Process parameter influence on deformation and recrystallization textures in Al alloys

Jurij J Sidor¹, Qingge Xie², Purnima Chakravarty¹, Gyula Pal¹

¹Savaria Institute of Technology, Faculty of Informatics, Eötvös Loránd University (ELTE), Szombathely, Hungary
²Collaborative Innovation Center of Steel Technology, University of Science and Technology Beijing, Beijing 100083, China
Corresponding author’s e-mail: js@inf.elte.hu

Abstract. Mesoscopic changes take place during the entire thermomechanical processing (TMP) of polycrystalline metallic materials. Controlling the TMP parameters is of key importance since both microstructure and texture evolution of a particular technological step severely influence the final properties of a given product. This study is focusing on the crystallographic aspect of microstructural formation in Al alloys. The effect of straining level on the crystallographic texture evolution and texture-dependent properties is discussed. This contribution clearly demonstrates that the plastic strain ratio can be modified via TMP route. Implementation of relatively small deformation could account for improvement of anisotropy of plastic yielding, necessary for advanced deep drawing qualities.

1. Introduction

Production of polycrystalline Al alloys involves continuous change on microstructural level and these mesoscopic changes strongly influence the mechanical properties of materials. In case of sheet metal products, the thermomechanical processing (TMP) involves direct chill casting, homogenization, hot and cold rolling (intermediate annealing, if necessary) and final recrystallization annealing. Although it seems that each stage of TMP chain is an independent event, the results of numerous investigations [1, 2] clearly indicate that there is a “genetic” connection between the individual processing steps. Furthermore, each TMP step affects the behavior of a consequent one as well as the final properties of a material.

Microstructure of polycrystalline aggregate, exposed to a sequence of deformation and annealing processes, is subjected to morphological changes, which depend on the straining level, strain rate, and annealing history. When it comes to crystallographic aspect of mesoscopic changes, the evolution becomes less apparent, compared to morphological one, since one type of texture tends to replace the preceding one and the connection between the two counterparts is not always obvious. A typical example is the evolution of recrystallization texture during discontinuous annealing in face centered cubic (FCC) metals. Particularly, the α and β fibers, which typically evolve during rolling, tend to transform to a qualitatively new one, characterized by the strongly evolved θ-fiber components mixed with the weakly developed α and γ orientations (see Tab. 1).

Both experimental evidence and implementation of numerical approaches are necessary in order to reveal the process parameters influence on the evolution of deformation and recrystallization textures. The present study aims to clarify the correlation among the TMP parameters, crystallographic texture...
and related properties. The major goal of this contribution is to show how the deformation affects the recrystallization texture and the plastic strain ratio.

Table 1. Main orientations and fibers, observed in materials with FCC crystal structure [1, 3].

| Fiber | Common axis of orientations |
|-------|-----------------------------|
| α     | ⟨011⟩ // Normal Direction   |
| β     | \( \{h, 1, h+1\} \left\{ \frac{h(h+1)}{3/4-h}, \frac{2h(h+1)}{1/2-h}, \frac{h^2}{h-3/4} + \frac{2h}{h-1/2} \right\} \) |
| γ     | ⟨111⟩ // Normal Direction   |
| θ     | ⟨001⟩ // Normal Direction   |
| η     | ⟨001⟩ // Rolling Direction  |

2. Materials, experimental and computational procedure

The evolution of crystallographic texture was studied in Al-Mg-Si alloy (AA6016 series) in the present investigation. Two investigated materials (A and B) featured different pre-rolling textures and were subjected to diverse straining levels. Material A was cold rolled with 86% thickness reduction, whereas material B was subjected to 30% (material B1) and 47% (material B2) reductions. Finally, the deformed sheets were annealed at 550°C with the aim to ensure a fully recrystallized state.

The initial, rolled and annealed samples were exposed to textures measurements across the thickness by means of electron backscattering diffraction (EBSD) technique. The orientation data, obtained from the plane perpendicular to the sample transverse direction (TD-plane), were collected and post-processed by the commercial OIM-TSL-8® software. The texture data were measured by the Hikari-type EBSD detector, which was attached to the scanning electron microscope (SEM) FEI®. In case of annealed samples, the orientation imaging microscopy (OIM) data were collected at the acceleration voltages ranging from 18 to 30 V with A2 Struers® electrolyte, cooled to temperatures ranging between -5 and 0°C. Application of 12-16 kV guaranteed appropriate OIM data quality. During OIM measurements, the investigated sheets were 70° tilted with respect to the EBSD detector and the mapping was performed on hexagonal scan grids. The textures were displayed in the \( \varphi_2=45°, \varphi_2=65° \) and \( \varphi_2=90° \) sections respectively, which show majority of components evolving during TMP.

Sample preparation for OIM was performed according to the standard procedure, which comprises mechanical grinding and polishing as well as electrolytic polishing. The mechanical polishing procedure was finished with the DiaDuo-2 Struers®-type 1 μm diamond paste. The electrolytic polishing, as a final step of sample preparation for EBSD, was conducted for ~1 min at voltages ranging from 18 to 30 V with A2 Struers® electrolyte, cooled to temperatures ranging between -5 and 0°C.

The plastic strain ratio \( (q\text{-value}) \) was simulated by a well-established Taylor-type homogenization approach Alamel [4]. The plastic strain ratio \( (q\text{-value}) \) is defined as a fraction of width to elongation in tension. The average \( q\text{-value} \), measured at different angles with respect to RD, describes the normal anisotropy whereas the planar anisotropy \( \Delta q \) is defined as a difference between the maximum and minimum \( q\text{-values} \). The calculation of \( q\text{-value} \) profiles was performed by crystal plasticity model by taking into account the \{111\}<110> octahedral slip systems, which typically operate during deformation at room temperatures in materials with FCC crystal structure.
3. Results and Discussions

3.1. Pre-rolling textures
The pre-rolling textures of materials A and B are shown in Fig. 1. The bunch of orientations, mainly distributed along the \( \alpha, \beta, \theta \) and \( \eta \) fibers, has evolved in the investigated materials. The textures presented in Fig. 1 a and b reveal significant qualitative and quantitative diversities. Material A is characterized by the mixture of cube \( \{001\}\langle100\rangle \) and \( \beta \)-fiber orientations. In contrast, material B shows strongly developed cube component, \( \{113\}\langle121\rangle \) orientation, a weak Goss texture \( \{011\}\langle100\rangle \) and traces of \( \beta \)-fiber orientations.

![Fig.1. Pre-rolling texture of investigated materials: a) material A; b) material B.](image)

3.2. Deformation textures
The development of texture during cold rolling, shown in Fig. 2, is characterized by the \( \alpha \) and \( \beta \) fibers. In case of severe straining level (material A, 86%), the rolling components are aligned along the \( \beta \)-fiber. The reference orientations of the \( \beta \)-fiber can be computed by means of equation presented in Table 1, according to which the position of each \( \beta \)-fiber component is a function of a single miller index \( h \). The deformation texture orientations in material A (Fig. 2a) show somewhat unequal distribution along the \( \beta \)-fiber \( (6.5<f(g)<11) \) with a maximum at \( \sim\{213\}\langle9 15 11\rangle \). As it is shown in Fig. 2a, the severe reduction of 86% is not capable of complete reorientation of cube component towards the deformation fiber texture. Compared to material A, material B, with strongly developed cube component, was subjected to less intense straining. The intensities of components along the skeleton line in samples B1 and B2, connecting the copper and brass in Euler space, tend to decline with an increase of \( h \), i.e., orientation intensity \( f(g) \) follows slight downward trend while moving from the \( \{112\}\langle111\rangle \) \( (h=1) \) via \( \{213\}\langle9 15 11\rangle \) \( (h=2) \) to \( \{011\}\langle112\rangle \) \( (h \to \infty) \). The cube component in samples B1 and B2 revealed fairly high stability against reorientation, as it was also observed in the material A. The major qualitative
The difference between the texture evolutionary patterns in materials A and B is related to the appearance of \( \alpha \)-fiber in samples B1 and B2, which was not present in material A (see Fig. 2 a-c).

3.3. Recrystallization texture

By comparison of Figs. 2 and 3, it becomes evident that the evolution of deformation texture has a great impact on the crystallographic aspect of microstructure development during recrystallization. The combined cube-\( \beta \)-fiber type deformation texture (Fig. 2a) has ensured a well evolved \( \theta \)-fiber, Goss, P (\( \{011\}\langle 231 \rangle \)) texture mixed with a week \( \gamma \)-fiber orientations (Fig. 3a) during recrystallization. In case of less intense deformation (30% and 47%), the cube-\( \alpha \)-\( \beta \)-fiber type rolling textures (Fig. 2 b and c) tended to transform to a cube dominating annealing texture (Fig. 3b and c). Both qualitative and quantitative differences in texture change might be explained by the varieties in the pre-rolling and deformation textures as well as diversities in the degree of reduction. Although both the initial and deformation textures of materials B1 and B2 showed strong cube components, the recrystallization ODFs of these samples exhibited weaker \( \theta \)-fiber components as compared to material A. Analysis of texture evolution presented in Figs. 2 and 3 suggests that a broad spectrum of orientations in the deformed state, obtained after relatively small straining levels, is capable of ensuring a qualitative diversity in the recrystallized matrix.

![Fig. 2. Deformation textures of investigated materials: a) material A, subjected to 86% reduction; b) material B, subjected to 30% reduction (B1); c) material B, subjected to 47% reduction (B2).](image)
The evolution of recrystallization textures shown in Fig. 3 can be explained by employing modeling approaches [2, 5, 6], where the nucleation spectrum is determined by orientation selection, which is based on the diversity of stored energy in differently oriented crystals. Apart from the cube component, which typically appears in metals with FCC crystal structure, weak γ-fiber orientations, P, Goss, Q (\{013\}\{231\}), or a ~22° rotated cube (CH) texture components tend to evolve in the recrystallized matrix of the investigated materials (Figs. 3 a–c). These orientations originate from diverse annealing phenomena and accounting for strain mode heterogeneities in the vicinity of non-deformable particles [2, 7] or texture evolution in the shear-bands [8], the recrystallization models are capable of explaining the evolution of mentioned orientations. The modelling issues, are out of the scope of this paper, but it should be underlined that the local events such as nucleation in the shear bands or particle stimulated nucleation [2], which takes place within a particle deformation affected zone, have a great impact on both qualitative and quantitative characteristics of texture evolved during final annealing.

Fig.3. Recrystallization textures of investigated materials annealed at 550°C: a) material A; b) material B1; c) material B2.
3.4. Plastic strain ratio

As it was already claimed in the previous studies [1, 9], the crystallographic texture strongly influences the anisotropy of plastic yielding, and therefore, texture control during TMP is of key importance. Both normal ($\bar{q}$) and planar ($\Delta q$) anisotropy appear to show evidence whether a given material has appropriate deep drawing quality (ideally, $\bar{q}$ should be above 0.5 and $\Delta q=0$). The assessment of plastic strain ratio (see Fig. 4) in the recrystallized materials A, B1 and B2 (shown in Fig. 3), is performed by employing a crystal plasticity model Alamel. This type of Taylor-type homogenization approach is capable of simulating the plastic anisotropy with a high accuracy [10] for a vast variety of textures. The investigated recrystallization textures (Fig. 3) reveal noticeable quantitative varieties. Since the ODFs of materials B1 and B2 are weaker in terms of texture intensity, the corresponding $q$-value profiles reveal less anisotropy, compared to material A, which shows a pronounced V-shaped profile. Although the $\bar{q}$ and $\Delta q$ values are far from ideal, the current study clearly shows that the proper choice of TMP parameters is capable of changing the behavior of a material during deep drawing operations.

4. Conclusions

Examination of texture evolution during deformation and recrystallization suggests that a broad spectrum of orientations aligned along the $\alpha$ and $\beta$ fibers in the deformed state, obtained after relatively small straining levels (30-47%), is capable of ensuring qualitative diversity in the recrystallized matrix. By way of contrast, relatively large quantity of cube component embedded in the deformation matrix, characterized solely by the $\beta$ fiber orientations account for stronger recrystallization texture.

The emerged weak textures are of particular interest due to enhanced formability characteristics while the strongly textured materials ensure strong planar anisotropy which is detrimental for deep drawing.

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