply defined, reflecting a progressive reduction in texture strength and a redistribution of orientations. A more pronounced change occurs at 6 s, where the maximum intensity drops to 2.2 and the pole figure becomes markedly diffuse. From 6 s to 10 s, only a faint ghost of the initial rolling texture remains. At 8 s and 10 s, the maximum intensities further decrease to around 2, and the pole figures approach a nearly random texture, with no dominant components that can be clearly distinguished. This evolution shows that heat treatment progressively suppresses the rolling-induced texture and leads to a weak, nearly random texture state.

The evolution of texture during heat treatment at 520 °C can be described using the volume fractions of the main FCC texture components shown in Figure 5-31. As discussed previously, the as-extruded AA6082 is initially characterized by strong extrusion-related components with RCu and Bs fractions of approximately 35% and 31%, respectively, together with about 16% Cb and 5% RGs. After 60% UCR, the texture becomes strongly dominated by Bs, which increases to approximately 63%, with smaller contributions from S (11%), Cb (5%), and RGs (4%), while RCu decreases to approximately 1%.

During annealing, the most notable change is the progressive weakening of the deformation-related Bs component. At 1 s, Bs remains the dominant component at approximately 43%, while S accounts for approximately 9% and RGs for 6%. By 2 s, Bs decreases markedly to approximately 21%, whereas S remains at about 9% and RCu increases to approximately 5%. At 4 s, Bs remains at a similar level of approximately 22%, while S decreases slightly to 8% and RCu increases to approximately 6%, indicating a redistribution of the texture components.

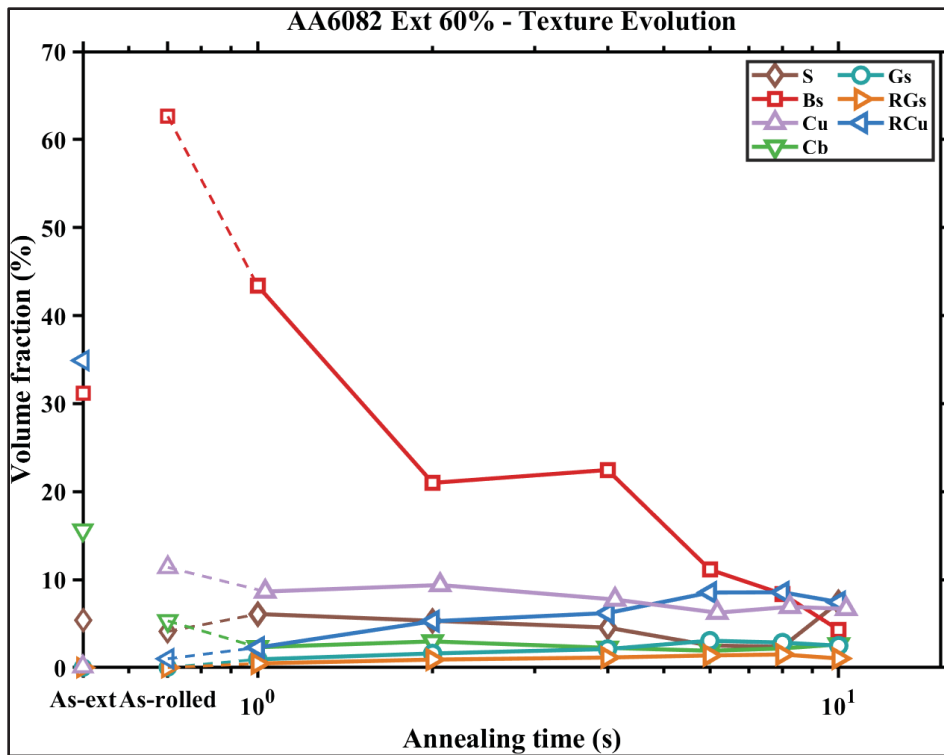


Figure 5-31 Volume fractions of the main FCC texture components for rolled as-extruded AA6082 (60% UCR) during heat treatment at 520 °C, including the as-extruded, as-rolled, and annealed conditions from 1 s to 10 s

This trend becomes more pronounced at longer annealing times. At 6 s, Bs decreases further to approximately 11%, while RCu reaches approximately 9%. At 8 s, Bs decreases to about 8%, while RCu remains at approximately 9% and S is about 7%. By 10 s, Bs is reduced to approximately 4%, whereas RGs, S, and RCu each reach approximately 7%. In comparison, Cu, Cb, and Gs remain relatively low throughout the heat treatment, generally ranging between approximately 1% and 3%. By 10 s, no single texture component strongly dominates the texture, consistent with the pronounced weakening of the initial rolling texture and the development of a weak, more broadly distributed crystallographic texture.

II. Stability of the Non-Recrystallized Fraction at 520 °C Evolution of the NR Subset

The annealing conditions at 2 s and 6 s were selected for comparative analysis of texture stability because they capture a major transition in the recrystallization fraction, which increases from 49% to 85%, respectively. By separating the non-recrystallized (NR) subset from the EBSD data and weighting its texture fractions by the corresponding NR fraction, the microstructural and crystallographic characteristics of the remaining deformation features can be followed through this critical stage of transformation (Figure 5-32 and Figure 5-33).

At 2 s of annealing (Figure 5-32a, c and Figure 5-33a, c), the NR subset is still mainly composed of elongated deformation bands aligned along TD. The corresponding $\{111\}$ pole figure shows a maximum intensity of 5.6 (Figure 5-32c). Quantitative volume fraction analysis (Figure 5-33a) indicates that the Bs component is the dominant component in the initial texture and also makes the largest contribution from the NR regions. Its NR-weighted volume fraction is approximately 23%, compared with approximately 63% in the initial texture. Other deformation-related components, such as S and RG, show NR-weighted fractions of approximately 7% and 3%, respectively, compared with about 11% and 4% in the initial texture. The NR-weighted Cb fraction is also reduced to approximately 2%, compared with about 5% initially. In contrast, the RCu component remains limited to approximately 1% in both the NR-weighted and initial textures, while Cu and Gs remain negligible. The area-weighted grain size distribution for the NR subset at 2 s (Figure 5-33c) exhibits a broad range extending up to about 150 μm .

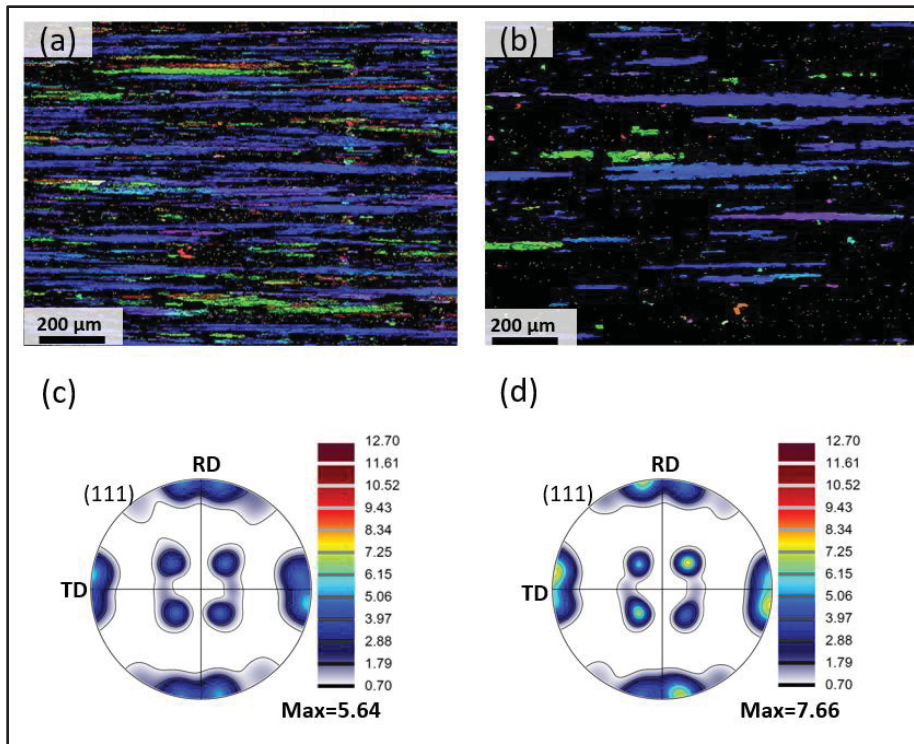


Figure 5-32 EBSD maps and corresponding $\{111\}$ pole figures of the non-recrystallized (NR) subset for rolled as-extruded AA6082 (60% UCR) after annealing at 520 °C: (a) EBSD map at 2 s, (b) EBSD map at 6 s, (c) $\{111\}$ pole figure at 2 s, and (d) $\{111\}$ pole figure at 6 s

By 6 s of annealing (Figure 5-32b, d and Figure 5-33b, d), the microstructure has reached a much more advanced stage of transformation (85% recrystallization), and the NR subset is reduced to a smaller number of elongated grains. Despite this strong reduction in NR area, the remaining regions exhibit a sharper crystallographic character, as reflected by the increase in the $\{111\}$ pole figure intensity to 7.7 (Figure 5-32d). This behavior is primarily associated with dominant contribution from the Bs component, which continues to provide the largest NR-weighted texture contribution with an NR-weighted fraction of approximately 9% (Figure 5-33b). At the same time, the NR-weighted fraction of the S component is approximately 1%, indicating that it is less persistent than Bs during recrystallization. The RCu component remains at a very low NR-weighted fraction of below 1%, while the Cu and Gs components remain close to zero. The corresponding grain size distribution (Figure 5-33d) reveals that the surviving NR remnants remain highly heterogeneous; although their area is strongly reduced, the remaining streaks still retain significant lengths along TD, with area-weighted sizes

reaching approximately 190 μm . This apparent increase in the maximum grain dimension compared with the 2 s condition is not an indication of grain growth, but rather a result of the heterogeneous nature of the structure. These results show that the surviving NR regions are characterized by a dominant Bs contribution and confirm that these persistent remnants correspond to highly stable deformation orientations that effectively resist the restoration process.

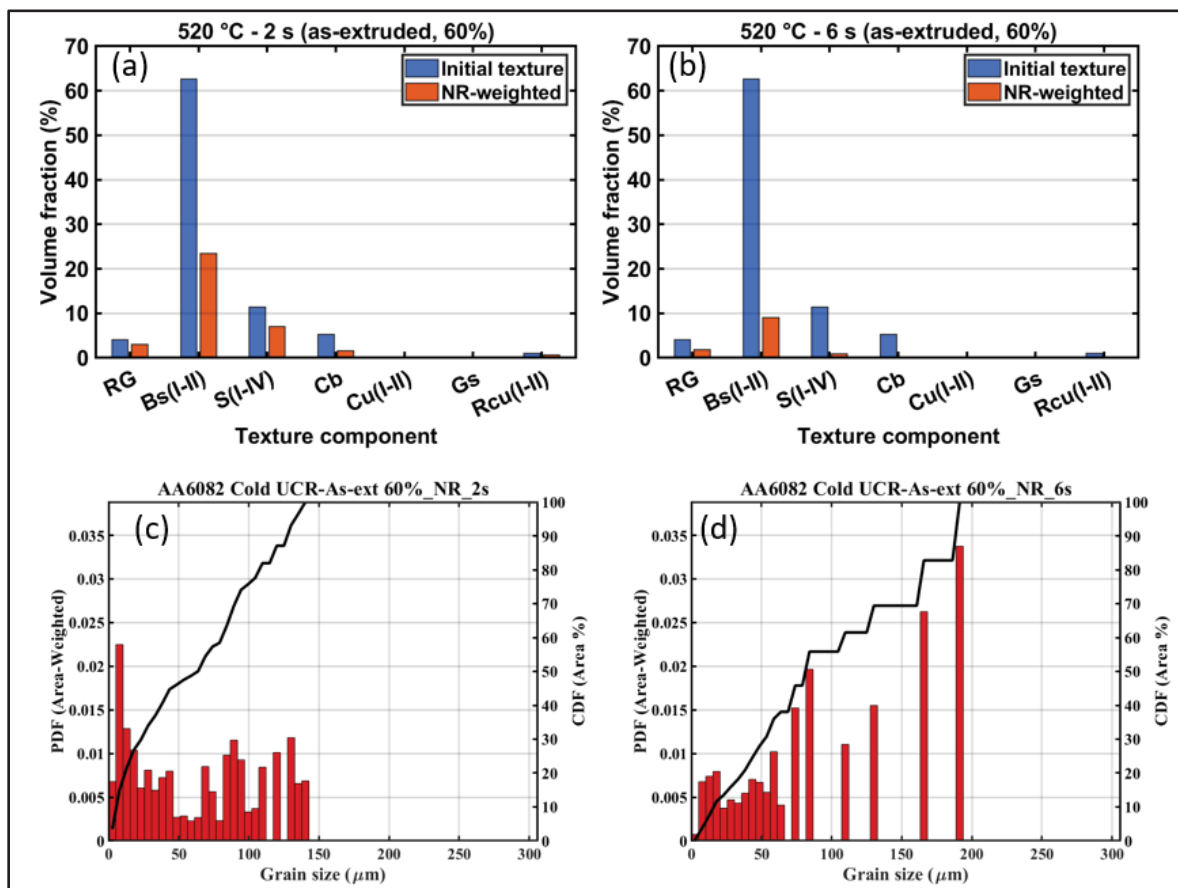


Figure 5-33 Texture component fractions (total vs. NR-weighted) and grain size distributions of the non-recrystallized (NR) regions in rolled as-extruded AA6082 (60% UCR) after annealing at 520 °C: texture component fractions after (a) 2 s and (b) 6 s; grain size distributions (PDF/CDF) of the NR regions after (c) 2 s and (d) 6 s

III. Texture Persistence and Substructural Stability

A quantitative assessment of texture persistence and substructural stability in AA6082 at 520 °C was carried out using the Retention Index (RI) and the resistance score (RS) over short annealing durations (1–10 s). While the RI describes the tendency of a given texture component to remain within the non-recrystallized (NR) fraction, the RS reflects the internal state of the surviving NR regions based on the normalized combination of GOS and GOS/D, thereby linking resistance to intragranular misorientation and its spatial distribution within the grain structure.

The RI map (Figure 5-34) indicates that, at the initial stage (1 s), the RGs and Bs(I–II) components exhibit the two highest retention levels, with RI values of 0.77 and 0.70, respectively, while S(I–IV) exhibits a slightly lower RI of 0.63. As annealing proceeds from 2 to 4 s, RGs remains the dominant retained orientation, with RI values of 0.74 and 0.80, respectively. Over the same interval, Bs(I–II) decreases from 0.37 to 0.33, whereas S(I–IV) decreases from 0.62 to 0.35. A pronounced decline in retention occurs after 4 s: at 6 s, the RI values fall to 0.44 for RGs, 0.14 for Bs(I–II), and 0.08 for S(I–IV). At 8 s, RGs remains the most strongly retained component with an RI of 0.43, while Bs(I–II) and S(I–IV) decrease further to 0.10 and 0.08, respectively. Other deformation components, such as Cu(I–II), are rapidly consumed and display negligible retention throughout the investigated time range. At 10 s, the same overall hierarchy persists, with RGs remaining the component with the highest RI, while Bs(I–II) and S(I–IV) decline to only 0.03 and 0.06, respectively. Thus, the RI map demonstrates that RGs is consistently the most persistent component and becomes increasingly dominant relative to Bs(I–II) and S(I–IV) as the heat-treatment time increases beyond 4 s.

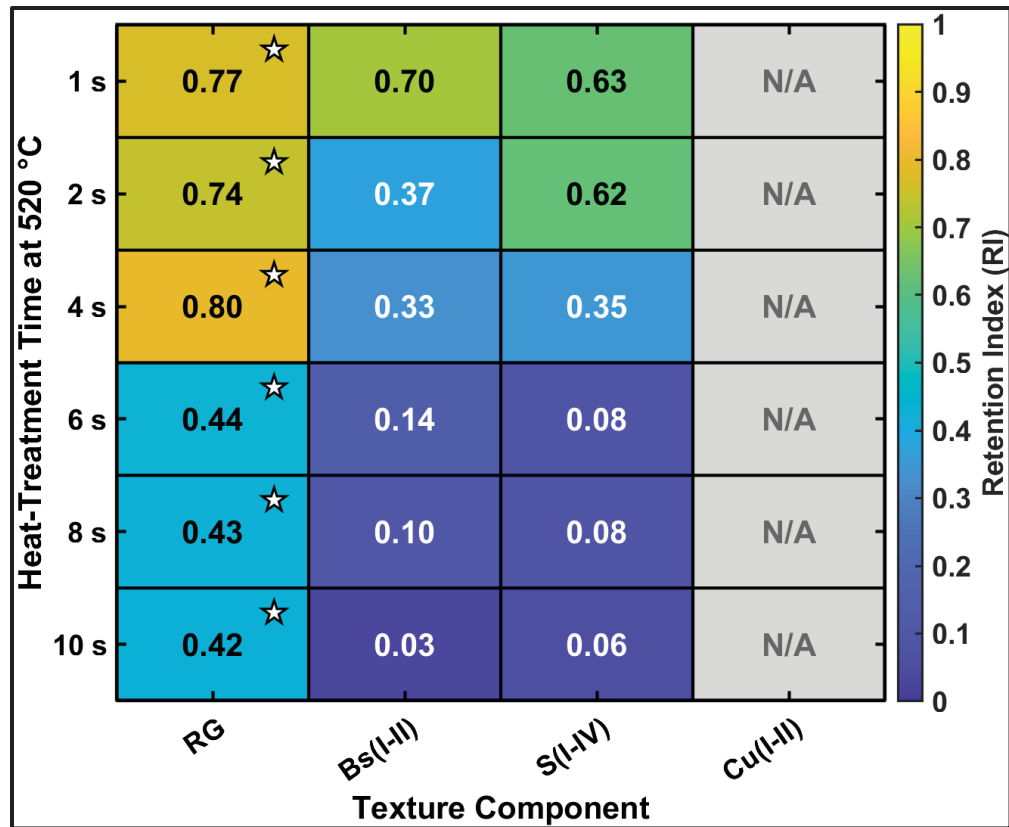


Figure 5-34 Retention Index (RI) for each texture component during heat treatment at 520 °C (1–10 s). ☆ marks the highest-RI component in each condition

The RS (Figure 5-35) reveals a complementary trend regarding the internal state of these orientations. Over the full annealing duration, the RGs component consistently exhibits the highest RS, ranging from 0.75 to 0.98. In contrast, Bs(I–II) and S(I–IV) display moderate resistance values, generally between 0.41 and 0.63, with a localized peak of 0.78 for Bs at 2 s.

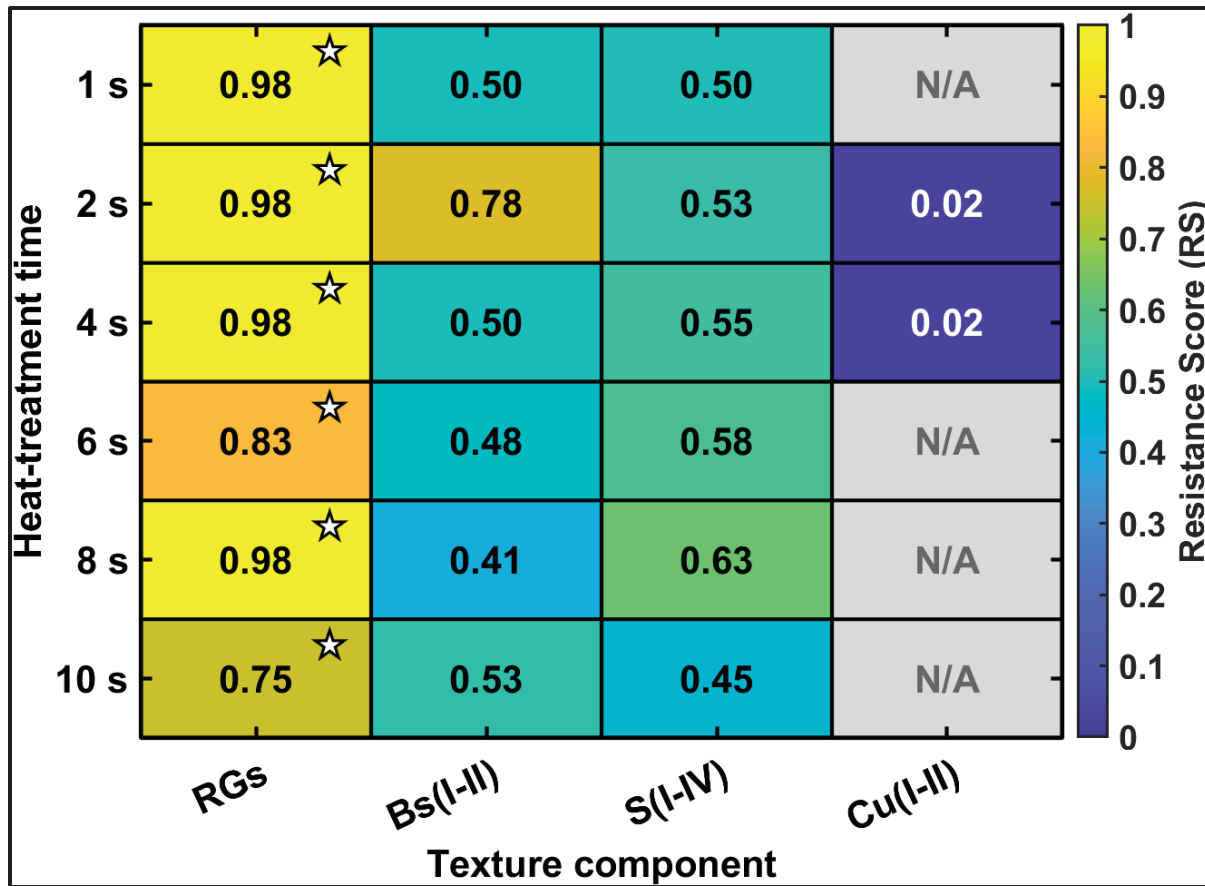


Figure 5-35 Resistance Score (RS), calculated from the equal-weight normalized combination of GOS and GOS/D, for each texture component during heat treatment at 520 °C (1–10 s). ☆ indicate all components exhibiting the maximum RS within each condition

5.3.1.2 Annealing of 90% Rolled As-extruded AA6082

The initial state of the rolled as-extruded AA6082 at 90% reduction under UCR was characterized by a highly deformed and banded structure, as detailed in Section 4.3.1.2 (Figure 4-20). In that as-rolled condition (0 s), the microstructure was defined by continuous deformation bands aligned along TD and a sharp rolling texture with a maximum intensity of 5.6.

The microstructural evolution of this sample during annealing at 520 °C is shown in Figure 5-36 and Figure 5-37. In contrast to lower reduction levels, the microstructure after only 1 s of

annealing already exhibits a fully transformed state, characterized by a uniform distribution of equiaxed grains. The band contrast maps (Figure 5-36) show a boundary network entirely dominated by black HAGBs, with only a minimal presence of red LAGBs. The EBSD orientation maps (Figure 5-37) demonstrate a complete loss of the prior elongated TD-aligned morphology, replaced by a distribution of diverse grain orientations. Quantitatively, the recrystallized area fraction reaches 93% at the 1 s and remains stable between 93% and 95% throughout the 10 s interval. Similarly, the HAGB fraction is measured at 95% at 1 s and reaches a maximum of 96% by 10 s.

The grain size distribution analysis (Figure 5-38) quantifies the immediate refinement occurring at this strain level. The as-rolled state (0s) is characterized by a maximum grain size of 72 μm and an 85% area fraction size of 43 μm . Following 1 s of annealing, these values drop significantly to 32 μm and 18 μm , respectively. Between 1 s and 10 s, the grain size remains nearly constant, with the cumulative area-fraction curves in Figure 5-38 almost entirely overlapping. At the final 10 s mark, the maximum grain size is 27 μm and the 85% cumulative size is 15 μm , indicating that the microstructure reaches a stabilized state nearly instantaneously upon immersion in the salt bath.

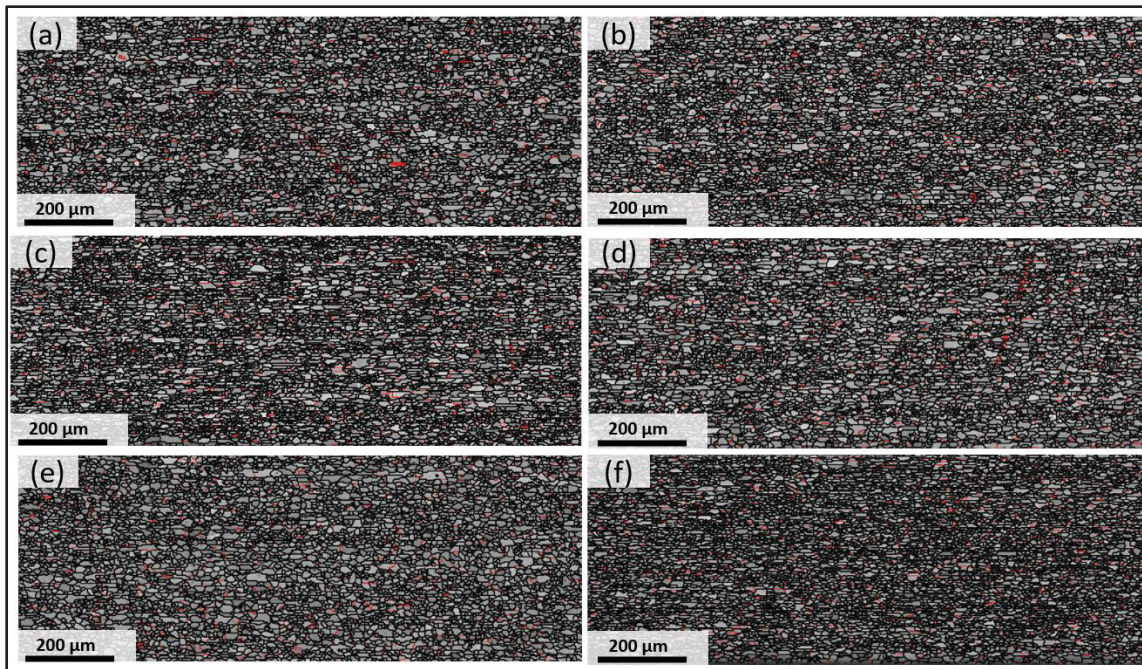


Figure 5-36 Band contrast maps of rolled as-extruded AA6082 at 90% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

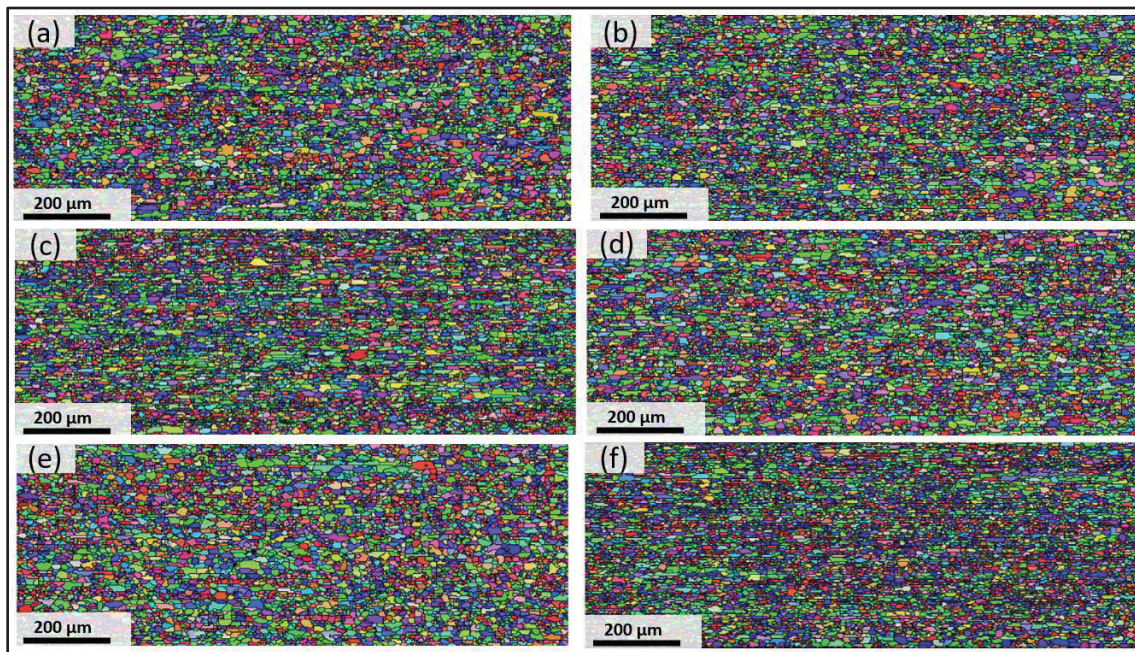


Figure 5-37 EBSD orientation map of rolled as-extruded AA6082 at 90% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

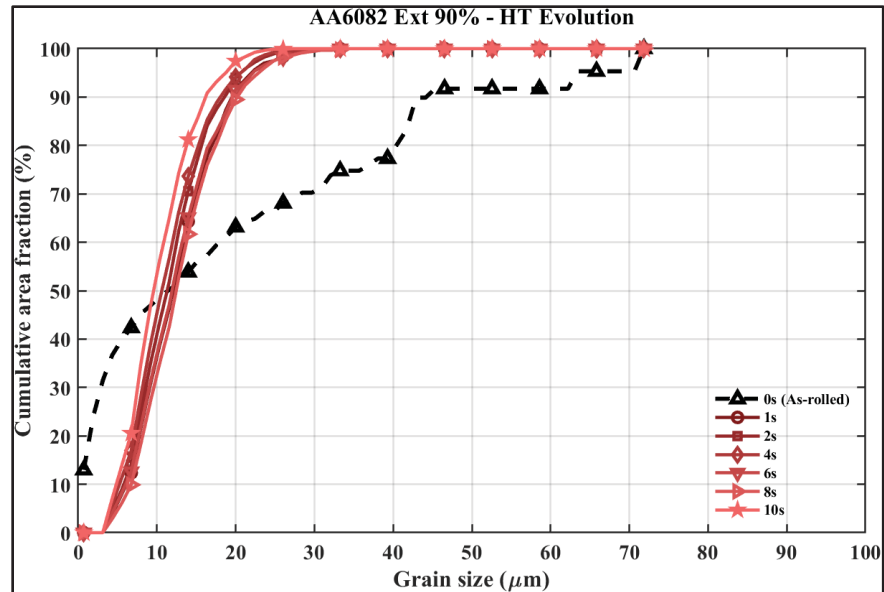


Figure 5-38 Evolution of grain size distribution for rolled as-extruded AA6082 (90% UCR) during heat treatment at 520 °C. The cumulative area-fraction curves are shown for the as-rolled condition (0 s) and annealing durations of 1 s, 2 s, 4 s, 6 s, 8 s, and 10 s

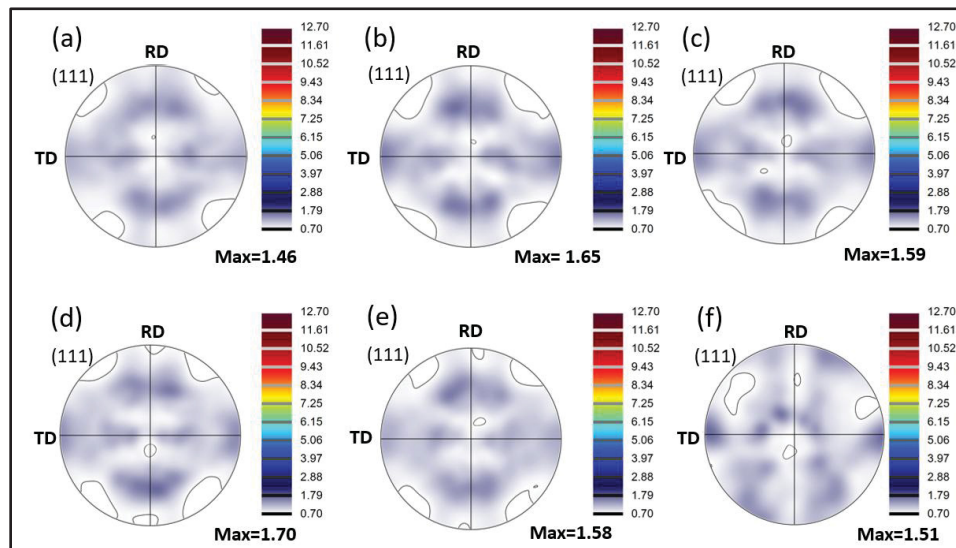


Figure 5-39 {111} pole figures of rolled as-extruded AA6082 at 90% reduction after annealing at 520 °C for (a) 1 s, (b) 2 s, (c) 4 s, (d) 6 s, (e) 8 s, and (f) 10 s

The crystallographic texture evolution is illustrated by the {111} pole figures in Figure 5-39. Within 1s of annealing, the texture already exhibits a maximum intensity of 1.5, representing

a nearly random orientation distribution. This low intensity is maintained across all annealing durations, with maximum values fluctuating between 1.5 and 1.7. The pole figures do not show any dominant peaks or preferred recrystallization fibers, indicating that the transformation results in a crystallographically isotropic grain population during the initial stage of annealing.

5.3.2 Discussion

5.3.2.1 Macro-Kinetics: Effects of Temperature and Stored Energy

The restoration kinetics of the 60% cold-rolled AA6082 alloy are strongly affected by processing conditions. In particular, the interaction between temperature and initial strain controls the available driving force and therefore the rate of microstructural transformation.

I. Temperature Effects and the Thermal Regime Shift at 60% UCR

The microstructural evolution of the 60% cold-rolled AA6082 alloy shows strong sensitivity to annealing temperature. This is consistent with thermally activated restoration processes, where recovery, nucleation, and boundary migration are expected to become faster as temperature increases (Raabe, 2014). Miszczyk et al. (2017) showed that recrystallization in deformed aluminum increases rapidly over a narrow annealing-temperature range, indicating a strong temperature dependence of the transformation (Miszczyk et al., 2017). The large difference between the minute-scale and second-scale transformations observed here therefore reflects a major increase in restoration kinetics with temperature.

At 350 °C (Section 5.2.3.3), restoration is very slow. Even after 1600 min of annealing, the recrystallized fraction reaches only 26%. The microstructure therefore remains predominantly non-recrystallized, with elongated deformation bands and recovered substructures still clearly preserved. This suggests that although HAGBs can form, their effective migration remains strongly restricted at this temperature. This interpretation is consistent with the observation that the 350 °C condition reaches only about 26% recrystallization despite a HAGB fraction

of approximately 46%. In addition, second-phase particles may further restrict boundary motion through particle/Zener pinning. Such pinning can limit subgrain and grain-boundary migration, while recovery simultaneously reduces the stored deformation energy available to sustain recrystallization (Baton, Geslin, & Moussa, 2021; Humphreys & Hatherly, 2004). In the present alloy, the greater predicted stability of Mg_2Si at 350 °C also supports a stronger particle-related restriction of migrating boundaries than at 400 °C

At 400 °C (Section 5.2.3.3), recrystallization becomes more evident, although it still develops slowly. The recrystallized fraction increases from 41% at 800 min to 87% after 1600 min. During this stage, the deformed structure is gradually replaced by a more recrystallized grains, accompanied by a marked reduction in internal misorientation. The intermediate temperature provides sufficient boundary mobility for recrystallization to proceed, but the process still requires considerable time to overcome microstructural obstacles.

The strongest kinetic change occurs at 520 °C, where the transformation time is reduced from thousands of minutes to seconds. A recrystallized fraction of 85%, which requires 1600 min at 400 °C, is achieved in only 6 s at 520 °C. This indicates a transition toward a faster restoration regime. At the higher temperature, increased boundary mobility allows recrystallization to proceed much more rapidly. At the same time, the high stored-energy state produced by 60% rolling reduction promotes particle-stimulated nucleation and growth around coarse second-phase particles. This interpretation is consistent with Adam and Field (2023), who showed that coarse particles are associated with local deformation heterogeneity and high-stored-energy regions that favor particle-stimulated recrystallization (K. Adam & Field, 2023). By 10 s, the microstructure is largely recrystallized, with the REX fraction stabilizing near 87%, a predominantly HAGB-controlled boundary network, a much finer grain-size distribution, and a weak crystallographic texture. The persistence of a small non-recrystallized fraction even under these high-temperature conditions further indicates that local orientation and microstructural heterogeneity continue to influence the final stages of recrystallization.

II. Impact of Rolling Reduction

A comparison of 60% and 90% reduction at 520 °C shows that stored deformation energy is the main factor controlling restoration kinetics and grain refinement. Increasing the reduction level increases the stored energy, which in turn accelerates recrystallization and produces a finer final structure.

At 60% reduction, the microstructure evolves progressively from a predominantly deformed state at 1 s ($\text{REX} \approx 21\%$) to an advanced recrystallized state by 6 s ($\text{REX} \approx 85\%$), reaching about 87% at 10 s. In contrast, the 90% reduction condition provides a much higher driving force and leads to almost instantaneous transformation upon immersion in the salt bath. At this strain level, the material is essentially fully recrystallized within 1 s, with a recrystallized fraction of approximately 93%.

This behavior is consistent with the stored deformation energy framework described by (Hansen & Juul Jensen, 2011). During plastic deformation, grains are progressively subdivided by dislocation boundaries, including geometrically necessary boundaries and incidental dislocation boundaries. With increasing strain, the deformation substructure becomes more refined, boundary spacing decreases, and the misorientation across deformation-induced boundaries increases. These changes increase the stored-energy heterogeneity of the deformed matrix and provide a higher driving force for recrystallization. Therefore, the faster recrystallization observed after 90% reduction can be attributed to the more heavily subdivided deformation structure and the higher stored-energy state. As a result, the recrystallization timescale is compressed into the range of seconds.

Initial reduction strongly influences both the pre-annealing microstructure and the subsequent refinement during annealing. Prior to heat treatment, increasing rolling reduction already leads to significant grain subdivision, with the maximum grain size decreasing from $\sim 239 \mu\text{m}$ at 60% reduction to $\sim 72 \mu\text{m}$ at 90% reduction (Figure 4-23). This indicates that a substantial portion of the refinement is achieved during deformation through the formation of substructures and boundary networks.

During annealing, this difference in initial state directly affects the final microstructure (Figure 5-29 and Figure 5-38). In the 60% rolled samples, the coarse deformation bands are progressively replaced by a refined structure, but the maximum grain size remains relatively high ($\sim 106\ \mu\text{m}$ after 10 s). In contrast, the 90% reduction condition produces a much finer and more uniform microstructure almost immediately, with the maximum grain size reduced to $\sim 32\ \mu\text{m}$ at 1 s and further to $\sim 27\ \mu\text{m}$ at 10 s. So, higher stored energy not only accelerates recrystallization kinetics but also amplifies the refinement already introduced during deformation, leading to a finer and more homogeneous final grain structure.

From a crystallographic perspective, both reduction levels weaken the rolling texture, but the evolution differs in rate. The 90% reduced samples quickly approach a nearly random texture state, whereas the 60% reduced samples show a more gradual suppression of the deformation fibers. In particular, the Brass-dominated fiber persists through the intermediate stages before being consumed. In the 90% reduction condition, the high density of nucleation sites and greater driving force likely promote a wider variety of orientations, preventing the selection of a single dominant recrystallization fiber.

5.3.2.2 Restoration Mechanisms during Rapid Annealing at 520 °C

I. Early-Stage Restoration at 520 °C: Morphological Persistence and Selective Consumption

At the onset of annealing at 520 °C in the 60% UCR condition, a morphological paradox is observed: the recrystallized area fraction increases rapidly to approximately 49% by 2 s, while the elongated TD-aligned deformation-band morphology remains clearly visible. This shows that a large increase in the recrystallized fraction does not immediately result in the disappearance of the pre-existing deformation morphology. At 2 s, the HAGB fraction has already increased to approximately 57%, indicating that recrystallization and boundary migration are well underway, while a substantial part of the TD-aligned deformation structure is still retained. The persistence of this morphology therefore reflects the spatially heterogeneous character of the transformation rather than an absence of recrystallization.

The observation of approximately 21% recrystallization after only 1 s indicates that the incubation period, if present, is very short under these conditions. During the first stages of annealing, the newly recrystallized regions appear to remain spatially constrained by the original elongated deformation structure. This indicates that recrystallization begins locally within or adjacent to the existing deformation-band structure rather than immediately producing a fully equiaxed microstructure. Lin et al. (2022) showed that recrystallization-related boundary migration is affected not only by stored energy, but also by the crystallographic relationship between the recrystallized region and the surrounding deformed matrix. Therefore, the coexistence of recrystallized grains with persistent elongated regions is consistent with locally different conditions for nucleation and boundary migration, while the overall morphology remains influenced by the prior deformation pattern (F. Lin, Zhang, Godfrey, & Juul Jensen, 2022).

As annealing progresses from 2 s to 6 s, the grain-size distribution of the NR subset shows an apparent increase in its maximum dimension from approximately 150 μm to 190 μm . This does not necessarily represent grain growth of the NR regions. Instead, it reflects the increasingly heterogeneous population that remains after a large part of the deformed matrix has recrystallized. At 6 s, the total recrystallized fraction reaches approximately 85%, leaving only a small fraction of the original deformation structure non-recrystallized. As recrystallization selectively consumes different parts of the deformation structure, the surviving NR population becomes increasingly dominated by elongated remnants that extend over relatively large distances along TD. Consequently, the maximum size measured within the NR subset can increase even though the overall microstructure is becoming substantially finer.

This survivorship effect provides clear evidence that restoration in AA6082 is controlled by heterogeneous stored energy and nucleation behavior. Recrystallization initiates preferentially in highly strained regions, whereas less strained regions persist longer and are consumed at later stages, consistent with selective consumption mechanisms reported in the literature (K. F. Adam, Long, & Field, 2017).

II. Functional Roles of Rotated Goss and Brass

The RG orientation shows the strongest persistence within the non-recrystallized fraction during annealing at 520 °C. This behavior is consistent with the distinct deformation substructure associated with this orientation. Previous studies have shown that recrystallization is strongly controlled by the local deformation structure and the crystallographic relationship between neighboring regions (Paul, Driver, Maurice, & Jasiński, 2002). In RG-oriented regions, deformation develops through a different substructural pattern, with a more distributed arrangement of dislocation boundaries formed by slip on multiple planes (W. C. Liu, Juul Jensen, & Morris, 2001). Such a substructural configuration may influence the interaction between migrating recrystallization boundaries and the remaining deformed matrix, allowing RG-oriented regions to persist preferentially during annealing.

This behavior is clearly reflected by the Retention Index (RI) (Figure 5-34). RG exhibits the highest RI throughout the investigated annealing times. Its RI is approximately 0.77 at 1 s and remains high at 0.74 and 0.80 after 2 and 4 s, respectively. Although retention decreases as recrystallization progresses, RG still exhibits RI values of approximately 0.44 and 0.43 at 6 and 8 s and remains the most strongly retained component at 10 s. Thus, the surviving NR texture becomes progressively dominated by RG as recrystallization advances. The resistance score (RS) (Figure 5-35) provides complementary information about the internal state of these surviving regions. RG consistently exhibits the highest RS, with values ranging approximately from 0.75 to 0.98. Because a high RS corresponds to relatively high GOS and GOS/D compared with the other NR components, the surviving RG regions retain substantial intragranular misorientation and orientation gradients. The combination of high RI and high RS indicates that RG remains strongly retained while preserving substantial intragranular misorientation and orientation gradients. This behavior points to restricted effective boundary mobility associated with the local crystallographic and substructural configuration of RG-oriented regions. A similar tendency was observed during annealing at 350 and 400 °C, where RG-rich regions remained strongly retained despite relatively high GOS and GOS/D values.

In contrast, the Brass (Bs) component shows a different retention behavior. Although Bs is initially strongly represented in the rolled deformation texture and remains relatively persistent during the earliest stage of annealing, it is progressively consumed as recrystallization proceeds. At 1 s, its RI is still relatively high at approximately 0.70, but it decreases to 0.37 at 2 s, 0.33 at 4 s, and only 0.14 at 6 s. By 8 s, its RI falls to approximately 0.10. The rapid decrease in Bs retention compared with RG indicates that Bs-oriented regions become increasingly susceptible to consumption once recrystallization progresses through the deformation structure.

The comparison between RG and Bs therefore demonstrates that persistence of a texture component cannot be interpreted from its internal deformation state alone. RG combines high retention with high RS, suggesting that restricted boundary mobility plays an important role in its survival. Bs, in contrast, retains appreciable deformation-related substructure at early times but is progressively consumed as recrystallization advances. The RI and RS therefore provide complementary measures that distinguish the amount of an orientation that survives from the internal state of the surviving regions.

5.4 Conclusions on Annealing Behavior of Cold-Rolled AA6082

This study shows that the recrystallization behavior of cold-rolled AA6082 is governed by the combined effects of annealing temperature, prior deformation, and local crystallographic orientation. Increasing the annealing temperature strongly accelerates restoration, shifting the transformation from prolonged annealing at 350–400 °C to a second-scale process at 520 °C. Increasing the rolling reduction from 60% to 90% further enhances recrystallization and promotes the formation of a finer and more homogeneous grain structure.

The recrystallization process also shows clear orientation dependence. RG is preferentially retained within the non-recrystallized regions and consistently exhibits the highest resistance score, while other deformation components are progressively consumed as annealing proceeds. This behavior shows that some crystallographic orientations are more resistant to the progression of recrystallization and that the local deformation substructure influences their

persistence. The evolution of RG therefore highlights the important role of crystallographic orientation in controlling the final stages of recrystallization and the resulting texture of AA6082.

CONCLUSIONS

This thesis investigated the microstructure and texture evolution of AA6xxx series aluminum alloys under cold rolling, pseudo-hot rolling, and subsequent annealing conditions. By systematically comparing strain paths (unidirectional vs. cross-rolling), alloy chemistry (AA6063 vs. AA6082), and initial states (as-cast vs. as-extruded), the results establish a consistent framework linking deformation, restoration, and recrystallization behavior.

During cold rolling, structural refinement proceeds through progressive dislocation accumulation, subgrain formation, and the gradual conversion of LAGBs into HAGBs. At high reductions, both UCR and CCR lead to a heavily subdivided and banded microstructure. However, the strain path governs the kinetics of this evolution. UCR promotes continuous lattice rotation and rapid grain subdivision, whereas CCR, through repeated 90° rotations, interrupts boundary alignment and delays fragmentation at intermediate stages. This difference is even more pronounced in the texture evolution. UCR produces a classical β -fiber rolling texture with strong S-component development, while CCR promotes a transition toward a Brass-dominated texture. This shift arises from the redistribution of slip activity under non-proportional loading, demonstrating that strain-path modification alone can alter texture development even in high stacking-fault energy alloys.

Under pseudo-hot rolling conditions, the microstructural response reflects a balance between strain accumulation and restoration. In AA6063, the relatively low solute content and limited particle density result in high boundary mobility and weak substructure stability. As a result, recovery processes locally dominate at intermediate reductions, leading to a non-monotonic evolution characterized by reduced LAGB density and transient coarsening. In contrast, AA6082 retains a higher level of stored energy due to solute drag and particle pinning, enabling sustained subdivision at 90% reduction. The observed microstructures at this stage are characterized by localized grain-boundary protrusions consistent with bulging, suggesting local boundary migration driven by stored-energy differences within the severely deformed microstructure.

The initial microstructure further influences the deformation response. In as-extruded AA6082, the inherited elongated grain morphology and texture persist through intermediate reductions, indicating a strong memory effect. This inherited structure begins to break down only at high strain, where restoration mechanisms become sufficiently active. The strain path modifies this transition. UHR maintains directional alignment and preserves aspects of the inherited structure, whereas CHR redistributes deformation and disrupts the continuity of elongated features, producing a more homogeneous and equiaxed microstructure.

Annealing behavior introduces a further transition controlled by temperature and stored energy. At intermediate temperatures (350–400 °C), restoration proceeds slowly and is dominated by recovery, as particle pinning limits boundary mobility. At 520 °C, a rapid transformation regime is observed, where recrystallization occurs within seconds due to increased boundary mobility and high stored energy from prior deformation. The level of cold reduction strongly affects this response. Higher reductions increase the stored energy and accelerate recrystallization while also leading to finer final grain structures through a higher density of nucleation sites.

The evolution during rapid annealing shows that recrystallization does not occur uniformly. In the early stages, high recrystallized fractions can be reached while deformation morphology is still partially preserved, indicating that internal substructure rearrangement precedes large-scale boundary migration. At intermediate stages, selective consumption of highly strained regions leads to temporary coarsening in grain-size statistics, before full recrystallization is achieved at later stages.

Texture evolution across all processing stages reflects the interaction between deformation and restoration. Rolling textures dominated by β -fiber components (S and Brass) are maintained under both cold and pseudo-hot rolling, while strain-path changes redistribute their relative contributions. During annealing, the deformation texture weakens, but key components persist, indicating that texture evolution is controlled by both kinetic constraints and local substructure stability. The persistence of Brass and the stability of rotated Goss orientations further highlight the heterogeneous nature of stored energy and its influence on recrystallization.

Overall, the results demonstrate that the final microstructural and texture state of AA6xxx alloys is determined by the interaction between strain path, alloy chemistry, initial condition, and thermal exposure. UCR and UHR favor directional alignment and stronger texture development when recovery is limited, whereas CCR and CHR promote redistribution of deformation, reduce anisotropy, and enhance microstructural homogeneity. The annealing response further depends on the stored energy introduced during deformation, which controls both the kinetics and the scale of refinement. These findings provide a mechanistic basis for tailoring thermomechanical processing routes to achieve targeted combinations of grain structure, texture, and anisotropy in 6xxx series aluminum alloys.

RECOMMENDATIONS

Several directions can be identified to extend the findings of this study and further develop the understanding of deformation and restoration behavior in AA6xxx alloys.

A first priority is the development of a quantitative modeling framework linking stored energy distribution to recrystallization kinetics. Integrating the present experimental data into deformation and texture evolution models would enable predictive simulations under complex strain paths. In particular, coupling these results with Crystal Plasticity Finite Element (CPFE) models would allow for a more detailed description of orientation evolution, slip activity, and the spatial distribution of stored energy during both deformation and annealing.

Further investigation of second-phase particles and precipitation effects is also required. While the present work highlights the influence of solutes and particles on substructure stability, the role of the precipitation sequence (GP zones $\rightarrow \beta'' \rightarrow \beta' \rightarrow \beta$) remains to be fully addressed. Future studies should consider pre-aging or over-aging treatments prior to deformation to evaluate how precipitate size, distribution, and coherency affect dislocation motion, Zener pinning, and recrystallization behavior. Such an approach would help establish a direct link between thermomechanical processing and final age-hardening potential.

The influence of strain-path modification can be explored beyond the 90° rotation used in CCR/CHR. Intermediate rotation angles (e.g., 45° or 60°) should be investigated to determine whether a critical threshold exists for disrupting deformation textures and promoting Brass-dominated states in high stacking-fault energy alloys. In addition, asymmetric rolling (ASR), which introduces through-thickness shear strain, could be compared with cross-rolling to evaluate its effectiveness in achieving more uniform grain refinement and texture redistribution.

From an experimental standpoint, advanced characterization techniques would provide further insight into transient microstructural evolution. In-situ annealing EBSD is particularly recommended, as it enables direct observation of boundary migration, substructure rearrangement, and orientation changes during heating. High-resolution orientation mapping

could also be used to better resolve local misorientation gradients and early-stage nucleation phenomena during recrystallization.

Finally, a more detailed analysis of particle characteristics, including size distribution, morphology, and spatial arrangement, would improve understanding of their contribution to boundary pinning and microstructural stability. Combining such characterization with modeling efforts would help clarify the interaction between particles, stored energy, and grain boundary mobility.

APPENDIX I

THERMOMECHANICAL AND STRAIN PATH EFFECTS ON THE ANISOTROPY OF AA6082 AND AA6063

This appendix presents supplementary mechanical data for the AA6082 and AA6063 aluminum alloys. The following sections present the true stress–strain responses, R-values, and strain hardening exponents across the selected thermomechanical conditions. The data are categorized by deformation temperature (cold versus pseudo-hot rolling), strain path (unidirectional versus cross rolling), and initial microstructural state (as-cast versus as-extruded), providing the supporting datasets for the analyses discussed in the CHAPTER 4.

I.1 Mechanical and Microstructural Evolution during Cold Rolling (UCR and CCR)

The true stress-strain curves for the cold rolled as-cast AA6082 (Figure-A I-1) and the cold rolled as-extruded AA6082 (Figure-A I-2), alongside the cold rolled as-cast AA6063 alloys (Figure-A I-3), demonstrate the strain hardening characteristic of severe plastic deformation. Increasing the rolling reduction from 40% to 90% yields a corresponding increase in the Ultimate Tensile Strength (UTS) across the selected conditions. The as-extruded AA6082 subjected to 90% UCR exhibits the highest overall strength, exceeding 500 MPa, reflecting the combined effects of the initial extruded state and subsequent cold working.

For the as-cast AA6082 alloy, a clear dependence on tensile orientation and rolling path is observed, with variations specific to the alloy composition (Figure-A I-1 and Table-A I-1). In the UCR condition at 40% reduction, the UTS values remain closely clustered across the 0°, 45°, and 90° directions, indicating limited initial strength anisotropy. By 90% UCR, the directional profile becomes more pronounced; the lowest UTS is observed at 45° and the highest at 90°, while the R-value sharply peaks at 45° with a maximum of 1.6. Under the CCR path, the evolution of anisotropy takes a different form. At 40% CCR, both UTS and R-values remain within a narrow range, similar to the UCR state. At 90% CCR, the UTS values remain

relatively close across all directions, with only a modest increase at 45° (329 MPa). The R-value under severe CCR again maximizes at 45° , while dropping significantly in the transverse and rolling directions. The strain-hardening exponent (n-value) for cold-rolled as-cast AA6082 remains uniformly low across all conditions, reflecting the material's limited capacity for continued uniform strain.

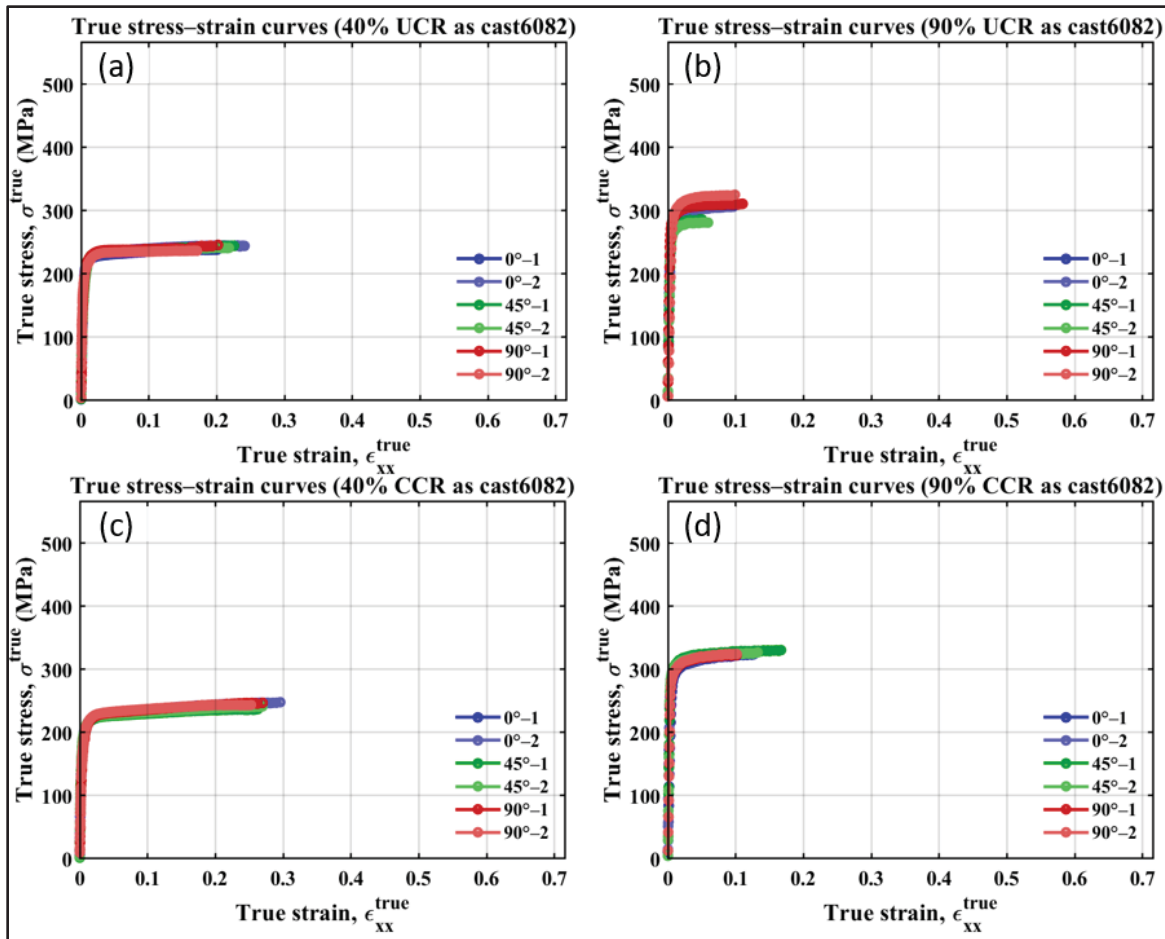


Figure-A I-1 True stress-strain curves of as-cast AA6082 under cold rolling conditions: (a) 40% UCR, (b) 90% UCR, (c) 40% CCR, and (d) 90% CCR. Tensile specimens were tested at 0° , 45° , and 90° relative to RD

Table-A I-1 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-cast AA6082 under cold rolling conditions (UCR and CCR) at 40% and 90% reduction, measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UCR 40%_as-cast6082	0°	242 ± 4	0.8 ± 0.0	0.04 ± 0.01
	45°	243 ± 2	1.4 ± 0.2	0.05 ± 0.01
	90°	242 ± 6	1.3 ± 0.1	0.04 ± 0.00
UCR 90%_as-cast6082	0°	309 ± 3	0.4 ± 0.0	0.04 ± 0.01
	45°	285 ± 4	1.6 ± 0.3	0.04 ± 0.01
	90°	318 ± 10	0.7 ± 0.4	0.05 ± 0.02
CCR 40%_as-cast6082	0°	247 ± 2	1.1 ± 0.1	0.06 ± 0.01
	45°	239 ± 3	1.2 ± 0.1	0.04 ± 0.00
	90°	246 ± 2	1.3 ± 0.1	0.06 ± 0.00
CCR 90%_as-cast6082	0°	323 ± 3	0.6 ± 0.0	0.05 ± 0.01
	45°	329 ± 3	1.4 ± 0.1	0.03 ± 0.00
	90°	324 ± 1	0.7 ± 0.0	0.04 ± 0.00

The deformation behavior of the as-extruded AA6082 alloy subjected to 90% UCR confirms the persistence of strong orientation-dependent characteristics even in a pre-processed material state (Figure-A I-2 and Table-A I-2). The UTS exhibits a clear directional trend, recording its minimum at 0° (509 MPa) and steadily increasing to its maximum at 90° (537 MPa). The plastic strain ratio follows the established anisotropic pattern, reaching a distinct maximum R-value of 1.2 at the 45° orientation. The strain-hardening exponent (n-value) remains constant at a low value of 0.07 across all directions, indicating an exhausted capacity for further work hardening.

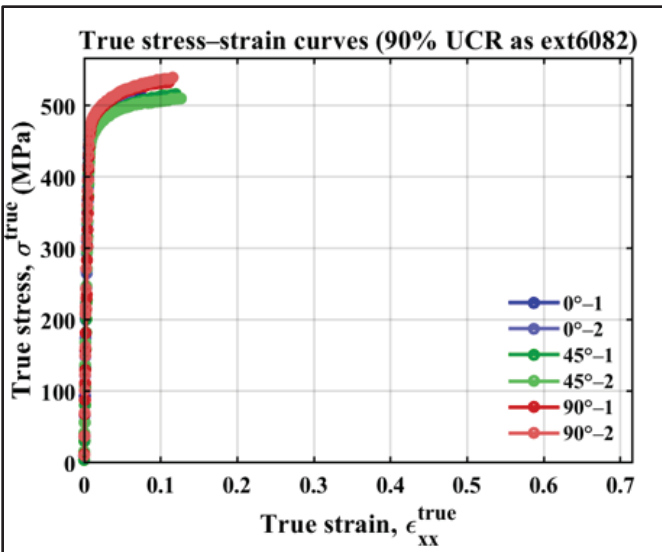


Figure-A I-2 True stress–strain curves of as-extruded AA6082 after 90% UCR. Tensile specimens were tested at 0°, 45°, and 90° relative to RD

Table-A I-2 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-extruded AA6082 after 90% UCR, measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UCR 90%_as-extruded 6082	0°	509 ± 3	0.5 ± 0.0	0.07 ± 0.00
	45°	515 ± 2	1.2 ± 0.2	0.07 ± 0.00
	90°	537 ± 5	0.9 ± 0.0	0.07 ± 0.01

In the as-cast AA6063 alloy, the mechanical response demonstrates a distinct evolution of orientation-dependent behavior under different cold rolling paths (Figure-A I-3 and Table-A I-3). Under UCR, the UTS develops a V-shaped anisotropic profile, with the minimum strength consistently observed at the 45° orientation at both 40% and 90% reduction levels. The plastic strain ratio (R-value) complements this trend, reaching its maximum at 45° and 90°, while demonstrating its lowest value at 0°. When subjected to CCR, the material exhibits a fundamentally different anisotropic signature. At 40% CCR, the UTS remains relatively uniform across all test directions, showing limited directional dependence. However, after 90%

CCR, the anisotropic profile reverses compared to the UCR condition; the UTS reaches a maximum of 391 MPa at 45°, while the R-value is distinctly concentrated at 45°. Across all these cold-rolled states for the as-cast AA6063, the strain hardening exponent (n-value) remains consistently low (typically not exceeding 0.11), indicating a limited capacity for further work hardening.

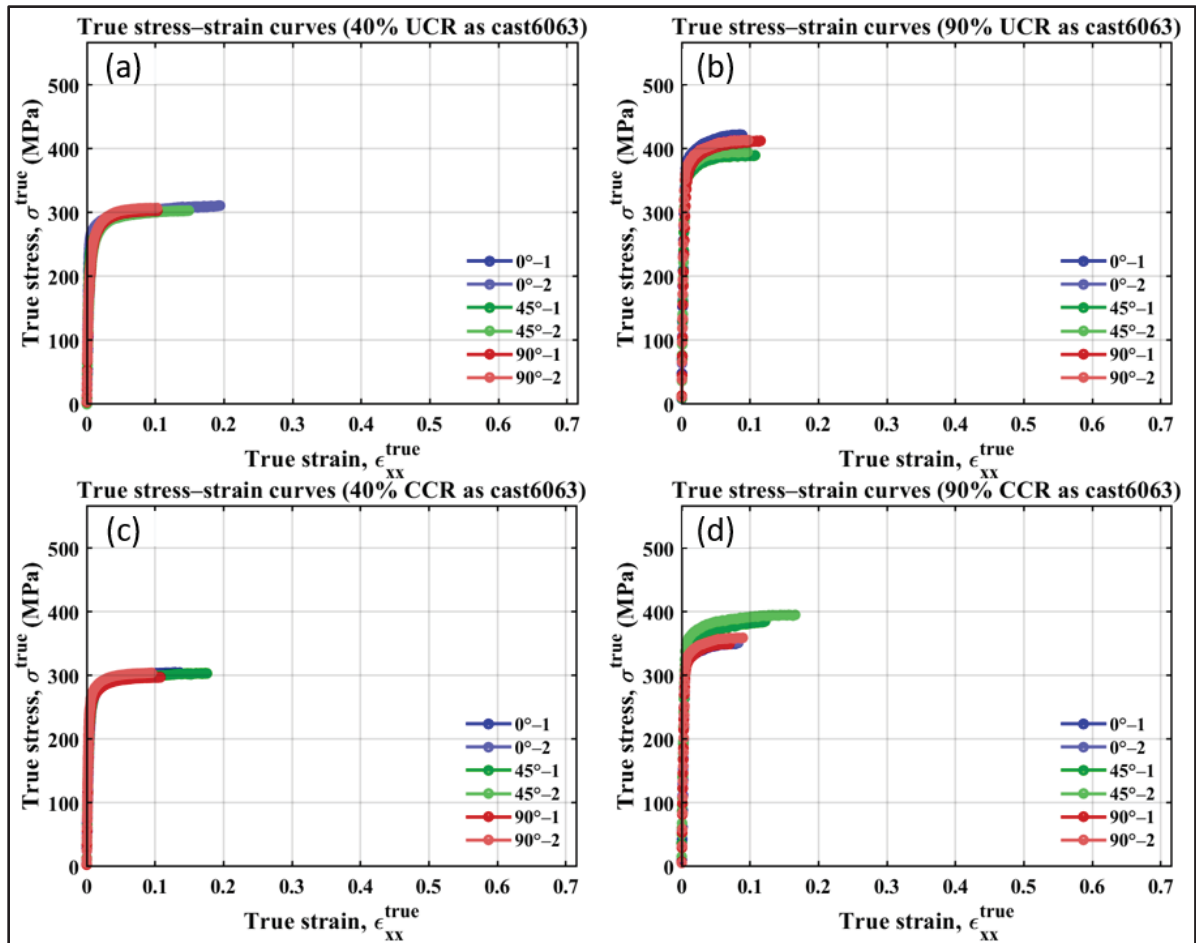


Figure-A I-3 True stress-strain curves of as-cast AA6063 under cold rolling conditions: (a) 40% UCR, (b) 90% UCR, (c) 40% CCR, and (d) 90% CCR. Tensile specimens were tested at 0°, 45°, and 90° with respect to the rolling direction (RD)

Table-A I-3 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-cast AA6063 under cold rolling conditions (UCR and CCR) at 40% and 90% reduction, measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UCR 40%_as-cast6063	0°	311 ± 1	0.8 ± 0.0	0.07 ± 0.04
	45°	303 ± 0	1.2 ± 0.1	0.08 ± 0.02
	90°	306 ± 3	1.3 ± 0.1	0.11 ± 0.02
UCR 90%_as-cast6063	0°	418 ± 7	0.4 ± 0.0	0.05 ± 0.00
	45°	393 ± 4	1.3 ± 0.0	0.04 ± 0.00
	90°	414 ± 1	1.1 ± 0.0	0.06 ± 0.02
CCR 40%_as-cast6063	0°	302 ± 5	0.9 ± 0.0	0.09 ± 0.02
	45°	301 ± 5	1.1 ± 0.0	0.07 ± 0.02
	90°	301 ± 5	1.0 ± 0.0	0.06 ± 0.01
CCR 90%_as-cast6063	0°	353 ± 1	0.6 ± 0.1	0.04 ± 0.00
	45°	391 ± 7	1.0 ± 0.0	0.04 ± 0.01
	90°	355 ± 6	0.7 ± 0.1	0.05 ± 0.00

I.2 Mechanical and Microstructural Evolution during Pseudo-hot Rolling (UHR and CHR)

The true stress-strain curves for the pseudo-hot rolled as-cast AA6082 (Figure-A I-4) and the cold rolled as-extruded AA6082 (Figure-A I-5), alongside the cold rolled as-cast AA6063 alloys (Figure-A I-6), illustrate the competing effects of strain hardening and dynamic restoration at elevated temperatures. Unlike the cold-rolled conditions, increasing the pseudo-hot rolling reduction from 40% to 90% results in a decrease in the UTS for both as-cast alloys. For instance, the as-cast AA6082 under UHR conditions drops from approximately 350 MPa at 40% reduction to nearly 300 MPa at 90% reduction. This strength reduction at higher deformation levels is consistent with dynamic softening mechanisms (such as recovery or recrystallization) counteracting the work hardening process. The as-extruded AA6082 subjected to 90% CHR (Figure-A I-5b) maintains a higher strength profile, exceeding 417

MPa, reflecting the influence of its distinct initial extruded microstructure prior to hot deformation.

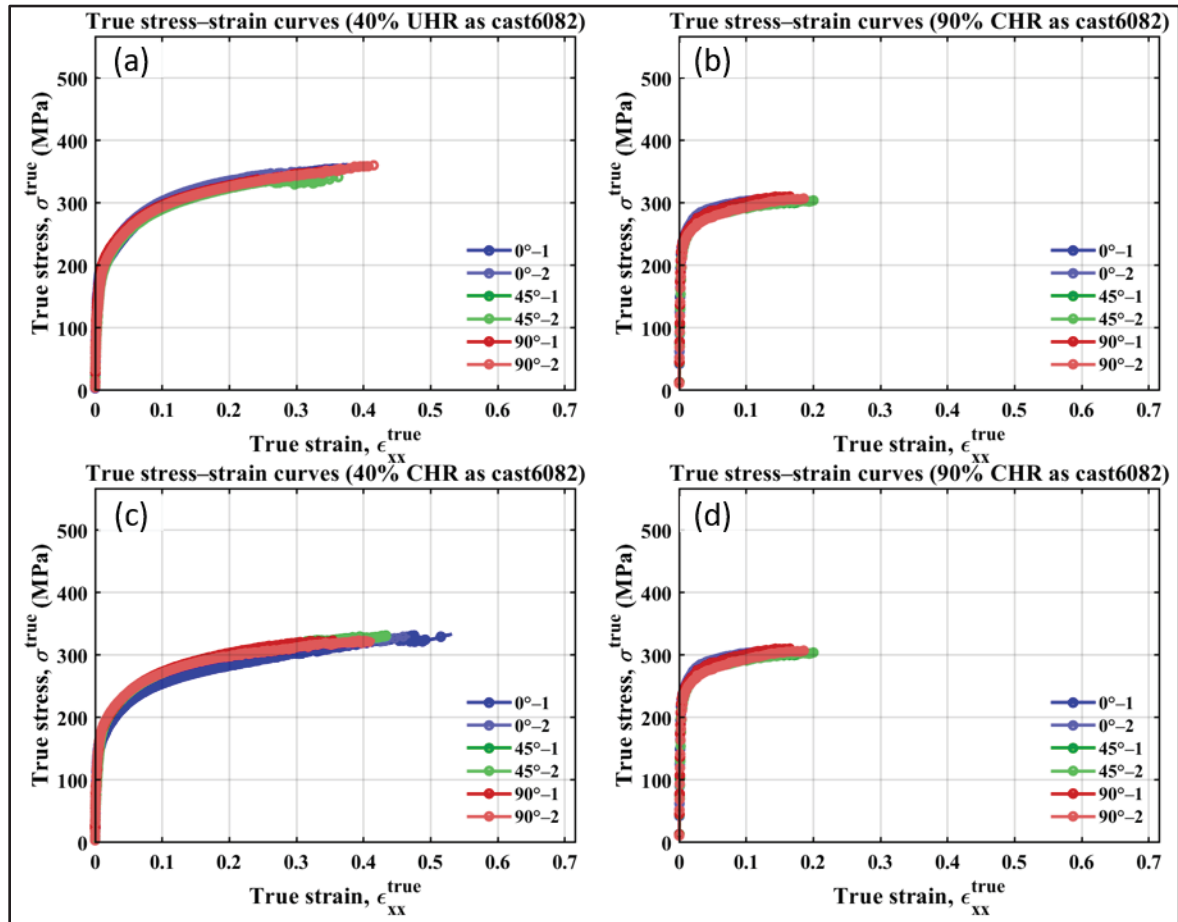


Figure-A I-4 True stress–strain curves of as-cast AA6082 under pseudo-hot rolling conditions: (a) 40% UHR, (b) 90% UHR, (c) 40% CHR, and (d) 90% CHR. Tensile specimens were tested at 0°, 45°, and 90° relative to RD

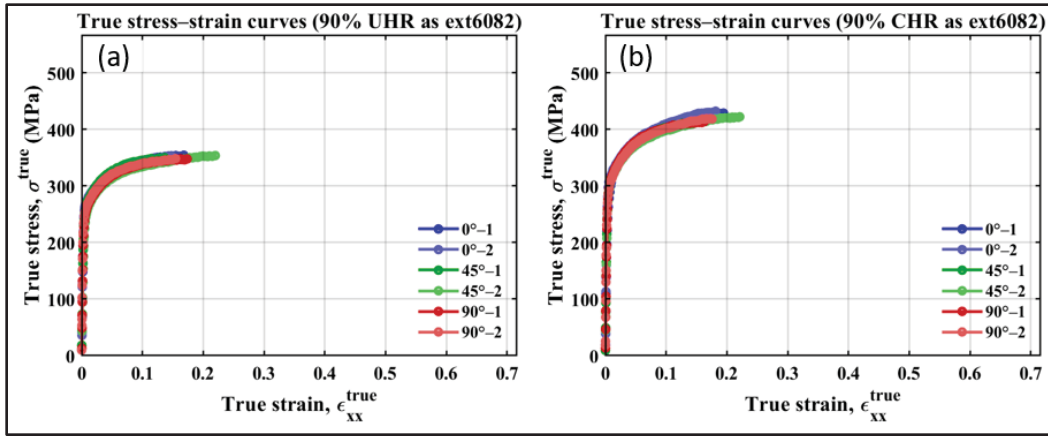


Figure-A I-5 True stress-strain curves of as-extruded AA6082 after 90% pseudo-hot rolling: (a) UHR and (b) CHR. Tensile specimens were tested at 0°, 45°, and 90° relative to RD

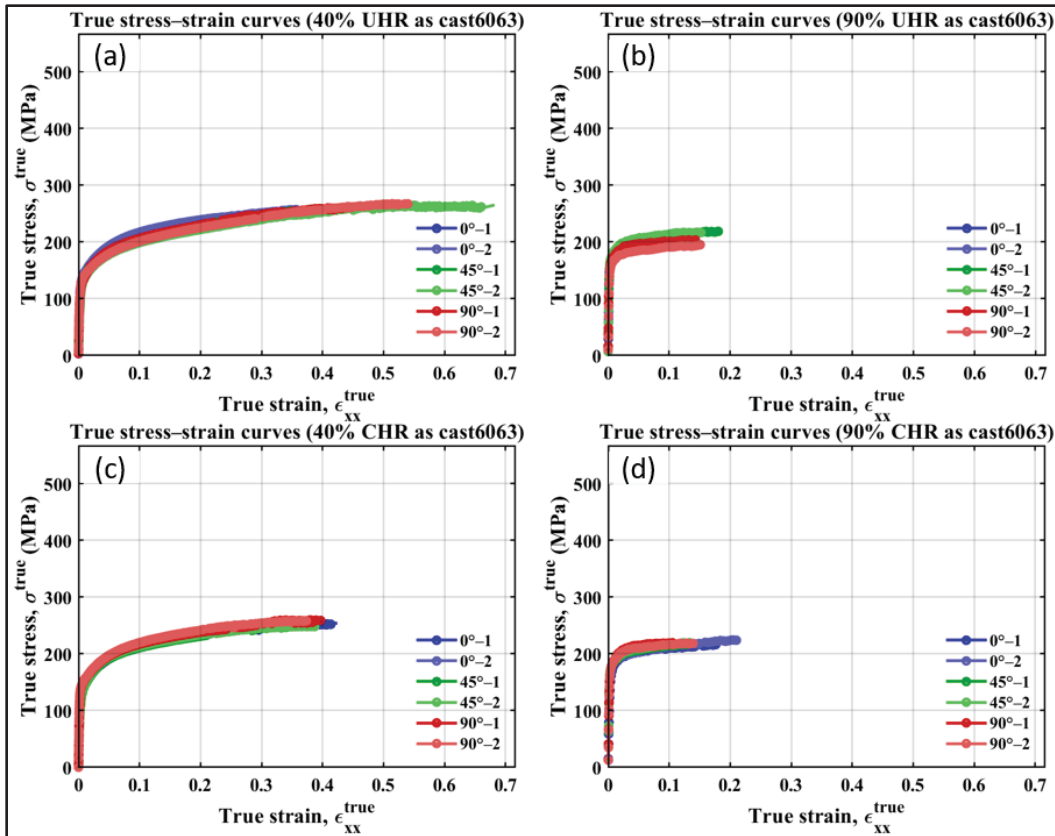


Figure-A I-6 True stress-strain curves of as-cast AA6063 under pseudo-hot rolling conditions: (a) 40% UHR, (b) 90% UHR, (c) 40% CHR, and (d) 90% CHR. Tensile specimens were tested at 0°, 45°, and 90° with respect to the rolling direction (RD)

For the as-cast AA6082 alloy, a small directional variations in UTS but limited overall anisotropy (Figure-A I-4 and Table-A I-4) is observed. Under UHR at 40% reduction, the UTS values remain closely grouped across 0°, 45°, and 90°, with a slightly lower value at 45°. A similar tendency is observed at 90% reduction, where the change in UTS is not significant, although the highest value is observed at 0° and the lower values at 45° and 90°. Under CHR, the UTS remains within a narrow range at both 40% and 90% reduction, with only minor differences between orientations. The R-values vary between approximately 0.4 and 0.9, with a slight tendency to peak at 45°, but remain generally below unity. This confirms that the material exhibits only weak plastic anisotropy, even when some directional variation in strength is present. The n-values remain moderate at 40% and decrease at 90%, indicating a progressive reduction in strain hardening capability.

Table-A I-4 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-cast AA6082 under pseudo-hot rolling conditions (UHR and CHR) at 40% and 90% reduction, measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UHR 40%_as-cast6082	0°	354 ± 3	0.8 ± 0.0	0.18 ± 0.02
	45°	344 ± 4	0.9 ± 0.1	0.19 ± 0.00
	90°	356 ± 7	0.7 ± 0.1	0.18 ± 0.01
UHR 90%_as-cast6082	0°	303 ± 2	0.6 ± 0.1	0.09 ± 0.01
	45°	297 ± 5	0.8 ± 0.1	0.09 ± 0.00
	90°	295 ± 4	0.6 ± 0.0	0.09 ± 0.01
CHR 40%_as-cast6082	0°	332 ± 2	0.7 ± 0.1	0.17 ± 0.01
	45°	331 ± 1	0.9 ± 0.0	0.18 ± 0.00
	90°	325 ± 1	0.9 ± 0.1	0.17 ± 0.01
CHR 90%_as-cast6082	0°	306 ± 2	0.5 ± 0.1	0.07 ± 0.00
	45°	303 ± 2	0.6 ± 0.1	0.08 ± 0.00
	90°	310 ± 2	0.4 ± 0.0	0.08 ± 0.00

In the as-extruded AA6082 alloy at 90% reduction, the UTS shows a clearer directional dependence compared to the as-cast conditions (Figure-A I-5 and Table-A I-5). Under UHR, the UTS remains nearly constant across all directions, with only a slight decrease from 0° to 90°. In contrast, under CHR, the UTS is significantly higher and shows a noticeable decrease from 0° to 90°, indicating a more pronounced orientation dependence in strength. Despite this, the R-values remain low in both cases, ranging from approximately 0.5 to 0.7, with a modest maximum at 45°. These values indicate that, even in the presence of directional strength variation, the material does not exhibit strong normal anisotropy. The n-values remain low and relatively stable, reflecting a limited but non-negligible capacity for further plastic deformation.

Table-A I-5 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-extruded AA6082 after 90% pseudo-hot rolling (UHR and CHR), measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UHR 90%_as-ext 6082	0°	354 ± 0	0.5 ± 0.02	0.09 ± 0.00
	45°	352 ± 3	0.6 ± 0.2	0.10 ± 0.03
	90°	349 ± 0	0.5 ± 0.0	0.10 ± 0.00
CHR 90%_as-ext 6082	0°	431 ± 2	0.5 ± 0.03	0.11 ± 0.01
	45°	420 ± 5	0.7 ± 0.0	0.10 ± 0.00
	90°	417 ± 4	0.5 ± 0.05	0.11 ± 0.02

In the as-cast AA6063 alloy, the hot-rolled conditions show only a limited and non-systematic dependence of UTS on tensile orientation (Figure-A I-6 and Table-A I-6). Under UHR at 40% reduction, the UTS increases slightly from 0° to 90°, indicating a mild directional effect, whereas at 90% UHR the trend changes, with the maximum at 45° and the minimum at 90°. Under CHR, the UTS again remains relatively uniform at 40% reduction and becomes nearly identical across all directions at 90%, indicating a largely isotropic strength response at high reduction. The R-values remain consistently below or close to unity in all cases (≈ 0.5 – 1.0), with only minor directional variation. This indicates that, despite small changes in UTS with

orientation, the material does not exhibit strong normal anisotropy in the Lankford sense. The n-values decrease with increasing reduction, reflecting the reduced strain hardening capacity at higher deformation levels.

Table-A I-6 Mechanical properties (UTS, R-value, and strain hardening exponent n) of as-cast AA6063 under pseudo-hot rolling conditions (UHR and CHR) at 40% and 90% reduction, measured at 0°, 45°, and 90° relative to RD

Condition	Angle	UTS	R	n
UHR 40%_as-cast6063	0°	256 ± 4	0.8 ± 0.0	0.16 ± 0.01
	45°	265 ± 0	0.8 ± 0.0	0.17 ± 0.02
	90°	266 ± 5	1.0 ± 0.2	0.17 ± 0.02
UHR 90%_as-cast6063	0°	211 ± 2	0.6 ± 0.1	0.07 ± 0.03
	45°	219 ± 1	0.6 ± 0.1	0.06 ± 0.01
	90°	199 ± 6	0.8 ± 0.1	0.05 ± 0.03
CHR 40%_as-cast6063	0°	252 ± 4	0.8 ± 0.1	0.15 ± 0.04
	45°	249 ± 0	0.8 ± 0.2	0.15 ± 0.04
	90°	260 ± 1	0.8 ± 0.0	0.14 ± 0.04
CHR 90%_as-cast6063	0°	221 ± 6	0.5 ± 0.1	0.06 ± 0.02
	45°	219 ± 2	0.6 ± 0.0	0.06 ± 0.01
	90°	220 ± 1	0.7 ± 0.1	0.06 ± 0.00

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