Microstructure, texture, and strain-hardening behavior of extruded Mg–Gd–Zn alloys

M.M. Hoseini-Athar a, R. Mahmudi a,*, R. Prasath Babu b, P. Hedström b

a School of Metallurgical and Materials Engineering, College of Engineering, University of Tehran, Tehran, Iran
b Department of Materials Science and Engineering, KTH Royal Institute of Technology, SE-100 44, Stockholm, Sweden

ARTICLE INFO

Keywords:
Mg–Gd–Zn alloy
Rare-earth element
Strain-hardening
Texture

ABSTRACT

The effect of Zn content on the microstructure, texture, mechanical properties and strain-hardening behavior of extruded Mg–2Gd–xZn (x = 0, 1, 2 and 3 wt%) sheets was investigated. Evaluation of texture revealed that while all of the alloys exhibited weak textures, the texture component was altered from a basal to a non-basal one by the addition of Zn. A typical transverse direction (TD) split texture with basal poles rotated about 40° from the normal direction (ND) toward TD was observed for the Zn-containing alloys, the effect being more pronounced at higher Zn contents. Furthermore, the Mg–2Gd–1Zn alloy exhibited the weakest texture due to solute drag imposed by co-segregation of Zn and Gd atoms at grain boundaries. Addition of Zn also resulted in a general increase in yield stress, ultimate tensile strength and elongation along the extrusion direction from 99 to 172 MPa, 178 to 263 MPa, and 25 to 35% for Mg–2Gd and Mg–2Gd–3Zn alloys, respectively. However, increasing Zn content was accompanied by an initial decrease in anisotropy of mechanical properties and strain-hardening behavior, followed by an increase at higher Zn contents. This was due to the difference of orientation of basal planes with regard to tension direction. As a result, lower yield stress, higher elongation and strain-hardening capacity was obtained along TD (with higher Schmid factor for basal slip) compared to ED. It was concluded that excellent mechanical properties and low anisotropy can be achieved in the Mg–2Gd–1Zn alloy.

1. Introduction

Application of magnesium alloys has been restricted because of their poor room temperature formability, originated from the limited number of slip systems in the hcp structure and development of strong basal texture during deformation [1]. It has been suggested that grain refinement and texture modification through alloying and deformation processes, are beneficial to formability of Mg alloys [2–4]. The crystallographic texture influences the deformation mechanism either by changing the resolved shear stress on basal planes (i.e. affecting Schmid factor), or promoting activation of non-basal slip systems through altering the grain orientation [5].

Addition of rare-earth (RE) elements to Mg alloys has been proposed as an effective approach to deal with the formability problem, through grain refinement and texture weakening. It has been reported that RE elements influence the texture development of Mg alloys during the recrystallization process, mostly through shear band nucleation (SBN), particle stimulated nucleation (PSN) and solute drag effects [6]. As a result, a relatively random texture with lower intensity and a broader distribution of the basal planes, oriented favorably for basal slip, is usually developed in the RE-containing Mg alloys [7]. For RE elements with high solubility in α-Mg solid solution, it is expected that the dominant mechanism affecting the texture evolution is solute drag imposed by the RE elements segregated at grain boundaries rather than Mg–RE particles [8]. Al-Samman and Li [9] showed that gadolinium (Gd) has the most pronounced effect on texture modification and improvement in the room temperature ductility and planar anisotropy among different RE elements. Kula et al. [10] have shown that Gd can improve both strength and ductility of the Mg–Gd solid solutions through more homogeneous deformation and prolonged strain hardening, which are caused by the activation of non-basal slip systems.

Zn is usually added to Mg–RE alloys to improve tensile strength and age hardening response. Zn is also known for its effect on texture and mechanical properties of Mg alloys through reducing the critical resolved shear stress (CRSS) of the prismatic (a) and pyramidal (c+a) slip by reducing the c/a ratio [11]. In addition, it has been reported that simultaneous presence of RE and Zn could lead to a more significantly weakened texture and a TD-split component, where basal poles are tilted.
from ND toward TD \[12\]. It has been shown that high ductility and excellent stretch formability can be achieved in Mg–Gd–Zn alloys, as a result of the formation of such a non-basal texture with low intensity \[13\]. Despite improved formability observed in the Mg alloys with the TD-split texture, it might be accompanied by more pronounced anisotropy of mechanical properties, e.g. yield stress and ductility \[12,13\].

Strain-hardening rate (SHR) is a major parameter influencing the formability of Mg alloys by delaying the onset of the deformation instability. Strain-hardening is associated with dislocation accumulation and annihilation. Dislocation tangling hampers further movement of other dislocations, causing further resistance to deformation \[14\]. Several studies have investigated the effect of alloying elements \[15\], second phase particles \[16,17\], grain size \[18\], twinning \[19\], temperature \[20\] and texture \[21\] on the strain-hardening behavior of Mg alloys. It has been demonstrated that in the binary Mg–Zn alloys, SHR and strain-hardening capacity are increased with increasing Zn content, due to the weaker basal texture and higher concentration of solute atoms in the α-Mg matrix \[14\].

Despite several studies on the texture and mechanical properties of Mg–Gd–Zn alloys, the effect of Zn content on textural evolution, mechanical properties, and strain-hardening behavior requires a more systematic investigation. Furthermore, since the addition of Zn can lead to mechanical anisotropy, the mechanism of anisotropy as well as the optimum Zn content to obtain enhanced mechanical properties and low planar anisotropy must be clarified. Therefore, the present study investigates the texture, mechanical properties and anisotropy of the Mg–2Gd–xZn (x = 0, 1, 2 and 3 wt%) alloys with an aim to elucidate the deformation mechanism and obtain the optimum Zn content.

2. Materials and methods

In the present study, Mg–2Gd–xZn alloys containing 0, 1, 2 and 3 wt % Zn were investigated. The chemical compositions of the studied alloys, obtained by ICP-AES method, are summarized in Table 1. The alloys were prepared by melting proper amounts of pure (99.9%) Mg and Zn and an Mg–20 wt% Gd master alloy at 1053 K in a graphite crucible under a protective covering flux. After pouring in a permanent mold preheated to 623 K, the as-cast plates were homogenized at 773 K for 10 h. Homogenized plates were extruded to 3-mm thick sheets under an extrusion ratio of approximately 6 at a temperature of 653 K. The sheet extrusion die is schematically shown in Fig. 1.

Microstructural characterization was carried out by a JEOL JSM–7800F scanning electron microscope equipped with energy dispersive x-ray spectroscopy (EDS). Metallographic specimens, sectioned perpendicular to the normal direction (ND) were prepared by grinding, polishing, and chemically etching in a solution of 4 g picric acid, 60 ml ethanol, 20 ml distilled water, and 20 ml acetic acid. Crystallographic texture was studied by using X-ray diffraction (XRD) pole figures obtained in a Rigaku X-ray diffractometer and electron back-scattered diffraction (EBSD) data provided by a Bruker e-flash 3D detector, taken from the extrusion direction (ED)–TD plane (Fig. 1). For EBSD analysis, specimens were first polished by diamond suspension and then electropolished in a solution of 80 ml ethanol, 8 ml HNO₃, and 20 ml acetic acid. EBSD data were analyzed using MTEX \[22\], considering a misorientation angle greater than 10° for defining the high angle grain boundaries (HAGBs).

Table 1

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Chemical composition (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Gd</td>
</tr>
<tr>
<td>Mg–2Gd</td>
<td>1.96</td>
</tr>
<tr>
<td>Mg–2Gd–1Zn</td>
<td>2.10</td>
</tr>
<tr>
<td>Mg–2Gd–2Zn</td>
<td>2.08</td>
</tr>
<tr>
<td>Mg–2Gd–3Zn</td>
<td>1.93</td>
</tr>
</tbody>
</table>

SEM micrographs of the extruded alloys are shown in Fig. 2. A fine equiaxed microstructure is observed for all of the studied alloys. Mean grain sizes of 12.2, 8.5, 9.2 and 9.6 μm were measured for the Mg–2Gd, Mg–2Gd–1Zn, Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, respectively. In addition to the α-Mg matrix, there exist some dispersed second phase particles in the microstructure. The volume fraction of these particles is relatively small, since the respective room temperature solubility levels of Zn and Gd in Mg are as high as 2 wt% and 1.5 wt% \[8\]. Nevertheless, addition of Zn increases the volume fraction of the second phase particles, so that the approximate values of 0.8, 2.7, 3.8 and 5.4 vol% are obtained for the Mg–2Gd, Mg–2Gd–1Zn, Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, respectively. According to EDS analysis, the compositions of the large precipitates are very close to those reported for the binary Mg-Gd and the ternary Mg–Gd–Zn phases \[23\]. Based on the EDS results, while the Mg–2Gd alloy contains only MgGd phase, the Zn-containing alloys include mostly MgGdZn phase with a very small amount of the MgGdZn phase.

EBSD orientation maps and misorientation angle distribution for the studied alloys are shown in Fig. 3. The microstructure of all alloys consists of equiaxed grains with a high fraction of high angle grain boundaries (HAGBs). However, there is a difference in the misorientation angle distributions, as manifested by a peak at ~30° in the Mg–2Gd alloy, which shifts to ~86° at higher Zn contents. The (0002) XRD pole figures (Fig. 4) reveal that for the Mg–2Gd alloy, a basal texture is
Fig. 2. SEM micrographs of (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys.

Fig. 3. EBSD orientation maps and misorientation angle distribution of (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys.

Fig. 4. 0002 XRD pole figures of (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys.
developed during extrusion, in which basal planes are mostly parallel to the ED–TD plane. Addition of Zn results in a split in the texture component, so that a typical “TD-split” non-basal texture with approximately 40° rotation of basal poles from ND toward TD is obtained. The so-called “TD-split” texture has been previously reported in some Mg–RE–Zn alloys [24]. Besides a change in texture component, addition of Zn also affects the peak intensity of the texture. As shown in Fig. 4, a drop in the peak intensity from 5.4 multiples of random distribution (mrd) to 3.1 mrd occurs by addition of 1 wt% Zn to the Mg–2Gd alloy. However, further increase in the Zn content leads to an increase in the peak intensity to 4.3 and 4.9 mrd for the Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, respectively.

3.2. Mechanical properties

Room temperature engineering stress–strain curves of the studied alloys are presented in Fig. 5. Mechanical properties obtained from the tensile tests are also summarized in Table 2. Room temperature plastic behavior of the alloys can be described by a power-law constitutive equation:

\[ \sigma = K \varepsilon^n \]  

where \( \sigma \), \( \varepsilon \), \( K \) and \( n \) are the stress, plastic strain, strength coefficient and strain-hardening exponent, respectively. Since the strain-hardening exponent is an indication of the uniform elongation, specimens with higher \( n \)-values also exhibit higher \( \varepsilon_f \).

As observed in Table 2, the Mg–2Gd alloy exhibits rather similar yield stress (YS), ultimate tensile strength (UTS) and elongation to fracture (\( \varepsilon_f \)) values along various directions. By the addition of Zn, a distinct yield point appears for the specimens tested along 45° and TD. Furthermore, Zn addition results in anisotropy in the YS and \( \varepsilon_f \) along different directions, while the UTS experiences a less significant variation. For the Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, YS values along ED are much higher than those obtained along TD. On the other hand, a higher \( \varepsilon_f \) is observed for the TD specimens compared to ED specimens. It should be mentioned that the mechanical properties along 45° is between those obtained for ED and TD specimens. In order to further investigate the distinct yield point in Mg–2Gd–3Zn, EBSD maps of the ED and TD specimens strained up to \( \varepsilon = 0.2 \), are shown in Fig. 6. It can be noted that several twins are seen in the specimen loaded along TD, while the specimen loaded along ED shows less twins. Evaluation of the orientations confirms that the twins are \{1012\} tension twins.

Since orientation of the basal planes (i.e. texture) has a major effect on mechanical properties, Schmid factor (SF) distribution for the basal slip as well as the mean SF values were calculated from the EBSD results and are demonstrated in Fig. 7. For the Mg–2Gd alloy, relatively similar distributions are observed along different directions. However, it is clear that increasing the Zn content leads to a marked difference in the SF distribution and mean SF values along ED, 45° and TD, so that for the Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, a significantly higher SF value is observed along TD.

3.3. Strain hardening behavior

In order to study the strain hardening behavior, Kocks–Mecking diagrams were plotted as the strain-hardening rate (SHR) against net flow stress (\( \sigma - \sigma_{0.2} \), where \( \sigma_{0.2} \) is the 0.2% proof stress based on true stress–strain curves), as shown in Fig. 8. In this regard, the SHR (\( \theta \)) is calculated as:

\[ \theta = \frac{d\sigma}{d\varepsilon} \]  

As can be seen, all specimens show an initial steep decrease in the
Table 2
Room temperature tensile properties of the studied alloys.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>ε (%)</th>
<th>π</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg-2Gd</td>
<td>99</td>
<td>102</td>
<td>106</td>
<td></td>
</tr>
<tr>
<td>Mg-2Gd-1Zn</td>
<td>146</td>
<td>151</td>
<td>158</td>
<td>30</td>
</tr>
<tr>
<td>Mg-2Gd-2Zn</td>
<td>155</td>
<td>123</td>
<td>117</td>
<td>32</td>
</tr>
<tr>
<td>Mg-2Gd-3Zn</td>
<td>172</td>
<td>133</td>
<td>121</td>
<td>35</td>
</tr>
</tbody>
</table>

Fig. 6. EBSD maps and misorientation angle distribution of the specimens loaded along the ED and TD of the Mg–2Gd–3Zn alloy, strained to ε = 0.2.

Fig. 7. Distribution of the Schmid factor for the basal slip in (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys.
strain hardening rate, due to elastoplastic transition. For the Zn-containing specimens tested along TD, the steep decrease is followed by a positive change in strain hardening rate, associated with stage II of strain hardening, which has been previously observed in Mg–Gd–Zn alloys [13]. At higher stress levels, a linear decrease in SHR is observed for all materials, which is attributed to stage III of strain hardening. By extrapolating the stage III to \( \sigma - \sigma_{0.2} = 0 \), as shown in Fig. 8a, the initial hardening rate of stage III \( \theta_{III}^{0} \) is obtained and summarized in Fig. 9. In addition to \( \theta_{III}^{0} \), the rate of decline in \( \theta \) is direction dependent in the Zn-containing alloys.

The hardening capacity, \( H_c \), of a material can be determined as:

\[
H_c = \frac{UTS - YS}{YS}
\]  

Fig. 9. Initial hardening rate of stage III \( \theta_{III}^{0} \) along ED, 45\(^\circ\) and TD.

Fig. 8. Strain-hardening curves of (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys, along ED, 45\(^\circ\) and TD.

Fig. 10. Strain hardening capacity of the studied materials along different tensile directions.

Hardening capacity of the studied materials are demonstrated in Fig. 10. As can be seen, addition of Zn leads to an initial decrease in \( H_c \), followed by an increase with raising the Zn content. The SHR and hardening are closely related to dislocation evolution. It has been proposed that the flow stress can be expressed as:

\[
\sigma = \sigma_0 + \sigma_{HP} + \sigma_d
\]
\[ \sigma_d = \alpha \sigma b \sqrt{\frac{\rho}{d}} \]  \hspace{1cm} (5)

where \( \sigma_0 \) is the frictional stress, \( \sigma_{HP} \) is the Hall–Petch stress, \( \sigma_d \) is the Taylor dislocation stress, \( M \) is the Taylor factor, \( \alpha \) is a constant, \( G \) is the shear modulus, \( b \) is the Burgers vector and \( \rho \) is the dislocation density. Considering that \( \rho = \rho_0 + \rho_{GB} \), Eqs. (4) and (5) can be combined and the dislocation density can be related to the net flow stress:

\[ \rho^{1/2} \propto \sigma_d \approx (\sigma - \sigma_{0.2}) \]  \hspace{1cm} (6)

On the other hand, the relation between macroscopic and microscopic hardening rates can be stated as:

\[ \frac{d\sigma}{dt} = \frac{1}{m^2} \frac{d\sigma}{d\epsilon} = \frac{1}{m} \frac{d\sigma}{d\epsilon} = \frac{\alpha Gb}{m} \frac{d\rho}{d\epsilon} \]  \hspace{1cm} (7)

where \( \tau \) and \( \gamma \) are resolved shear stress and strain on slip plane and \( m \) is the Schmid factor. According to Eqs. (6) and (7), the relationship between \((\sigma - \sigma_{0.2})\theta\) and \((\sigma - \sigma_{0.2})\) is equivalent to dislocation storage rate \((d\rho/d\epsilon)\) and dislocation density \((\rho^{1/2})\) [25], as demonstrated in Fig. 11.

The initial dislocation storage rate can be fitted with a straight line. Zhao et al. [15] have reported that the initial slope is mainly influenced by grain size. However, a difference in the slope of the linear part is also observed along different directions (for example Fig. 11d). It can be inferred that in addition to grain size, texture has a major effect on the dislocation storage rate.

4. Discussion

4.1. Effect of Zn content on microstructure and texture

As shown in Figs. 2 and 3, microstructure of the extruded alloys consists of equiaxed grains and a high fraction of HAGBs, because of complete dynamic recrystallization (DRX) during hot extrusion. A slight difference is observed in the grain sizes of the extruded materials. For the binary Mg–Gd alloys, it has been shown that solute drag imposed by segregation of Gd atoms can retard recrystallization, thereby achieving small grain sizes and a weak texture [26,27]. It can be inferred that by addition of small amounts of Zn, co-segregation of Zn and Gd leads to a more severe solute drag effect, resulting in grain refinement from 12.2 to 8.5 \( \mu \)m for the Mg–2Gd alloy and Mg–2Gd–1Zn alloy. However, increasing the Zn to 2 and 3 wt% results in the formation of second phase particles, reducing the solute drag effect, which is manifested by a slight increase in grain size. This has been observed in a previous study [28].

Texture evaluation (Figs. 3 and 4) suggests that addition of Zn not only affects the texture component, but also alters the peak intensity of the developed texture. The Mg–2Gd alloy exhibits a relatively weak basal texture. Such a texture has been previously reported for the hot rolled Mg–1 wt% Gd alloy [29]. It should be mentioned that the RE-texture observed in some RE-containing alloys [30] is not seen here. The 30° peak in the misorientation angle distribution of Mg–Gd, which is probably related to a near-coincidence site lattice (CSL) boundary \( \Sigma_{15a} \), indicates continuous dynamic recrystallization (CDRX) during extrusion. This can stem from the increased stacking fault energy by addition of Gd, and consequently a higher dynamic recovery rate [31]. It has been reported that evolution of RE-texture can be suppressed by CDRX [32]. It has also been shown that RE-texture can be obtained under certain extrusion speeds [33]. Therefore, both CDRX and inadequate extrusion speed can be responsible for the formation of a basal texture instead of the RE-texture. Regarding the texture intensity, solute drag effect imposed by Gd atoms is expected to contribute to texture weakening (peak intensity of 5.4 mrd) by balancing the differences in grain boundary energies, thereby hindering nucleation and growth of the preferred orientations [34,35].

By the addition of 1 wt% Zn, a lower peak intensity is obtained,

Fig. 11. The relationship between \((\sigma - \sigma_{0.2})\theta\) and \((\sigma - \sigma_{0.2})\) for (a) Mg–2Gd, (b) Mg–2Gd–1Zn, (c) Mg–2Gd–2Zn, and (d) Mg–2Gd–3Zn alloys, along ED, 45° and TD.
which can be attributed to intensified solute drag mechanism due to co-segregation of large Gd (atomic radius of 180 p.m.) and small Zn (atomic radius of 133 p.m.) atoms at grain boundaries and dislocations. It has been argued that in the Mg–RE–Zn alloys, the RE–non-RE solute pairing/clustering can amplify the solute drag and change texture evolution [28, 36]. However, further increase of Zn results in the formation of second phase particles, reducing the amount of Gd in the a-Mg matrix, and decreasing co-segregation and the solute drag effects. Hence, texture intensity is increased by raising the Zn content. Such composition dependent co-segregation has been previously reported for the Mg–Ce–Zn alloys [37].

In addition to texture intensity, tilting of basal poles from ND toward TD, which is more obvious in alloys with higher Zn content, is also observed. The split of the texture toward TD has been attributed to deformation by non-basal slip. It has been proposed that the TD-split texture can be related to the activation of prismatic (a) slips due to the presence of Gd [13]. However, TD-split texture is not observed in the Mg–2Gd alloy in the present study. The RD–TD double split texture observed in Mg–Zn–Ce alloy has been attributed to the growth advantage of non-basal orientations as a result of particle pinning or solute drag, changing the orientation relationships for high boundary mobility [38]. Furthermore, misorientation angle distribution (Fig. 3) indicates that for the Zn-containing alloys, discontinuous dynamic recrystallization (DDRX) and twinning induced dynamic recrystallization take place. This can be manifested by the shift of distribution to higher angles and a peak intensity at ~86°, respectively [31]. For the binary Mg–Zn alloys, a recent study has shown that a strong basal texture is evolved, despite activation of (c+a) slip. This has been ascribed to rapid growth of tension twins and preferential growth of basal orientation, due to the insignificant segregation of Zn at grain boundaries and low pinning effect [39]. The above observations suggest that addition of Zn stimulates the development of the TD-split texture by reducing the c/a ratio and activation of non-basal slip. In addition, co-segregation of Zn and Gd atoms as well as stronger pinning effect of the Mg–Gd and Mg–Gd–Zn precipitates, can restrict the preferential growth of basal orientations, leading to development of a non-basal texture.

4.2. Effect of texture on mechanical behavior

Considering the tensile specimen tested along the ED, YS and UTS are increased by the addition of Zn. This can be attributed to the Hall-Petch effect, solid solution strengthening and second phase hardening. According to the slight differences in grain sizes, the Hall-Petch strengthening effect seems to be negligible. Due to the small volume fraction as well as large (mostly larger than 1.5 μm) size of the second phase particles, second phase strengthening can be neglected. Therefore, the solid solution strengthening has the major contribution in the strengthening of the tested alloys. This is in accordance with the view of Xu et al. [40] who have proposed solid solution as the main mechanism contributing to the yield stress in Mg-Gd alloys. Texture can affect mechanical properties through: (i) change in SF of the basal planes as the main contributor to deformation, and (ii) possibility of non-basal slip activation, caused by a change in the grain orientations. Room temperature deformation of Mg alloys usually takes place through slip on the basal plane in (1120) directions. However, non-basal (a) and (c+a) slip [4] as well as (10T2) twinning [25] have been observed to be active at room temperature. It should be mentioned, however, that since the SF along ED is almost the same in all alloys (Fig. 7), texture is not likely to have a significant role in this case. Based on Fig. 5 and Table 2, increasing the Zn content results in a substantial difference in yield stress values along the ED, 45° and TD. This difference can be attributed to the corresponding mean SF (Fig. 7). In the case of Mg–2Gd alloy, there is a similar distribution of SF values along different directions, as the c-axes are mostly parallel to ND (Fig. 4a). Since basal slip is the dominant deformation mechanism at the beginning of deformation, similar YS values are obtained, as suggested by the Schmid law, $\tau_{\text{thr}} = \tau_{\text{cRSS}}/m$. On the other hand, the Mg–2Gd–3Zn alloy demonstrates a clear difference in distribution of SF values along ED and TD, i.e. the mean SF values of 0.17 and 0.24, respectively. This is also observed for the Mg–2Gd–1Zn and Mg–2Gd–2Zn alloys to a lesser extent. Under a tensile stress applied along ED, most basal planes are in hard orientation (SF ≤ 0.35), leading to a high yield stress. However, basal planes have a more favorable orientation when the specimen is tested along TD, since the TD-split texture provides a higher fraction of basal planes with soft orientation (SF ≥ 0.35). This texture softening is the main reason for achieving a lower YS along TD. In contrast to YS, the effect of texture on the UTS along different directions is less significant. This is because yielding is mostly associated with basal slip, and thus, substantial difference in the SF distribution results in considerable anisotropy in YS. On the other hand, UTS is affected by strain-hardening behavior (particularly in stage III), which is related to multiple slip, twinning and dynamic recovery. After yielding by basal slip and some extent of strain-hardening, non-basal slip and twinning can be activated. Strain-hardening compensates some texture softening effects observed for the yield stress, leading to a less significant difference in the UTS along ED, 45° and TD.

Comparing the stress–strain curves of the Zn-containing specimens tested along ED and TD, a small plateau (a concave-up region after the yield) is observed along TD. These concave-up stress–strain curves have been observed for Mg alloys, where it has been ascribed to the twinning dominated deformation in this region [33,41]. On the other hand, previous studies suggest that extension twinning readily takes place, under tensile loading parallel to c-axis [21]. Fig. 4b shows that too many grains are oriented with their c-axes parallel to the TD, indicating high SF for extension twins. Therefore, twinning takes place more easily along TD, as shown in Fig. 6.

It has been suggested that the strain hardening capability of a material can be considered as a measure of its resistance to develop tensile mechanical instabilities, thereby controlling ductility [25]. Based on Considère criterion, instability occurs when $\sigma_{\text{y}}/\sigma_{\text{flow}} = \kappa$, i.e. when SHR is exhausted and becomes lower than the flow stress. It is obvious that a higher strain-hardening capability in the TD specimens (Fig. 10) of the Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys, delays the onset of the deformation instability, i.e. necking. As a result, a significantly higher uniform plastic strain is observed along TD for Mg–2Gd–2Zn and Mg–2Gd–3Zn alloys. The strain-hardening capability of materials depends on the YS, which in turn is related to the texture and grain size [18]. Slightly higher hardening capability of Mg–2Gd compared to Mg–2Gd–1Zn can be attributed to its larger grain size, as grain refinement reduces the hardening capacity by diminishing the difference between flow resistance of the grain boundary and grain interior. Mg–2Gd–3Zn along TD demonstrates the highest hardening capacity due to the low YS, activation of non-basal slip by Zn addition, and the TD-split texture. In addition to the strain hardening capability, twinning in the specimens tested along TD can also contribute to strain accommodation, leading to ductility improvement. Luo et al. [4] have also suggested extension twins responsible for higher ductility of Mg–Gd–Zn alloys.

In summary, addition of Zn yields a weak non-basal texture. Due to the type of texture, a significantly lower yield stress is obtained along TD as a result of texture softening. In addition to texture softening, activation of non-basal slip systems as well as twinning along TD result in lower yield stress along TD. This effect is more pronounced in the high Zn-containing alloys.

4.3. Effect of texture on strain hardening

Similar to the strain-hardening behavior of fcc metals, three stages can be recognized for the strain-hardening of Mg alloys, including an initial elastoplastic transition with a very rapid decrease in SHR, an increase in SHR to a peak value and a short plateau (stage II), and a
linear decrease associated with dynamic recovery (stage III). However, the extent of the strain-hardening stages in Mg alloys is greatly influenced by the microstructure, texture, and deformation mechanisms. As observed for Mg–2Gd (Fig. 8a), stage II was suppressed and the linear stage III, which is associated with multiple slip and dynamic recovery, can be initiated from the earlier stages of deformation. Both texture and increased SFE, due to the presence of Gd, can be responsible for the assisted dynamic recovery. Regarding the effect of orientation, due to the basal texture, c-axes of the grains in the ED specimen are nearly perpendicular to the tension direction, which is a hard orientation for basal slip. Therefore, multiple slip (especially prismatic (a) slip with an average SF of 0.26, is activated at lower strains, suppressing stage II hardening followed by a high $\theta^a_{\text{III}}$ and a rapidly declining SHR. According to previous reports, it can be inferred that activation of prismatic (a) slip and cross-slip of (a) dislocation onto the prismatic planes can assist dynamic recovery and lead to a rapid transition to stage III [25,42]. Hu et al. [43] have also reported that in Mg alloys with basal texture, activation of prismatic (a) is highly favored. Early activation of multiple slip in Mg alloys leads to the formation of dislocations forests, causing rapid increase in dislocation storage rate, as shown in Fig. 11. However, it is followed by a fast decrease in dislocation storage rate. A similar trend is also obtained for the specimen tested along ED in the Zn-containing alloys. Since the c-axes of most grains are almost perpendicular to ED, the stage II is suppressed, resulting in the appearance of stage III from earlier stages of deformation by enforcing the prismatic slip, while materials with ideal high SF values (i.e. TD) exhibit stage II [25,44].

Evaluation of SHR indicates that addition of Zn is accompanied by a change in the strain hardening behavior along various directions, particularly the presence of a marked stage II hardening in the 45° and TD specimens. The distinct yield point in the stress–strain curves (Fig. 5) corresponding to the stage II hardening is associated with activation of tension twins (Fig. 6), which are more easily activated along TD [33,45]. Tension twins influence SHR through; (i) hampering of dislocation motion by twin boundaries acting as obstacles (a Hall-Petch like effect), and (ii) texture hardening by reorienting the grains into hard orientations, as usually observed in Mg alloys [46,47]. Therefore, stronger interactions between twins and gliding dislocations brings about a more apparent stage II and a higher dislocation storage rate (Fig. 11) [48]. Shi et al. [12] have suggested a critical tilting degree of 10°–25° for basal poles toward the tension direction, above which stage II would be observed.

In addition to the stage II hardening, strain-hardening behavior of different specimens is also different in term of $\theta^a_{\text{III}}$ and hardening capacity. As can be seen in Fig. 9, Zn-containing alloys demonstrate a higher $\theta^a_{\text{III}}$ along ED rather than TD. Higher SF of basal slip, as the dominant deformation mechanism especially at the onset of deformation, along TD leads to a lower $\theta^a_{\text{III}}$ along this direction. Despite higher $\theta^a_{\text{III}}$ of the ED specimens, the strain hardening rate decreases more rapidly that leads to a lower hardening capacity and a lower ductility in this direction. However, the 45° and TD specimens can maintain their high SHR to a larger extent, resulting in higher hardening capacity and better ductility. Dynamic recovery originated from the cross-slip of (a) dislocations from basal to prismatic planes is accompanied by a softening effect, decreasing the decline rate in SHR [3]. In the TD specimen, the TD-split texture, which is favorable for basal slip (SF > 0.35), postpones the activation of non-basal slip. At the early stages of deformation, only basal slip is the dominant deformation mechanism in the TD specimens. This would result in a lower SHR compared to the ED specimens in which multiple slip occurs more readily, as manifested by earlier stage III. By further straining, the relative orientation of the hard slip systems compared to the soft ones changes and a transition in the deformation mechanism from basal to mixed basal + non-basal slip occurs. Furthermore, rotation of the basal poles from ND to TD also increases the tendency for twinning, especially at higher elongations (Fig. 6) [21]. As a result, a slower decline in SHR is observed along TD compared to ED. The higher strain hardening rate at high stress levels can also delay initiation of the deformation instability, providing a higher ductility along TD.

In conclusion, although the alloys containing 2 and 3 wt% Zn demonstrate higher SHR at high stress levels, a high anisotropic strain hardening behavior is observed for these alloys. It seems that the Mg–2Gd–1Zn alloys shows the best combination of strain hardening capacity and low anisotropy.

5. Conclusions

Texture, mechanical properties and strain-hardening behavior of the extruded Mg–2Gd–xZn (x = 0, 1, 2 and 3 wt%) sheets were investigated. The extruded alloys consisted of equiaxed grains and a high fraction of HAGBs, because of complete dynamic recrystallization (DRX) during hot extrusion. Texture evaluation revealed that the processing condition was inadequate for the formation of RE-texture, and thus, a weak basal texture was formed in the Mg–2Gd alloy. This type of texture resulted in similar mechanical properties in various directions. Addition of Zn resulted in a ~40° rotation of basal poles from ND toward TD, giving a typical TD-split texture. This type of texture led to a planar anisotropy in mechanical properties and strain-hardening behavior, with lowerYS and UTS and higher $\theta^a$ and strain-hardening capacity along TD compared to ED. Variation in the SF of basal planes and deformation mechanism was recognized as the reason for anisotropy. Generally, TD specimens exhibited lower $\theta^a_{\text{III}}$ and a slower decline in SHR, due to easier activation of basal slip in the beginning of deformation and activation of tension twinning and non-basal slip at later stages of deformation. The Mg–2Gd–1Zn alloy showed excellent mechanical properties and low YS and UTS anisotropy as a result of a weak non-basal texture.

Data availability statement

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

Funding

This research did not receive any specific grant from funding agencies in the public, commercial, or not-for-profit sectors.

Author contribution

M. M. Hoseini-Athar: Methodology, Data processing, Original draft preparation, Investigation. R. Mahmudi:Conceptualization, Supervision, Validation, Writing- Reviewing and Editing. R. Prasath Babu: Methodology, Data processing, Editing. P. Hedström: Methodology, Data processing, Validation, Editing.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

References

[3] J. Suh, J. Victoria-Hernández, D. Letzig, R. Golle, W. Volk, Enhanced mechanical behavior and reduced mechanical anisotropy of AZ31 Mg alloy sheet processed by