Final Draft
of the original manuscript:

Liu, L.; Pan, F.; Chen, X.; Huang, Y.; Song, B.; Yang, H.; Hort, N.: 
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First published online by Elsevier: June 19, 2018

DOI: /10.1016/j.vacuum.2018.06.048
https://dx.doi.org/10.1016/j.vacuum.2018.06.048
The effect of Y addition on recrystallization and mechanical properties of Mg–6Zn–xY–0.5Ce–0.4Zr alloys

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Abstract

Microstructure and mechanical properties of extruded Mg–6Zn–xY–0.5Ce–0.4Zr sheets were systematically investigated. Mg–Zn–Ce, I-phase (Mg$_3$Zn$_6$Y) and W-phase (Mg$_3$Zn$_3$Y$_2$) with different morphologies were observed in as-cast and extruded conditions. Numerous nano-precipitates were found along grain boundaries and inside grains in both dynamically recrystallized (DRXed) and unDRXed regions after hot extrusion. The area fraction of DRXed regions increases with Y additions mainly due to particle stimulated nucleation (PSN). An extruded fiber texture was observed after hot extrusion. The texture is strengthened firstly and then weakened. The extruded alloys with 0.5–1 wt.% Y and 0.5 wt.% Ce exhibit much better mechanical properties compared with other Mg alloys such as Mg–5Zn–1Y–1Zr (wt.%) and Mg–6Zn–1Ce–0.5Zr (wt.%) alloys under the same extrusion ratio. The extruded alloy with 0.2 wt.% Y and 0.5 wt.% Ce has the yield strength of 322 MPa, ultimate tensile strength of 378 MPa and elongation of 8.6%, which is the highest value among the studied alloys. Grain boundary strengthening and precipitation strengthening are the two most important factors for improving yield strength of Mg–6Zn–xY–0.5Ce–0.4Zr alloys. The addition of Y is an effective way in significantly improving the mechanical properties of Mg–6Zn–xY–0.5Ce–0.4Zr alloys.
1 Introduction

Magnesium (Mg) alloys with low density receive strong research interest for applications to various structural components of automobiles and aircrafts due to the urge demands for weight reduction of transportation vehicles for better fuel efficiency [1–9]. However, the extensive utilizations of Mg alloys are limited by their unfavorable mechanical properties, such as poor absolute strength and formability at room temperature compared with some Al alloys and steel. Additions of rare-earth (RE) metals to Mg alloys are known as an effective approach to improve mechanical properties [10–14].

Among the different Mg–RE alloys, Mg–Zn–RE alloys have recently gained increasing attentions from numerous researchers because of their excellent strength. The formation of a large number of fine stable precipitates in these alloys can result in their advantageous mechanical properties. For example, an extruded, rolled, and aged Mg-8.2Gd-3.8Y-1.1Zn-0.4Zr (wt.%) alloy shows a yield strength (σ_y) of 416 MPa, an ultimate tensile strength (σ_{UTS}) of 505 MPa, and elongation to failure of 12.8% [15]. An extruded Mg-10Gd-5.7Y-1.6Zn-0.6Zr (wt.%) alloy shows a high σ_y of 473 MPa and σ_{UTS} of 542 MPa [16]. However, additions of large quantities of RE metals increase the cost for these alloys. This high cost is undesirable for industrial applications. Therefore, RE-free or low-RE-content Mg alloys with high performance are highly desired.

Y and Ce are alloying elements commonly used in Mg alloys, and they are also two representative RE elements [17]. Ce possesses low solid solubility in α-Mg and easily forms the second phase [18]. Its addition to Mg can improve its strength and elongation through grain refinement and texture weakening [19–22]. Y can effectively refine grains and reduce stacking fault energy of Mg [23–26]. We have studied previously the microstructure and mechanical properties of Mg–Zn–Y–Zr [27] and Mg–Zn–Ce–Zr [28] alloys. If Y and Ce are simultaneously added to Mg-Zn-Zr alloy, their synergic strengthening effects are expected to be much better. The new Mg–Zn–Y–Ce–Zr alloys were developed [29]. The strength is much higher than Mg–Zn–Zr alloys separate with Y or Ce in the same extrusion ratio (11.5:1). However, the size of particles in above 4 μm is too large that deteriorate mechanical property after extrusion. Therefore, a low-RE-content alloy series with total RE content less than 2 wt.%, Mg-6Zn-xY-0.5Ce-0.4Zr (x = 0.2, 0.5, 1 and 1.5 wt.%), was designed in this work. The new Mg-6Zn-xY-0.5Ce-0.4Zr alloys are improved by changing Zn/Y ratio that the alloys contain more fine strengthening particles (I-phase) than before.

To date, several investigations have been carried out for Mg–Zn (–Zr) alloys added with Y and Ce. Guo et al. [30] studied the effects of rapid solidification and reciprocating extrusion on the microstructure and mechanical behavior of Mg–6Zn–1Y–0.6Ce–0.6Zr alloy. A large number of particles was observed in
Table 1: Chemical compositions of Mg–6Zn–xY–0.5Ce–0.4Zr alloys.

<table>
<thead>
<tr>
<th>Samples</th>
<th>Nominal Composition</th>
<th>Composition (wt.%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Mg-6Zn-xY-0.5Ce-0.4Zr</td>
<td>Mg</td>
</tr>
<tr>
<td>Alloy 1</td>
<td>Mg-6Zn-0.2Y-0.4Zr</td>
<td>bal.</td>
</tr>
<tr>
<td>Alloy 2</td>
<td>Mg-6Zn-0.5Y-0.4Zr</td>
<td>bal.</td>
</tr>
<tr>
<td>Alloy 3</td>
<td>Mg-6Zn-1Y-0.4Zr</td>
<td>bal.</td>
</tr>
<tr>
<td>Alloy 4</td>
<td>Mg-6Zn-1.5Y-0.4Zr</td>
<td>bal.</td>
</tr>
</tbody>
</table>

rapidly solidified Mg–6Zn–1Y–1Ce alloy [31]. They predominantly composed of T phase, W phase and Mg$_4$Zn$_7$ phase. However, the effects of Y and Ce contents on the microstructure and mechanical performance are still unclear. Therefore, in the present study, dynamic recrystallization, texture and mechanical properties of Mg–6Zn–xY–0.5Ce–0.4Zr alloys with different Y contents were investigated.

2 Experimental

Commercially pure Mg (99.95 wt.%), Zn (99.95 wt.%), and Mg-30 wt.% Y, Mg-30 wt.% Ce and Mg-27.85 wt.% Zr master alloys were used to prepare the experimental alloys (Alloy I to IV). The nominal composition of Alloy I to Alloy IV is Mg-6Zn-0.5Ce-0.2Y-0.4Zr, Mg-6Zn-0.5Ce-0.5Y-0.4Zr, Mg-6Zn-0.5Ce-1Y-0.4Zr, and Mg-6Zn-0.5Ce-1.5Y-0.4Zr (wt.%), respectively. The actual chemical compositions listed in Table 1 were determined by Inductively Coupled Plasma (ICP-OES). The alloys were melted in an electric resistance furnace using a mild steel crucible under mixed protection atmosphere of CO$_2$ and SF$_6$ with a ratio of 99:1. Upon reaching the temperature 750 ºC, the melt was stirred for 8 min and subsequently held for 30 min. Then, it was poured into the steel mould pre-heated to 200 ºC. The ingots were cylindrical with a diameter of 130 mm and length of 250 mm. The ingots were homogenized at 400 ºC for 12 h to alleviate the solute segregation. Prior to hot extrusion, the bars machined from the homogenized ingots with 120 mm diameter and 130 mm length were pre-heated at 370 ºC for 2 h in a resistance furnace. The extrusion was then conducted at 370 ºC with a ratio of 11.5:1 and a ram speed of 4 mm/s. The cross-section area of the extruded plate was $8 \times 125 \text{ mm}^2$.

The constituent phases of as-cast alloys were identified with X-ray diffraction (XRD, Rigaku D/MAX-2500PC) using Cu-$k\alpha$ radiation with a scanning angle from 10º to 90º and a scanning speed of 2º/min. The macro-texture was also determined by XRD. Microstructural observations were performed using optical microscopy (OM), scanning electron microscopy (SEM, TESCAN VEGA II LUM) with energy dispersive spectrum (EDS) operated at 20 kV and transmission elec-
tron microscopy (TEM, FEI TECNAI G2 F20) with an accelerating voltage of 200 kV. For OM and SEM observations, the polished specimens were etched in a mixture of 1 g picric acid, 2.6 ml acetic and 16 ml ethanol.

Thin foil specimen for TEM observation was prepared by punching 3 mm diameter discs, mechanical polishing, dimple grinding, as well as ion milling using Gatan Precision Ion Polishing System. The dynamic microstructure and their corresponding micro-texture analyses of the alloys were examined by electron back scattering diffraction (EBSD, JEOL JSM-7800F equipped with an HKL channel 5 systems). The specimen for EBSD orientation mapping was prepared by mechanical grinding, followed by electro-polishing with a solution of AC2 electrolyte.

Tensile testing was carried out in a CMT-5105 material testing machine at a strain rate of $1 \times 10^{-3}/s$ at room temperature. The rectangular tensile specimens with a gauge length of 25 mm and a cross-sectional area of $2 \times 5 \text{ mm}^2$ were prepared by an electrical sparking wire cutting machine. Each tensile test was repeated three times. The tensile deformation direction was parallel to extrusion direction (ED).

3 Results and Discussion

3.1 Microstructural characterizations

The XRD patterns of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys are shown in Fig. 1. Four main phases were identified, namely, α-Mg, Mg–Zn–Ce, I-phase (Mg$_3$Zn$_6$Y) and W-phase (Mg$_3$Zn$_3$Y$_2$). The diffraction of W-phase is gradually intensified while the amount of I-phase reduces with increasing Y content.

The OM micrographs of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys are shown in Fig. 2. The average grain sizes of Alloy I-IV were about 74, 83, 94 and 88 $\mu$m, respectively measured by line intercept method. The grain size increases with Y addition and then decreases when Y content is further increased to 1.5 wt.%. The grains of the alloys with low Y content are finer than that with high Y content. This might be due to low I-phase/matrix interfacial energy that brought about strong bonding to I-phase/matrix interfaces [32].

Fig. 3 presents SEM images of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys. Most of second phases were formed as network at grain boundaries. The detailed morphologies of four main phases are given in Fig. 4. Table 2 is EDS analysis of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys related to Fig. 4. The block-shaped Mg–Zn–Ce phase (Fig. 4a and b) and the lamellar W-phase (Fig. 4c and d) were found at triple junctions of grain boundaries. The rod and granular I-phase (Fig. 4e and f) located inside the grains and at triple junctions of grain boundaries. Those Zr-rich particles (Fig. 4g and h) were found mostly inside the grains.
Figure 1: XRD patterns of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys.

Figure 2: OM micrographs of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys: 
(a) Alloy I, (b) Alloy II, (c) Alloy III, and (d) Alloy IV.
Figure 3: SEM micrographs of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys: (a) Alloy I, (b) Alloy II, (c) Alloy III, and (d) Alloy IV.

Table 2: EDS analysis of as-cast Mg–6Zn–xY–0.5Ce–0.4Zr alloys related to Fig. 4.

<table>
<thead>
<tr>
<th>Point</th>
<th>Mg</th>
<th>Zn</th>
<th>Y</th>
<th>Ce</th>
<th>Zr</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>60.1</td>
<td>32.6</td>
<td>-</td>
<td>7.3</td>
<td>-</td>
</tr>
<tr>
<td>2</td>
<td>63.9</td>
<td>30.5</td>
<td>-</td>
<td>5.6</td>
<td>-</td>
</tr>
<tr>
<td>3</td>
<td>35.6</td>
<td>46.3</td>
<td>21.1</td>
<td>-</td>
<td>-</td>
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<tr>
<td>4</td>
<td>78.3</td>
<td>13.1</td>
<td>8.6</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>5</td>
<td>77.7</td>
<td>18.9</td>
<td>3.4</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>6</td>
<td>88.7</td>
<td>9.8</td>
<td>1.5</td>
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<td>79.5</td>
<td>17.1</td>
<td>3.4</td>
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<td>8</td>
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<td>33.6</td>
<td>5.3</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>9</td>
<td>10.0</td>
<td>36.4</td>
<td>-</td>
<td>-</td>
<td>53.6</td>
</tr>
<tr>
<td>10</td>
<td>43.1</td>
<td>21.4</td>
<td>-</td>
<td>-</td>
<td>35.5</td>
</tr>
</tbody>
</table>
Figure 4: SEM micrographs showing the morphology of (a and b) Mg-Zn-Ce, (c and d) Mg$_3$Zn$_5$Y$_2$, (e and f) Mg$_3$Zn$_6$Y, and (g and h) Zn, Zr-rich phase in as-cast Alloy IV.
Fig. 5 shows OM microstructures of the extruded alloys. A bimodal structure with fine equiaxed grains and deformed grains was observed in all alloys. The area fractions of deformed regions reduce with Y and Ce additions. They were measured by the software Image-Pro Plus (Fig. 5 (e)). They are 43%, 40%, 37% and 24% for Alloy I-IV, respectively. The average size of those fine grains is about 2.4, 2.8, 2.9 and 2.7 µm measured by intercept method for Alloy I-IV, respectively.

Fig. 6 illustrates typical SEM images showing the as-extruded microstructure for the alloys. Many particles with different sizes were inhomogeneously distributed in α-Mg matrix. Evidently, the amount of fragmented particles increases with Y content. The initial particles in the as-cast alloy were broken into fine particles during hot extrusion. These broken particles were zonally distributed along ED (Fig. 6 (b)).

Fig. 7 shows the morphology of the particles and their corresponding EDS results in Alloy II. Alloy II was selected due to other three alloys are in similar morphology of particles based on XRD and SEM analysis. The Zr-rich phases are in rod-shape after hot extrusion (Fig. 7 (a)). The block-shaped Mg–Zn–Ce and I-phase located in the particle bands. The size of Mg–Zn–Ce particles is about 2–4 µm. I-phase (0.5–2.5 µm) is relatively finer than Mg–Zn–Ce phase.

Fig. 8 (a) shows the bright field TEM images of I-phase in the extruded Alloy II. Alloy II was selected due to other three alloys are in similar type of particles. There are three different zone axes for the icosahedral quasicrystalline phase (I-phase), i.e. 5-fold, 3-fold, 2-fold. The corresponding selected area electron diffraction (SAED) of the I-phase along 5-fold zone axis is presented in Fig. 8.
Figure 6: SEM micrographs of extruded Mg–6Zn–xY–0.5Ce–0.4Zr alloys:
(a) Alloy I, (b) Alloy II, (c) Alloy III and (d) Alloy IV.
Figure 7: (a and c) SEM images of Zr-rich rod shape phase, Mg-Zn-Ce and Mg$_3$Zn$_3$Y$_2$ particles, respectively. (b, d and e) its corresponding EDS results for the extruded Alloy II.
Figure 8: (a) Bright field TEM image of Mg₃Zn₆Y phase in the extruded Alloy II and the inset is its corresponding SAED pattern along 5-fold zone axis; (b) Mg-Zn-Ce phase and its corresponding EDX results.

(a). Fig. 8 (b) shows the bright field TEM images of Mg-Zn-Ce phase and its corresponding EDX results. Its crystal structure was described in our previous paper [33].

Fig. 9 (a) illustrates the bright field TEM observations of the precipitates in the extruded Alloy II. Some precipitates were found along grain boundaries. They are believed to exert a strong pining effect against the continued migration of grain boundaries. In addition, numerous nano-precipitates were distributed in DRXed grains to effectively strengthen the Mg matrix (Fig. 9 (b)). A large amount of nano-precipitates was also observed in the unDRXed regions (Fig. 10). Such fine precipitates were not observed in the as-cast samples, indicating that they were formed during hot extrusion at 370 °C.

### 3.2 Dynamic recrystallization and texture of extruded alloys

The particles with diameter greater than ~1 µm were reported to be capable of acting as nucleation sites for dynamic recrystallization during hot deformation (i.e. particle stimulated nucleation (PSN) recrystallization) [34,35]. The deformation processing created a strain gradient area being full of both high dislocation density and large orientation around the particles. The sub-boundaries migrations were formed around the particle rapidly in deformation domains during plastic deformation, resulting in the formation of the new high-angle grain boundaries [35]. The PSN phenomenon was demonstrated in Fig. 11 (a and b) which shows...
Figure 9: (a and b) Bright field TEM images showing fine precipitates along grain boundary and inside grains in the extruded Alloy II, respectively.

Figure 10: Bright field TEM image showing fine precipitates in unDRXed area in the extruded Alloy II.
Figure 11: (a) SEM image and Mg-Zn-Ce and Mg-Zn-Y phase EDX analysis; (b) IPF map corresponds to (a) for the extruded Alloy I; (c) (0001) pole figures of the whole, unDRXed and DRXed region.
the SEM image and its corresponding inverse pole figure (IPF) map for the extruded Alloy I. Obviously, dynamic recrystallization occurred in the regions containing the particles while the coarse non-recrystallized grains corresponded to the particle-free zones. These particles were Mg-Zn-Ce and Mg-Zn-Y particles based on the phase identification. They are the effective nucleation sites for recrystallization. It is worth noting that local deformation (indicated by gradual color change) could be seen inside the large deformed grain, indicating the possible existence of low angle boundaries (LAB). Continuous dynamic recrystallization (CDRX) might also happen in the large deformed grain. Fig. 11 (c) shows (0001) pole figures taken from the DRXed region, the unDRXed region and the whole region of the extruded Alloy I, respectively. The extruded Alloy I exhibits a strong fiber texture with basal planes parallel to the extrusion direction. It is clear that the unDRXed region exhibits stronger basal texture than the DRXed region.

Fig. 12 (a–d) show the IPF maps of extruded Alloy I-IV and (e–h) are corresponding completely DRXed regions. The extruded alloys exhibit a bimodal grain structure consisting of coarse unDRXed regions and fine DRXed grains, which is consistent with OM observations. The unDRXed deformed regions decreases with Y additions. The amount of fragmented particles increases with Y content as shown in Fig. 6. In addition, the unDRXed grain orientation of extruded Alloy I-IV is almost parallel to ED (Fig. 12).
Figure 13: (0002) and (1010) macro-texture of the as-extruded Mg–6Zn–xY–0.5Ce–0.4Zr alloys: (a) Alloy I, (b) Alloy II, (c) Alloy III and (d) Alloy IV.
The (0001) and (1010) macro-texture of the as-extruded alloys are presented in Fig. 13. The orientation distribution of the basal planes for Alloy I extends to transversal direction (TD) (Fig. 13a). In the (1010) pole figure the intensity is high at the top and bottom poles. It is a typical extruded fiber texture. This extruded fiber texture also exists in the alloys with the addition of 0.5–1.5 wt.% Y combined with 0.5 wt.% Ce. The values of peak intensity in (1010) pole figure are 8.2, 10.1, 12.1 and 11.5 for Alloy I–IV, respectively. The maximum intensity in (0001) pole figure for Alloy I–IV has the similar evolution tendency. The boundary strengthening and precipitate strengthening made a great contribution to the improvement of yield strength for Mg–6Zn–xY–0.5Ce–0.4Zr alloys.

The area fraction of DRXed regions is higher than that of unDRXed regions as measured through OM. Although the texture of unDRXed regions is stronger than that of the DRXed regions, its effect on macro-texture is limited due to its small area fraction and quite big grains. Therefore, the texture intensity of DRXed regions might have great effect on the macro-texture intensity. Fig. 14 shows the (0001) pole figures for the DRXed regions in Fig. 12. The texture of DRXed regions is strengthened firstly and then weakened by Y additions, showing the same evolution tendency with the macro-texture for Alloy I–IV. The second phases play an important role in DRX. A significant competition in DRX exist between the I-phase and the W-phase. I-phase is a harder phase and supposedly can create more strain in the vicinity, resulting in nucleation of new grains on its surface. The high symmetry of I-phase will lead to orientation randomization of newly formed DRXed grains.

The texture of Alloy I containing more I-phase is weaker compared with other alloys might be due to this fact. The eutectic W-phase particle with a net-like morphology is largely distributed at grain boundaries [36]. The mobile dislocations would move into the grain boundaries and then be absorbed, resulting in reduction in the dislocation density. The texture intensity increases with increasing the amount of W-phase. However, the texture slightly weakens when the Y content reaches to 1.5 wt.%. This might be due to the decreasing distance between those particles which could effectively hinder dislocation movement.

### 3.3 Mechanical properties of as-extruded alloys

The engineering stress-strain curves are illustrated in Fig. 15. Alloys I–III have good mechanical properties. Alloy I with 0.2 wt.% Y and 0.5 wt.% Ce exhibit the optimal yield strength of 322 MPa, ultimate tensile strength of 378 MPa and good elongation of 8.6%. The ultimate tensile strength decreases slightly with Y additions. Apparently, a higher addition with 1.5 wt.% Y substantially deteriorates the room-temperature tensile property. Ultimate tensile strength, 0.2% proof yield strength, and elongation for as-extruded alloys are shown in Table 3. Table
Figure 14: (0001) pole figures of DRXed regions measured by EBSD corresponding to Fig. 12(e–h) (a) Alloy I, (b) Alloy II, (c) Alloy III and (d) Alloy IV.

Figure 15: Typical tensile engineering stress-strain curves of as-extruded Mg–6Zn–xY–0.5Ce–0.4Zr alloys.
Table 3: Ultimate tensile strength (UTS), 0.2% proof yield strength (TYS), and elongation (El.) for the as-extruded Mg–6Zn–xY–0.5Ce–0.4Zr alloys and comparison with other Mg alloys.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>UTS</th>
<th>TYS</th>
<th>EL.</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>MPa</td>
<td>MPa</td>
<td>%</td>
<td></td>
</tr>
<tr>
<td>Alloy I</td>
<td>378 ±2</td>
<td>322±5</td>
<td>8.6±0.5</td>
<td>this work</td>
</tr>
<tr>
<td>Alloy II</td>
<td>369±4</td>
<td>327±4</td>
<td>7.3±0.4</td>
<td>this work</td>
</tr>
<tr>
<td>Alloy III</td>
<td>362±4</td>
<td>315±5</td>
<td>8.7±0.3</td>
<td>this work</td>
</tr>
<tr>
<td>Alloy IV</td>
<td>345±5</td>
<td>298±6</td>
<td>4.2±1.1</td>
<td>this work</td>
</tr>
<tr>
<td>Mg-6Zn-0.5Zr</td>
<td>302</td>
<td>231</td>
<td>11</td>
<td>[28]</td>
</tr>
<tr>
<td>Mg-5Zn-1Y-1Zr</td>
<td>321</td>
<td>265</td>
<td>8</td>
<td>[27]</td>
</tr>
<tr>
<td>Mg-6Zn-1Ce-0.5Zr</td>
<td>316</td>
<td>249</td>
<td>15</td>
<td>[28]</td>
</tr>
</tbody>
</table>

3 compares the mechanical properties of Mg-6Zn-xY-0.5Ce-0.4Zr alloys with a commercial Mg–6Zn–0.5Zr [28], Mg–5Zn–1Y–1Zr [27] and Mg-6Zn-1Ce-0.5Zr [28] alloys in the same extrusion ratio (11.5:1). The yield strength of the extruded alloys with 0.5–1 wt. % Y and 0.5 wt. % Ce in this study was even higher than that of Mg–6Zn–0.5Zr, Mg–5Zn–1Y–1Zr and Mg–6Zn–1Ce–0.5Zr alloys.

Alloy II exhibited the optimal yield strength of 327 MPa with an increment of 96 MPa compared with Mg–6Zn–0.5Zr. The yield strength of as-extruded alloys is generally influenced by contributions from grain boundary strengthening and precipitate strengthening in the DRXed regions, dislocation strengthening and precipitate strengthening in the unDRXed regions, and solid solution strengthening in both regions. It is approximately expressed using linear relationship [36–38]:

$$\sigma_y = \Delta \sigma_{gb} + M(\tau_0 + \Delta \tau_s + \Delta \tau_p + \Delta \tau_d)$$  \hspace{1cm} (1)

where $\sigma_y$ is yield strength, $\Delta \sigma_{gb}$ is the strength increment contributed by grain boundary strengthening, $M$ is the Taylor factor, $\tau_0$ is intrinsic critical resolved shear stress, $\Delta \tau_s$, $\Delta \tau_p$ and $\Delta \tau_d$ are precipitate strengthening, solid solution strengthening, and dislocation strengthening, respectively. The extruded Alloy II is selected as an example to perform the following analysis due to its highest yield strength. For the present alloy, $M(\tau_0 + \Delta \tau_s)$ is regarded as $\sigma_0$ which is taken as the as-cast yield strength (115 MPa), because the solid solution strengthening is limited and not significantly changed by extrusion. The $M\Delta \tau_d$ is neglected in the DRXed regions due to the low dislocation density. Besides, $\Delta \sigma_{gb}$ could also be ignored owing to the coarse grains in the unDRXed regions.

Thus, Eq. (1) is simplified into two parts as below for the DRXed and un-DRXed regions, respectively:
\[ \sigma_{y/DRX} = \Delta \sigma_{gb} + \sigma_0 + M_{DRX} \Delta \tau_\rho \] (2)

\[ \sigma_{y/unDRX} = \sigma_0 + M_{unDRX}(\Delta \tau_p + \Delta \tau_d) \] (3)

The \( \Delta \sigma_{gb} \) can be calculated by the Hall-Petch relationship:

\[ \Delta \sigma_{gb} = kd^{-\frac{1}{2}} \] (4)

Where \( k \) is the Hall-Petch slope, \( k = 220 \text{ MPa} \mu\text{m}^{1/2} \) [39], and \( d \) is the grain size. The average grain size of DRXed region is 2.81 \( \mu\text{m} \) for the as-extruded Alloy II. The \( \Delta \sigma_{gb} \) in the DRXed region was calculated as 131 MPa. The \( M \) depends on texture which ranges from 2.1 with very strong texture and pyramidal slip activated, to 4.5 with random texture.

\( M \) is 3.5 when a strong texture is formed in Mg and the basal slip, the pyramidal slip and the prism slip are all active [40]. In the present study, value of \( M_{DRX} \) is taken as 3.5 since DRXed regions exhibit strong basal texture, while value of \( M_{unDRX} \) is taken as 2.5 due to the much stronger basal texture. The \( \Delta \tau_p \) from the broken particles can be calculated using the following equation:

\[ \Delta \tau = \frac{0.19 G b}{\lambda} \ln \frac{0.08 r}{b} \] (5)

where \( \Delta \tau_p \) is the increase of CRSS, \( G \) is the shear modulus (17 GPa); \( b \) is the Burger vector (0.32 nm); \( \lambda \) is the distance of the particles; and \( r \) is the diameter of particles. The average diameter of precipitates in the DRXed region is about 61 nm and the interspacing of the precipitates is about 190 nm (Fig. 9b), respectively. Thus the \( M_{DRXed} \Delta \tau_\rho \) for DRXed regions was estimated to be 60 MPa. The average diameter of precipitates in the unDRXed region is about 9 nm and the interspacing of the precipitates is about 15 nm (Fig. 10). Thus the \( M_{unDRXed} \Delta \tau_\rho \) for unDRXed regions was estimated to be 140 MPa.

The \( \Delta \tau_d \) was calculated by the following equation:

\[ \Delta \tau_d = \alpha G b \sqrt{\rho} \] (6)

where \( \alpha \) is a constant with a value of 0.2, \( \rho \) is the dislocation density. The dislocation density for the as-extruded Mg-Zn-Zr Alloy II was evaluated to be about \( 2 \times 10^{14} \text{ m}^{-2} \). Thus \( M_{unDRX} \Delta \tau_d \) was estimated to be 39 MPa.

The yield strength of the DRXed region and unDRXed region predicted by Eqs. (2) and (3) are 306 MPa and 294 MPa, respectively. Since the fraction of unDRXed regions is 43\%, the yield stress of the extruded Alloy II was predicted as \( 306 \times 57\% + 294 \times 43\% = 301 MPa \), which is a little smaller than the measured value (327 MPa). This might be due to the fact that the dislocation strengthening
in the DRXed region and the grain boundary strengthening in the unDRXed region were not considered during quantitative analysis. From above analysis, grain boundary strengthening and precipitate strengthening made a great contribution to the improvement of yield strength for Mg–6Zn–xY–0.5Ce–0.4Zr alloys.

4 Conclusions

The microstructure and mechanical properties of Mg–6Zn–xY–0.5Ce–0.4Zr alloys were investigated. Following conclusions can be obtained:

1. Three phases Mg–Zn–Ce, I-phase ($\text{Mg}_3\text{Zn}_6\text{Y}$) and W-phase ($\text{Mg}_3\text{Zn}_3\text{Y}_2$) with different morphologies exist in the as-cast and as-extruded alloys. Nano-precipitates were formed inside grains and along grain boundaries in both DRXed and unDRXed regions by deformation-induced precipitation during hot extrusion.

2. Mg-Zn-Ce and Mg-Zn-Y particles with more than 1 $\mu$m can act as effective nucleation sites for dynamic recrystallization (PSN). The area fraction of DRXed regions increases with Y additions. After hot extrusion an extruded fiber texture was observed which intensity is strengthened firstly and then weakened with increasing Y content.

3. The extruded alloy with 0.2 wt.% Y and 0.5 wt.% Ce exhibit the highest yield strength of 322 MPa, ultimate tensile strength of 378 MPa and elongation of 8.6% among the studied alloys. The main mechanism of the enhancement in yield strength of these Mg–6Zn–xY–0.5Ce–0.4Zr alloys is grain boundary strengthening and precipitate strengthening.

Acknowledgements

The authors would like to thank the financial supports from the National Key Research and Development Program of China (2016YFB0301100), National Natural Science Foundation of China (51571043 and 51531002), and Chongqing Municipal Government (cstc2017jcyjBX0040).

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