Influence of Mg Content on Texture Development during Hot Plain-Strain Deformation of Aluminum Alloys

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Abstract: The study addresses the effect of magnesium and other alloying elements on rolling “β-fiber” texture formation during hot deformation of aluminum alloys. For the study, flat cast ingots from three aluminum alloys with variable magnesium content were deformed in a Gleeble testing unit with different parameters of thermomechanical treatment. Immediately after completion of deformation, the samples were quenched using an automatic cooling system and the microstructure and crystalline texture was analyzed by optical microscopy and X-ray analysis. The analysis demonstrated that an increase in alloying components, magnesium in particular, leads to an increase in brass-type texture and a decrease in S and copper-type texture. The reason was that the simulation of the deformation texture development revealed a great contribution of impurity atoms rather than the decrease in stacking fault energy.

Keywords: aluminum alloys; deformation; magnesium content; texture; simulation

1. Introduction

The improvement of casting and thermomechanical treatment practices is one of the most demanding aspect of modern material science [1–6]. Combination of high plasticity, fairly low density and reasonable strength makes aluminum alloys indispensable in multiple present-day industries [7–10]. Still, one of the main problems associated with the processing of aluminum alloys is the anisotropy of their physical and mechanical properties. Anisotropy is caused by the crystallographic texture formed during thermomechanical processing [11]. Significant reduction of anisotropy can be achieved by a careful selection of specific texture components, e.g., balancing the cube texture, giving high plasticity at 0°/90° to the rolling direction with the “β-fiber” texture of plane strain deformation (i.e., rolling) causing high plasticity at 45° [11]. The β-fiber itself is composed of three texture components (see Table 1 below), with the central S component giving maximum plasticity at 45° to the rolling direction.
Table 1. Euler angles of ideal texture components including “β-fiber” [12].

| Component | Miller Indices [hkl](uvw) | Euler Angles $\phi_1$ $\Phi$ $\phi_2$ |
|-----------|----------------------------|----------------------------------------|
| Cube      | [001](100)                 | 0° 0° 0°/90°                           |
| Cube RD   | [013](100)                 | 0° 22° 0°/90°                          |
| Cube ND   | [001](310)                 | 22° 0° 0°/90°                          |
| Goss      | [011](100)                 | 0° 45° 0°/90°                          |
| Bs        | [011](211)                 | 35° 45° 0°/90°                          |
| Cu        | [112](111)                 | 90° 30° 45°                            |
| S         | [123](634)                 | 60° 35° 65°                            |

The effect of the remaining two components on anisotropy has a more complicated nature. It can be illustrated by the earing profile, i.e., the non-uniform height and thickness of the wall after deep drawing of a cup from a cylindrical workpiece. Peaks are formed in the areas with higher plasticity along the rolling direction, while valleys form in the areas with lower plasticity. The authors of [13] showed that the earing profile and, as a result, anisotropy for each of the three β-fiber components is individual. Therefore, it is obvious that efficient anisotropy reduction will be possible by an individual texture component management, rather than controlling the overall volume of deformation texture.

Besides adequate control of the deformation process and the resulting distribution of β-fiber components, it is the management of the corresponding recrystallization processes that defines the quality of aluminum semi-products, based on final grain size and orientations and resulting mechanical properties. Here, the rolling texture component “S” has a high influence on the cube recrystallization texture formation via oriented growth due to its preferred 40° orientation relationship along the crystallographic direction <111> relative to the cube texture [14]. Therefore, the generation of an exact S texture component quantity is critical.

One of the factors influencing the distribution of texture components within the β-fiber is the stacking fault energy. For example, pure copper with medium stacking fault energy level develops as pronounced “β-fiber” rolling texture—named “copper texture” [12]. The main texture in low stacking fault energy FCC metals such as α-brass formed during plain-strain (e.g., rolling) deformation is also called “brass texture” [12,15,16]. Some of the recent studies [17] established a more continuous transition path from Cu-type texture to Bs-type texture and an orientation distribution “balance center” is identified, depending on SFE. High SFE values cause its shift towards the copper-type texture, while low values shift it towards the brass-type texture. The effect of a high value in SFE on deformation strengthening is caused by stimulating transverse and double transverse dislocations to glide, enabling dislocations to bypass obstacles. Such an effect reduces the intensity of deformation strengthening (strain hardening). During the simulation of plastic deformation by shear along slip planes, such a drop in strength intensity means a drop in critical shear stress as a function of shear strain between slip planes. In Reference [17], it is suggested that the intensification of strain hardening results in the activation of a larger number of slip planes for the same type of crystal deformation. Some slip systems may be favorably oriented, while others are less favorable for glide initiation, a behavior that is numerically expressed by the Taylor factor.

If there is no strain hardening, the glide pattern would follow the most favorably oriented slip systems with minimum shear stress action. In this case, the crystal would reach the theoretically stable position at relatively low strain. However, due to deformation strengthening, slip on the systems, which are favorably oriented with respect to the applied external force, can be less energetically favorable compared to slip on other slip systems with less strain hardening. A larger number of slip systems is activated and some orientation dispersion advances at the same strain level, resulting in less pronounced copper-type texture and relatively larger fraction of α-fiber textures, including the brass-type texture, which is attributed to both the α- and β-fiber. As mentioned in several studies [15–17],
the evolution of the deformation texture starts with the development of an α-fiber texture, which consists of multiple texture components in the Euler angle space between the Goss texture and Bs texture, followed by the development of the β-fiber texture as deformation progresses.

As far as aluminum is concerned, there are various options for the distribution of texture components. This strongly depends on the content of alloy elements, specifically magnesium. In industrially pure aluminum alloy Al 99.5 (AA1050) the β-fiber intensity is higher on the copper side (112)<111> ($\varphi_2 = 45^\circ$) and shifts at higher strains towards the S texture (123)<634> ($\varphi_2 = 65^\circ$), whereas the brass component (110)<112> ($\varphi_2 = 0^\circ / 90^\circ$) develops relatively low [18]. On the contrary, the Al-Mg alloy AA5182, having high (up to 5%) magnesium content, reveals a shift of the texture towards the brass component [19]. This texture pattern change is mainly associated with the effect of magnesium to maximally decrease the stacking fault energy. In pure aluminum, it reaches 174 mJ/m², and in the alloy with 6.28% magnesium content, it reaches 151.6 mJ/m² [20].

Thus, magnesium content increase has a strong effect on texture composition during plane hot deformation and subsequent recrystallization (due to oriented growth mechanism influence). Therefore, accounting for magnesium effect on texture evolution during aluminum alloys’ thermomechanical treatment presents one of the key factors of the product anisotropy reduction. It should be noted that the listed above papers on aluminum just state the significant effect of magnesium on texture composition emerging during deformation. However, there are no detailed studies of quantitative evaluation of magnesium effect on β-fiber texture distribution. At the same time, there are studies providing detailed quantitative evaluation of Zn effect on deformation texture distribution in silver [21] and copper [22] alloys. The authors of [23] evaluated the Mn effect on deformation texture composition. Likewise, there are no studies accounting for magnesium effect during deformation textures’ modeling. However, similar studies were conducted for copper alloys with Mn additions [24]. Adequate texture composition control during materials’ thermodynamic processing is not possible without proper understanding of the quantified magnesium effect on aluminum alloys’ deformation texture composition as well as specific features of this factor modeling.

The objective of this study is to investigate the effect of magnesium content on the development of the key β-fiber texture components during hot deformation. In addition, effective methods of this factor accounting during deformation texture modeling shall be demonstrated.

2. Experiment Description

To study the effect of magnesium content on texture formation during thermomechanical treatment, plain-strain deformation tests were performed on three representative aluminum alloys: Al95, AlMg1 and AlMg4. The exact alloy compositions are presented in Table 2 below.

**Table 2.** Chemical composition of the studied alloys.

| Alloy      | Fe    | Si    | Mn    | Cr   | Ti   | Al   | Cu   | Be   | Mg   | Zn   | Impurities                      |
|------------|-------|-------|-------|------|------|------|------|------|------|------|---------------------------------|
| Al95       | Below 0.03 | Below 0.03 | -     | -    | -    | 99.95 | Below 0.015 | -    | -    | -    |                                |
| Al-Mg1     | Up to 0.7 | Up to 0.3 | Up to 0.2 | Up to 0.1 | - | 97.15–99.5 | Up to 0.2 | - | 1 | Up to 0.2 | others, each 0.05; total 0.15 |
| Al-Mg4     | Up to 0.5 | Up to 0.4 | 0.2–0.7 | 0.05–0.25 | Up to 0.15 | 93.15–96.23 | Up to 0.1 | 0.0002–0.0005 | 4.4 | Up to 0.25 | others, each 0.05; total 0.15 |

Al95 has no magnesium, Al-Mg1 has about 1%, and Al-Mg4 has 4.4% magnesium content. The samples were cast and homogenized at:

- 600–615 °C during 6 h for Al95;
- 595–610 °C during 6 h for Al-Mg1;
• 485–500 °C during 8 h for Al-Mg4.

Prior to deformation, the dimensions of the initial samples were as follows: length $a = 30$ mm, width $b = 35$ mm, thickness $h = 10$ mm.

Table 3 lists plain-strain deformation temperatures and rates for the different alloys Al-95, Al-Mg1 and Al-Mg4. The degree of rolling strain was 80% for all modes.

Table 3. Plain-strain deformation temperature and rate modes for the studied alloys.

| Alloy | Mode ID | Plain-Strain Deformation Mode | Temperature, °C | Strain Rate, s$^{-1}$ |
|-------|---------|-------------------------------|-----------------|----------------------|
| Al95  | 91      |                               | 300             | 1                    |
|       | 92      |                               | 450             | 1                    |
|       | 93      |                               | 300             | 10                   |
|       | 94      |                               | 450             | 10                   |
| AlMg1 | 11      |                               | 300             | 1                    |
|       | 12      |                               | 450             | 1                    |
|       | 13      |                               | 300             | 10                   |
|       | 14      |                               | 450             | 10                   |
| AlMg4 | 41      |                               | 300             | 1                    |
|       | 42      |                               | 450             | 1                    |
|       | 43      |                               | 300             | 10                   |
|       | 44      |                               | 450             | 10                   |

Figure 1 illustrates the test device for the deformation which was close to plain strain. TC are thermocouples—their locations in the sample and in the tool are indicated with arrows. Trapezoidal hammers are located at right angles to the rolling direction (see Figure 1). The hammer head part’s width ($w$) is 10 mm.

The samples were heated in low vacuum up to deformation temperatures, listed in Table 3, at 3 °C/s heating rate. Slight compression (<5 kN) during heating ensured uniform contact between hammers and samples along the entire contact area. After the preset temperature was reached, followed by 3 min isothermal holding time, the sample was compressed to the defined strain degree with the preset strain rate. One second after deformation completion, the sample was cooled to room temperature by water flushing. Specimens were cut out from the workpiece middle section and prepared for grain structure and texture examination by optical microscopy and X-ray texture analysis.

The microstructure was examined using optical microscope Axiovert 40 MAT (Carl Zeiss AG, Oberkochen, Germany). Microsections of specimens were prepared by cutting out, mechanical grinding, polishing and electrolytic polishing in fluoroboric acid electrolyte,
having the following composition: boric acid—11 g; fluorhydric acid—30 mL; distilled water—2200 mL.

Deformed samples cut out from the middle sections were used for texture determination by direct pole figure measurement with the survey plane perpendicular to the direction of compression, exerted by the hammers. The textures were measured by four incomplete pole figures {111}, {200}, {220} and {311}, obtained by the “diffraction” method, using X-ray diffractometer DRON-7 (Bourevestnik JSC, Saint-Petersburg, Russia) in CuKα radiation. The incline angle range α was 0° to 70° and the rotation angle range β was 0° to 360° with 5° increments each. The intensity drop in the pole figure periphery due to defocusing effect was compensated by correction factors, calculated by accounting for the pole figures’ X-ray photography conditions. The orientation distribution function (ODF) was calculated from the measured pole figures, presented as a superposition of a large number (2000) of standard distributions with identical low scattering. The same orientation distribution function (ODF) was used to calculate complete pole figures and inverse pole figures for three mutually perpendicular sample directions: normal to the rolling plane, rolling direction and transverse direction.

In addition, the chemical composition of the aluminum matrix and large intermetallic particles after treatment according to modes 91, 11 and 41 shown in Table 3 was investigated by electron microscopy. The JEOL 6390 A scanning electron microscope (SEM) (JEOL Ltd., Tokyo, Japan) was used for this purpose. The sample preparation technique consisted of mechanical grinding, polishing and electropolishing. Electropolishing of the specimen was carried out at 85–110 °C and the voltage of 10–30 V in electrolyte of the following composition: 500 mL of H₃PO₄, 300 mL of H₂SO₄, 50 g of CrO₃ and 50 mL of H₂O. As mentioned earlier, rolled FCC lattice metals develop both copper (Cu) and brass (Bs)-type orientations. Rolling textures of high SFE metals and alloys show pronounced Cu orientations, while low SFE materials form strong Bs textures [24].

The deformation texture simulation model, described in [25], was applied for modeling the texture evolution during hot deformation. The incorporated modifications enabled to consider the effect of magnesium content on the evolution of the major β-fiber texture components due to the decrease in stacking fault energy. As stated in the Introduction, the evolution of texture intensity depends on the nature of deformation strengthening, thus requiring some modifications, introduced in the model in reference [25]. The model in reference [25] accounts only for the ratio, relating critical shear stress and shear strain rate, and is written as:

$$\gamma = \gamma_0 \left( \frac{\tau}{\tau_c} \right)^n,$$

where $\gamma$ is actual shear strain rate, $\gamma_0$ is basis shear strain rate value, $\tau_c$ is actual critical shear stress value, $\tau_c$—basis critical shear stress value.

This study offers the following ratio for $\tau_c$ calculation:

$$\tau_c = \tau_c \varepsilon_{ac}^{-n},$$

where $\varepsilon_{ac}$ is the parameter, characterizing deformation strengthening, (measure of crystalline structure imperfections (dislocations) accumulated in α-type slip system).

$$n = n_0 \left( \ln(\dot{\varepsilon}) + \frac{Q}{RT} \right)$$

where $\dot{\varepsilon}$ is macroscopic sample strain rate, $n_0$ is basis $n$ factor value within 0.3–0.5 range, $Q$ is plastic deformation activation energy, taken as 140,000 J/m³ for Al-95, 165,000 J/m³ for AlMg1, 175,000 J/m³ for AlMg4, $R$ is absolute gas constant.

The parameter $\varepsilon_{ac}$ is of generalized nature, accounting for various strengthening components: an SFE-specific ability of dislocations to bypass obstacles, dislocations’ damping in solid solution, dislocations’ retardation by forest dislocations and impurity atmosphere.
The parameter $\epsilon^{\alpha}_{ac}$ is calculated using shear strain increments $d\epsilon^{\alpha} = \gamma^{\alpha} dt$, where $dt$ is a time increment:

$$
\begin{bmatrix}
\begin{array}{cccccccc}
    d\epsilon^1_{ac} \\
    d\epsilon^2_{ac} \\
    d\epsilon^3_{ac} \\
    \vdots \\
    d\epsilon^{10}_{ac} \\
    d\epsilon^{11}_{ac} \\
    d\epsilon^{12}_{ac}
\end{array}
\end{bmatrix}
= 
\begin{bmatrix}
    q & p & p & r & r & \cdots & r & r & r \\
    p & q & p & r & r & \cdots & r & r & r \\
    p & p & q & r & r & \cdots & r & r & r \\
    \vdots & \vdots & \vdots & \vdots & \vdots & \ddots & \vdots & \vdots & \vdots \\
    r & r & r & r & r & \cdots & r & q & p \\
    r & r & r & r & r & \cdots & r & p & q \\
    r & r & r & r & r & \cdots & r & p & q
\end{bmatrix}
\begin{bmatrix}
    d\epsilon^1 \\
    d\epsilon^2 \\
    d\epsilon^3 \\
    \vdots \\
    d\epsilon^{10} \\
    d\epsilon^{11} \\
    d\epsilon^{12}
\end{bmatrix}
$$

(4)

Considering 12 slip systems, vector $d\epsilon^{\alpha} \alpha = 1.12$ was converted to vector $d\epsilon^{\alpha}_{ac} \alpha = 1.12$ using ratio (3). In ratio (3), $q$ is the factor accounting for an increase in dislocation density in this particular slip system, dislocations’ damping and dislocations’ retardation by impurities atmosphere. $p$ is the factor accounting for slip system dislocation interactions in intersecting planes. This factor is responsible for dislocations’ retardation due to the intersection with other dislocations and the formation of Lomer-Cottrell jogs.

3. Results and Discussion

The as-cast Al-95 has a coarse dendritic columnar structure with elongated grains up to 2 mm with 300 $\mu$m thickness, as shown in Figure 2a. However, more equiaxed grains with up to 1 mm size are located in the center. In the area of the deformed sample adjacent to the die, a retained dendritic structure zone is observed, which is due to locally impeded deformation, obstructing structure workover, as shown in Figure 2b [26]. It should be noted, however, that both columnar and equiaxed grains align in the horizontal direction and are in line with their plastic strain. At the same time, severely directionally elongated grains are observed in the center. With an increase in the deformation temperature to 450 °C in the alloy Al-95, the presence of fine newly recrystallized grains, with the size of several tens of microns are observed as shown in Figure 2c. The higher the deformation temperature, the greater the number of such formations. This can be explained by the inability to generate the microstructure instantly, since it takes several seconds to develop it after the onset of deformation. Due to the high temperature of the sample and therefore, high migration rate of large-angle boundaries [23], this is sufficient for the growth of the most desired texture components. In AlMg1 alloy, the dendritic structure has much smaller sizes and ranges, from 600 $\mu$m to 800 $\mu$m on average; besides, it is far more equiaxed, as shown in Figure 2d. Such conditions can be explained by a high content of impurities, maximizing undercooling between the liquid and the surface of nuclei, formed during crystallization, thus facilitating recrystallization [27]. During upsetting, an impeded deformation zone is also observed along the workpiece boundaries with elongated grains in the center, as shown in Figure 2e. Additionally, effective methods of this factor accounting during deformation texture modeling shall be demonstrated.
With Al-Mg1 temperature increase, as shown in Figure 2f, much more new grains’ nuclei are observed, which is associated with aluminum alloys’ recrystallization kinetics boost, caused by magnesium content increase [28].

Compared to the Al-Mg1 alloy, Al-Mg4 has a finer as-cast structure with average grain size ranging from 200 µm to 300 µm, as shown in Figure 2h, which is explained by even higher content of Mg as well as the presence of Mn, Cr, Ti and other elements. After thermomechanical treatment using a Gleeble unit, an impeded deformation zone is also observed, and elongated grains occur in the center. It should be noted that none of the alloys change the pattern of grain structure distribution over the deformation center thickness under the variation of temperature-rate parameters of the deformation, as shown in Figure 4h. In all cases, an impeded deformation zone with retained dendritic structures and elongated grains are observed in the workpiece center. At 450 °C, significantly larger amount of recrystallization nuclei is observed in this alloy, compared to A-95 and Al-Mg4, which is illustrated in Figure 2i.

Electron microscopy results demonstrated a negligible amount of intermetallic particles in Al-95 alloy, as shown in Figure 3a. Aluminum matrix chemical analyses revealed that it is almost completely composed of aluminum with a small fraction of silicon (within margin of error). This is explained by high purity of this aluminum alloy. Therefore, it
is rather difficult to detect the locations with other dissolved elements using chemical analysis by electron microscopy method. The results of the analysis enable the conclusion that the presence of such elements does not produce notable effect on texture evolution. Al-Mg1, as shown in Figure 3b, has a specific for such alloys’ ALFe6-type particles, and others contained in this alloy’s transition elements, e.g., Cu, dissolve in them. The latter attract Mg, which can be traced in such particles. However, main Mg content is observed in aluminum matrix, while other elements’ content lies within determination error margin. Thus, it may be concluded that it is magnesium which plays the key role in strengthening, hence texture evolution. AlFe6 particles with magnesium traces are observed in Al-Mg4, as shown in Figure 3c. Magnesium in similar Al(FeSi)6 particles was also observed in magnesium-rich 5454 alloy, which was explained by examination method imperfections (i.e., aluminum matrix portion trapping in addition to particles for analysis). In addition, Al3Mg2 particles are observed, they partially withdraw magnesium from solid solution and its content is slightly over 3%, while the content of other impurity elements in solid solution is within measurement error margin. Thus, it is not the other elements, but magnesium will produce the major effect on crystalline texture development during hot deformation.

![Electron microscopy results for the following alloys: (a) Al-95; (b) Al-Mg1; (c) Al-Mg4.](image)

Figure 3. Electron microscopy results for the following alloys: (a) Al-95; (b) Al-Mg1; (c) Al-Mg4.

It may be caused by Al3Mg2-type particles forming due to nonequilibrium crystallization conditions and magnesium concentration exceeding 1 (it is proven by electron microscopy results). The higher the magnesium content is, the less effect will its further build-up have on its increasing content in the solid solution (hence strengthening), it will be more represented by Al3Mg2-, Mg2Si-type intermetallic particles. Besides, it may be assumed that until 1% concentration is reached, texture formation behavior changes due to Cu and S abundance decrease and Bs increase, followed by equalization of slipping along all planes. However, fine studies, deploying scanning microscopy methods, are required for this aspect of the investigation.

In Figure 4 the location of the specific for aluminum alloys texture components are shown in the Euler space (for $\varphi_2 = 90^\circ$, $\varphi_2 = 45^\circ$ and $\varphi_2 = 45^\circ$ sections). The values of the Euler angles for Figure 4 is taken from Table 1.
The deformation of commercially pure aluminum with a deformation rate of 1 s$^{-1}$ and at 300 °C, results in the development of a strong pronounced Cu and S deformation texture (Figure 5a). This perfectly matches the microstructure analysis, namely the presence of severely elongated (i.e., deformed) grains. Besides, a moderately pronounced brass deformation texture, as well as a small number of Goss textured grains are also present. The latter can develop both during deformation and recrystallization. Deformation with 1 s$^{-1}$ rate and at 450 °C (Figure 5b) results in a moderately pronounced Cu deformation texture, with the Bs component being totally absent. This is associated with the temperature rise causing a reduction in strengthening (strain hardening) in all main slip planes specific to face-centered cubic lattice metals [29]. At a strain rate increase to 10 s$^{-1}$ at 300 °C (Figure 5c), the Goss texture is better pronounced, the volume of Cu and S also increases, while the Bs texture is greatly diminished at such thermomechanical treatment modes. This is explained by the fact that strengthening increases with strain rate, resulting in a maximal deployment of slip systems, specific to Cu and S texture components. At a rise in temperature to 450 °C at 10 s$^{-1}$ (Figure 5d), the strain rate decreases the volume of Cu and S and, in some way, boosts the brass texture (Bs) volume, while Goss disappears completely.

Such a pattern is explained by a decrease in strain hardening with temperature rise, while the brass-type texture-associated slip systems once again become favorable. In addition, a small volume of cube texture appears, as the result of a small number of fine recrystallized grains. The general feature, specific for this alloy, consists in the deployment of slip systems, associated with Cu and S texture components and—to a lesser degree—with the brass texture component Bs. It should be noted that commercially pure aluminum features a high stacking fault energy, which brings it closer to copper with respect to texture formation.

During AMg1 hot deformation at 1 s$^{-1}$ strain rate and 300 °C temperature (Figure 6a), all three β-fiber texture specific components, i.e., Cu, S and Bs, are observed. At the same time, unlike the previous alloy, it shows a strongly pronounced Bs texture component, while the other two rolling texture components are less pronounced. A temperature rise causes some dissipation of the Bs texture volume, while the amount of Cu and S slightly increases (Figure 6b). A strain rate increase (Figure 6c) to 10 s$^{-1}$ causes an increasing in volume of S and especially Cu rolling texture components. At 10 s$^{-1}$ strain rate and 450 °C (Figure 6d), no clearly defined deformation textures are observed, and all three rolling texture components are very weak.

Such conditions can be explained by a stronger effect of temperature on strength in such alloys, causing far more rapid strength decrease with temperature rise, compared to Al-95 alloy. In addition, due to the high temperature and high strain rate, a large number of recrystallized grains is also present. Some of them will contribute to cube texture, and others to a random texture constituent. Thus, the majority of elements, occurring in this alloy, will stimulate the development of multiple second-phase particles with the tendency of forming new recrystallized nuclei by particle-stimulated nucleation (PSN), having a mostly random texture constituent [30,31].
A strain rate increase leads to an increase in the Zener-Hollomon value, hence an increase of energy, required for the recrystallization process. Therefore, a rate increase to 450 °C temperature will facilitate the dispersion of textures’ components, rather than their strengthening, as might be expected. At the same time, the grain boundary mobility in this alloy at 300 °C temperature is not sufficient for significant grain growth during the metal cooling period. Thus, regardless of the high Zener-Hollomon parameter value in all thermomechanical treatment modes, recrystallization at 300 °C temperature does not have significant effect on the texture’s composition. It should be pointed out, unlike commercially pure aluminum, that mainly brass texture-associated slip systems are deployed in Al-Mg1 alloy. Strengthening, stimulated by the strain rate increase, starts deploying slip systems, associated with the Cu and S components. Strengthening related to temperature, strain rate and many impurity elements, especially, magnesium, will have a different influence on the activation of various slip systems.

Figure 5. ODF, aluminum 99.95% (for composition refer to Table 1) compressed under plain strain to 80% engineering strain (a) strain rate 1 s⁻¹, temperature 300 ºC, (b) strain rate 1 s⁻¹, temperature 450 ºC, (c) strain rate 10 s⁻¹, temperature 300 ºC (d) strain rate 10 s⁻¹, temperature 450 ºC.
At 1 s\(^{-1}\) strain rate and 300 °C, three well-formed major β-fiber components are observed in Al-Mg4 alloy (Figure 7a). As the temperature rises to 450 °C (Figure 7b), the volume of all three deformation texture components slightly decreases, which is explained by strain hardening. As the strain rate rises to 10 s\(^{-1}\) (Figure 7c), the volume of the three major rolling texture (β-fiber) components also increases. At 10 s\(^{-1}\) strain rate and 450 °C (Figure 7d), the total amount of β-fiber texture decreases compared to the mode, illustrated in Figure 3c. This can be also attributed to the initiation of the recrystallization process. Note that all deformation texture components are less pronounced in this alloy, compared to Al-Mg1. This is explained by the high Mg content of Al-Mg4, which is close to AA5182 alloy. The latter alloy fully recrystallizes within 10 s under these temperature-rate strain modes, but in all major cases, recrystallization duration varies from 50 to 100 s\(^{-1}\). Due to a larger amount of newly formed 10–30 μm grains, observed in Al-Mg4 (Figure 4i), the deformation texture is weaker. At the same time, like in Al-Mg1, this alloy has a clearly defined Bs texture.

**Figure 6.** ODF, Al + 1% Mg (composition, see Table 2) compressed under plain strain to 80% engineering strain (a) strain rate 1 s\(^{-1}\), temperature 300 °C, (b) strain rate 1 s\(^{-1}\), temperature 450 °C, (c) strain rate 10 s\(^{-1}\), temperature 300 °C (d) strain rate 10 s\(^{-1}\), temperature 450 °C.
Figure 7. ODF, Al + 4% Mg (composition, see Table 3) compressed under plain strain to 80% engineering strain, material initially as cast, strain rate (a,b) 1 s\(^{-1}\); (c,d) 10 s\(^{-1}\); temperatures (a,c) 300 °C; (b,d) 450 °C.

For clarity, the various β-fiber textures’ distribution within 45° < \(\varphi_2\) < 90° interval is illustrated in Figure 8.

It is obvious, from Figure 8a, that the copper-type texture and similar components are abundant in Al-95 under all processing modes. Their intensity gradually declines to 60°, i.e., to the S texture. It is followed by a sharp decline in texture intensity until, finally, the same values at \(\varphi_2 = 90°\) are reached, which is also characteristic for the Bs component. In the Al-Mg1 alloy, the texture intensity declines within 45° < \(\varphi_2\) < 60° interval, i.e., around the C and S components; at the same time, brass-type orientation intensity builds up. As mentioned earlier, this is due to strain hardening, which impairs the advantage of gliding on the slip systems, attributed to copper- and S-type textures. At the same time, the building up of the brass-type textures is observed, as gliding on its systems makes such strengthening level more favorable.

All texture intensities decline in the Al-Mg4 alloy, which is explained not only by better recrystallization capability of this alloy, but by strengthening reaching the values, which
disable gliding in slip systems, associated both with S and Cu components, preferring the brass-type textures related slip systems. It should be stressed that increase in strain hardening is caused not only by an increase in SFE, but also by an increase in magnesium impurity atoms, dissolved in solid solution [32]. It should also be considered that the SFE of pure aluminum (annealed) is equal to 174 mJ/m$^2$, while the SFE of the alloy with 6.28% magnesium content is 151.6 mJ/m$^2$. In other words, even deformed alloys with extremely high magnesium content maintain a high level of SFE, compared to other materials [20].

Considering the data, presented by this source, it may be assumed that Al-95 alloy (with account for its high purity) energy will be equal to 177 mJ/m$^2$, Al-Mg1—168.0 mJ/m$^2$, Al-Mg4—154.7 mJ/m$^2$, thus SFE difference for two extreme alloys will be slightly above 22 mJ/m$^2$.

Therefore, opposite to copper and copper-based alloys, SFE cannot be considered as a prevailing factor in shifting towards Bs-type textures when referred to aluminum alloys.

The calculations with variable values: $q = 0.25–1.5$, $p = 0.1–0.5$ and $r = 0.25–0.7$ demonstrate a greater effect of impurity atoms. The calculations show the best agreement for Al-95 alloy at $q = 0.32$, $p = 0.19$, $r = 0.32$; for Al-Mg1 alloy at $q = 0.51$, $p = 0.24$, $r = 0.43$; for Al-Mg4 at $q = 1.1$, $p = 0.36$, $r = 0.54$, thus, the impurity atoms-related factor $q$ has a greater effect than the SFE-related factors $p$ and $r$.

Figure 9 shows the results of ODF simulations of thermomechanical treatment for the three studied alloys at strain rate of 10 s$^{-1}$ and 300 $^\circ$C. For Al95 alloy, as shown in Figure 3a, the calculated and the observed ODF demonstrate a high convergence for the S and Cu components. At the same time, the Bs texture components are slightly overestimated, while Goss texture values are underestimated. The AlMg1 alloy has less pronounced Cu and S textures, while a minor Bs texture fraction is observed.

Note that in the model, unlike in the actual process, all deformation texture components are less clearly developed. Al-Mg4 has three sufficiently clearly defined components with a dominating cube texture, which accurately agrees with the observed pattern.
It is obvious, from Figure 8a, that the copper-type texture and similar components manifest themselves clearly at 300 °C temperature. To further improve the calculation accuracy requires a further in-depth investigation of physical effects on matrix components (1), various dislocation movements and crystalline strengthening due to their interactions with various obstacles, represented by dislocation forest and Lomer-Cottrell jogs.

For the calculations, \( n_0 = 0.46 \) was assumed and \( \tau^c \) to be the same in all slip systems. For aluminum, it was taken as 1, based on the data in ref. [33], for the Al 1% Mg alloy \( \tau^c = 2.5 \) and for 4% magnesium alloy \( \tau^c = 4.0 \). The results are shown in Figure 10.

As can be seen in Figure 8, the calculation matches the conditions, under which the data in Figure 7 were obtained; in general, the calculation adequately reflects the transformation of texture components in case of modifications in alloy composition and deformation parameters. However, it underestimates the intensity values of the copper and S-type texture components as well as of the brass-type texture; some inaccuracies manifest themselves clearly at 300 °C temperature. For Al-Mg1, inaccuracies consist in a slightly overestimated brass texture and underestimated copper texture. The best match is observed in Al-Mg4 alloy, emphasizing a greater effect of magnesium atoms and other components, dissolved in the aluminum matrix, on strain hardening (crystalline strengthening), compared to an increase in SFE. To further improve the calculation accuracy requires a further in-depth investigation of physical effects on matrix components (1), various dislocation movements and crystalline strengthening due to their interactions with various obstacles, represented by dislocation forest and Lomer-Cottrell jogs.
Magnesium content effect on the main texture components during different thermo-mechanical treatment modes is illustrated in Figure 11. At 300 °C, temperature and strain rate of 10 s\(^{-1}\) Cu and S textures abundance drops from 27 f(g) and 15 f(g) to 7 f(g) and 7.3 f(g) respectively with magnesium content increase. With increasing magnesium content, Bs component abundance grows from 0.7 f(g) to 9.5 f(g). At lower temperature, all orientations' abundance decreases, which is related to recrystallization effect and strengthening decrease. However, overall nature of magnesium effect on major texture components stays the same: with magnesium increase from 0% to 4% S and Bs abundance drops from 13 f(g) and 7 f(g) to 3.4 f(g) and 4.6 f(g) respectively. It shall be pointed out that the sharpest Cu and S orientations decrease and Bs abundance increase are observed with magnesium content ranging from 0% to 1%. It is followed by further texture components' abundance change; however, it has steadier behavior. The main cause could be incomplete magnesium solubility in aluminum matrix due to nonequilibrium crystallization conditions and magnesium content over 1%; in this case, Al\(_{3}\)Mg\(_{2}\)-type intermetallic compounds will be formed. The latter will be preserved during homogenization, though they will become smaller in size. Besides, it may be assumed that until 1% concentration is reached, texture formation behavior changes due to Cu and S abundance decreases and Bs increase, followed by equalization of slipping along all planes. However, fine studies, deploying scanning microscopy methods, are required for this aspect of the investigation.

**Figure 10.** Model-based calculation of texture components’ re-distribution with increasing deformation strengthening: (a) Aluminum Al 99.5 (AA1050); (b) Al-Mg1; (c) Al-Mg4.

**Figure 11.** Magnesium content effect on the main β-fiber texture components’ abundance: (a) 10 s\(^{-1}\), temperatures 300 °C; (b) 10 s\(^{-1}\), temperatures 450 °C.

4. Conclusions

The study showed that an increase in the content of magnesium and other alloying elements has a significant effect on the development of rolling texture (β-fiber) components during hot deformation. In commercially pure aluminum, mainly copper and S texture-
related slip planes are deployed. As the magnesium content increases, the intensities of these components deplete, and a pronounced brass texture appears. There are two explanations for this fact: Firstly, the stacking fault energy (SFE) decreases. This factor undoubtedly contributes to the stain hardening, since the SFE decreases for magnesium-rich alloys by 10–15%. The simulation demonstrates the greater effect of impurity atoms on crystallographic strengthening, i.e., strain hardening. To further improve the model accuracy requires in-depth analysis of strengthening matrix components’ relationship with physical phenomena, emerging during dislocations’ movements and interactions with various obstacles.

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