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Evolution of microstructure, texture and residual stress of AZ31 Mg alloy in hot extrusion process

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Abstract
In this work, we have investigated the microstructure, texture, and residual stress of AZ31 Mg alloy at several higher extrusion temperatures (360 °C, 380 °C, 400 °C) and speeds (1 m min⁻¹, 2 m min⁻¹, 3 m min⁻¹). Results show that the bimodal microstructure can be observed in all extruded Mg alloys, consisting of the fine grains in dynamic recrystallization (DRX) zone and the coarse grains in non-dynamic recrystallization (non-DRX) zone. The non-monotonic relation between average grain diameter and extrusion speed has been found. It is attributed to the promoted nucleation and inhibited grain growth at higher extrusion speed. The bimodal microstructure can maintain the stability of sharp {0002} basal texture. Schmid Factor (SF) is calculated to explain the mechanism of basal texture formation. By employing XRD with cosθ method, the residual stress has been measured. The major origin of residual stress release at higher extrusion temperature is the grain growth, rather than the strengthening of basal texture. The anisotropy of residual stress distribution is related to the coupling effect of grain growth and evolution of basal texture of extruded Mg alloys.

1. Introduction

The Mg alloys have been considered to be one of the most promising green engineering materials due to the outstanding properties such as low density, high specific strength, characteristic of electromagnetic shielding, and ability to be recycled [1, 2]. Many studies have been carried out to improve the mechanical properties and corrosion resistance of Mg alloys so that they can apply for more areas [3, 4]. At room temperature, Mg alloys exhibit a relatively low formability. It is associated with the limited slip systems of the hexagonal close-packed (HCP) crystal structure [5]. Hot extrusion process has been recognized as an effective way to improve the workability, refine the grain size, and enhance the strength of Mg alloys [6, 7]. Hot extrusion temperature and speed are important process parameters. Although the hot extrusion of Mg alloys has been studied in some detail, there has thus far been little investigation into the precious role of higher extrusion speed. Chen et al [8] suggested that the increased extrusion speed is beneficial to the formability of Mg alloys. Nevertheless, the surface cracks caused by relatively rapid extrusion (0.9 m min⁻¹) have been observed by Lapovok et al [9]. Therefore, a higher extrusion temperature is essential to improve the workability of rapid extruded Mg alloys. The extrusion speed and temperature exhibit contradictory effects. The hardening behaviour is induced by increased extrusion speed, while the softening process is caused by elevated extrusion temperature. It obviously affects the microstructure characteristics, and ultimately determines the mechanical properties of the extruded Mg alloys. The details deserve further study. In this work, the gradient extrusion speed parameter (1 m min⁻¹, 2 m min⁻¹, and 3 m min⁻¹) is employed. Several extrusion temperatures (360 °C, 380 °C, and 400 °C) are used, which are obviously higher than the recrystallization temperature (∼300 °C) [8]. The evolution of microstructure is investigated. The effect of high temperature and speed extrusion on the microstructure and mechanical properties is discussed. This work could provide references for optimizing extrusion parameters and controlling microstructure characteristics of hot extruded Mg alloys. Hot extrusion process facilitates the activation of slip systems and twinning, thus inducing the formation of texture [10, 11]. Despite many
fundamental works on the basal texture in hot extruded Mg alloys, the mechanism of basal texture formation and evolution has not been fully understood. Schmid Factor (SF) is an important criterion for judging the activation of slip systems and twinning [12]. We have calculated SF to theoretically analyse the texture formation mechanism of extruded Mg alloys. The correlation between microstructure features and basal texture intensity is also discussed to give further explanation.

Plastic deformation can induce significant residual stress, which causes the distortional failure, stress-corrosion cracking, and fatigue of the workpiece [13]. As suggested by Withers et al [14], the residual stress is more difficult to predict than the in-service stress. It is valuable to measure residual stress. The cosα method, a novel XRD means for detecting residual stress, has drawn interests [15, 16]. The determined residual stress is based on the measurement of distortion of Debye–Scherrer ring (DSR). The signal of the DSR can be collected by two-dimensional (2D) x-ray plane detector [17]. The data points of all angles (0°–360°) on the DSR can be used to estimate the residual stress to improve the measurement efficiency and reliability. The equation of cosα method is given below:

$$\sigma_x = -\left(\frac{E}{1 + \nu}\right) \frac{1}{\sin 2\Psi_0 \sin 2\theta} \frac{\partial \varepsilon_x}{\partial \cos \alpha}$$

where $\sigma_x$ is residual stress, $E$ is elastic modulus, $\nu$ is Poisson’s ratio, $\Psi_0$ is orientation angle between normal line of sample and input x-ray, $\theta$ is diffraction angle between reflection line and input x-ray, $\varepsilon_x$ is strain, and $\alpha$ is angle at the DSR. The residual stress $\sigma_x$ can be determined from the plot of $\varepsilon_x$ versus $\cos\alpha$, which is obtained from the recorded DSR. We have adopted the XRD with cosα method to calculate the residual stress of extruded Mg alloys. The residual stress distribution on the extruded surface is explored. The microstructure and texture origin of residual stress distribution are also analyzed. It is expected that this work could contribute to the controllable effect of hot extruded parameters, microstructure, texture, and residual stress to improve the mechanical properties of Mg alloys.

2. Experimental

The original billets were AZ31 Mg alloy semi-continuous casting rods. To eliminate possible elemental segregation, the Mg alloys rods were homogenized at 400 °C for 10 h before the hot extrusion process. The Mg alloys rods were extruded at 3 m min⁻¹ with different extrusion temperatures (360 °C, 380 °C, 400 °C), while at 380 °C with different extrusion speeds (1 m min⁻¹, 2 m min⁻¹ and 3 m min⁻¹). The extrusion ratio was kept as 23.58. After hot extrusion process, the Mg alloys sheets with thickness of 9 mm were obtained. The chemical composition of AZ31 Mg alloy used in this study is listed in table 1.

| Element, wt.% | Al  | Zn  | Mn  | Si  | Cu  | Ca  |
|---------------|-----|-----|-----|-----|-----|-----|
| AZ31          | 2.50–3.50 | 0.60–1.40 | 0.20–1.00 | <0.08 | <0.01 | <0.04 |

In order to observe the microstructure of the hot extruded Mg alloys, the standard metallographic samples were prepared. The microstructure characterization was performed by a ZEISS Axio Imager M2m metallographic microscope. Orientation analysis was carried out by a FEI QUANTA FEG450 field emission scanning electron microscope (FE-SEM) equipped with the Channel 5 back-scatter electron diffraction system (EBSD). The texture was examined by x-ray diffraction (XRD). The residual stress distribution along the extrusion direction (ED) and transverse direction (TD) of the compressive planes were determined by XRD with cosα method. The residual stress was measured by μ-X 360 n portable x-ray residual stress analyzer that performed using Cr target as x-ray source target material at the voltage of 30 kV, tube current of 1.2 mA, the x-ray incidence angle of 25°, and sample distance of 39 mm.

3. Results and discussion

3.1. Microstructure of extruded Mg alloys at different extrusion processes

We have investigated the microstructure characteristics of hot extruded Mg alloys. Microstructure analysis area is 25 mm × 12 mm. Figures 1 and 2 show the microstructure of extruded Mg alloys at various extrusion temperatures and speeds. It can be observed that the fine and coarse grains coexist in all extruded Mg alloys, illustrating the bimodal grain morphology. The bimodal microstructure characteristics have been observed in Mg alloys extruded at relatively lower temperatures and speeds [18, 19]. Nevertheless, the increased extrusion temperature and speed can still induce bimodal microstructure, as seen in figures 1(c) and 2(c). The fine grains are newly nucleated ones in the dynamic recrystallization (DRX) zone, while the coarse grains continue growing.
Figure 1. Microstructure of extruded Mg alloys at 3 m min$^{-1}$ with various extrusion temperatures: (a) 360 °C; (b) 380 °C; (c) 400 °C.

Figure 2. Microstructure of extruded Mg alloys at 380 °C with various extrusion speeds: (a) 1 m min$^{-1}$; (b) 2 m min$^{-1}$; (c) 3 m min$^{-1}$. 
in the non-dynamic recrystallization (non-DRX) zone. The microstructure characteristic of extruded Mg alloys strongly depends on the DRX behaviour. The dominant DRX mechanism varies significantly with the increased extrusion temperature and speed. In the early stage of hot extrusion (such as low extrusion temperature and speed), the twinning-induced DRX (TDRX) behaviour tends to occur, where grains nucleate around the twinning. Besides, the continuous DRX (CDRX) is the other major mechanism, where the movement, merger, and absorption of dislocations at subgrain boundaries are the main processes. Compared with TDRX and CDRX, the discontinuous DRX (DDRX) that requires more driving force tends to occur at higher extrusion temperature and speed. During hot extrusion, the transition of DRX mechanism, the region and degree of DRX determine the microstructure characteristics of Mg alloys. Grain boundaries are important locations for DRX\[20,21\]. The diffusion and movement of grain boundaries play an important role in the occurrence of DRX, which deserves more attention to understand the microstructure evolution deeply.

The imaging principle of EBSD is based on the orientation of each grain, which is believed to be an efficient method to analyse the feature of grain boundary. Figures 3 and 4 show the band contrast of extruded Mg alloys at different extrusion temperatures and speeds. The blue lines represent low angle grain boundaries (LAGBs) with the angles from 2° to 15°\[22\]. The black lines refer to high angle grain boundaries (HAGBs) with the angles more than 15°\[23\]. As seen in figure 3, LAGBs is sensitive to extrusion temperature. The proportion of LAGBs decrease gradually with increasing extrusion temperature. Especially at higher extrusion temperature (400 °C), LAGBs are almost invisible, while the fraction of HAGBs is relatively larger (figure 3(c)). It indicates that higher extrusion temperature induces most of the grains to coarsen although small DRX areas can still be observed. As seen in figure 4, the effect of increased extrusion speed on grain boundaries is more complicated. The fraction of LAGBs is relatively smaller at 2 m min\(^{-1}\) (figure 4(b)). Whereas more LAGBs and HAGBs can be observed at 3 m min\(^{-1}\) (figure 4(c)), which may indicate more grain nucleation is induced in DRX zone.

To analyse the deformation behaviour of extruded Mg alloys statistically, we have calculated the average diameter and number of grains at different extrusion parameters. The counted grains are at least 400 to ensure the repeatability of the results. As shown in table 2, the average grain diameter increases while the number of grains reduces at elevated extrusion temperature, showing the tendency of grain growth with increasing extrusion temperature. Different from the high-temperature extrusion, providing a driving force for the migration and diffusion of grain boundaries, the role of high extrusion speed on the microstructure still needs to be studied further. We have examined the size and number of grains of extruded Mg alloys at different extrusion parameters.
speeds, as given in Table 3. As the extrusion speed increases, the average grain diameter first increases and then decreases, which is accompanied with the opposite change trend of the number of grains. It indicates that the average grain size is not monotonous with the increased extrusion speed. The elevated extrusion speed can cause the uneven stress and high-density dislocation, which induces the nucleation of grains. In addition, the high-speed extrusion shortens the deformation time, making the grain boundary diffusion insufficient, thus inhibiting the grain growth. Therefore, we believe that the high-speed extrusion has a contradictory effect on the degree of DRX, that is, promoting nucleation and inhibiting growth. This result is different from the work [24], where the coarse grains are observed in DRX zone at lower speed extrusion (3.5 mm s⁻¹). As described in many reports [25, 26], the improved strength is strongly dependent on the reduced crystalline size. The refined grains are accompanied with more grain boundaries that can also induce the transfer of stress. It facilitates a uniform distribution of stress [27]. In contrast, the coexistence of fine and coarse grains resulted from the insufficient

| Extrusion temperature | 360 °C | 380 °C | 400 °C |
|-----------------------|-------|-------|-------|
| Average grain diameter/μm | 24.66 (±9.20) | 30.37 (±8.69) | 38.01 (±18.16) |
| Number of grains | 1216 | 918 | 413 |

Table 3. Size and number of grains of extruded Mg alloys at 380 °C with various extrusion speeds. The standard deviation is shown in parentheses.

| Extrusion speed | 1 m min⁻¹ | 2 m min⁻¹ | 3 m min⁻¹ |
|----------------|----------|----------|----------|
| Average grain diameter/μm | 30.39 (±12.23) | 33.97 (±15.01) | 30.37 (±8.69) |
| Number of grains | 764 | 569 | 918 |
3.2. Texture of extruded Mg alloys at different extrusion processes

We have performed texture analysis of extruded Mg alloys by using XRD. The surface size of the analysis sample is 25 mm × 12 mm. Figure 5(a) shows the XRD result of extruded Mg alloys at different extrusion temperatures. It is obvious that the diffraction peak of the {0002} crystal plane is the strongest. The diffraction peaks of other non-basal planes with lower intensity can also be observed, including {1010}, {1011}, {1012}, {1120}, {1013}, {1122}, and {2021}. As the extrusion temperature elevates, the diffraction peak intensity of the {0002} crystal plane increases. The decreased diffraction intensity of non-basal planes can also be observed, such as {1010}, {1011}, {1012}, and {1122}. It indicates that the extruded Mg alloys still has a clear preferred orientation of the {0002} plane at higher extrusion temperature. Figure 5(b) shows the XRD result of extruded Mg alloys at different extrusion speeds. It is seen that the diffraction peak intensity of the {0002} crystal plane increases with rising extrusion speed, similar to the tendency at elevated extrusion temperature. In addition, higher extrusion speed induces the formation of several non-basal orientations.

XRD results show that the {0002} basal texture may be dominant in extruded Mg alloys. To analyse the evolution of basal texture in extruded Mg alloys deeply, we have calculated the {0002} pole figure of extruded Mg alloys at different parameters. The {1010} pole figure has also been calculated to provide a clearer comparison. As the extrusion temperature increases from 360 °C to 400 °C (figure 6), the maximum polar density value of the basal texture is 21.53, 22.04, and 26.08, respectively. It means that the elevated extrusion temperature causes an increase in the intensity of basal texture. When the extrusion temperature is 400 °C, the isopycnal lines appear in the center of the {1010} pole figure, indicating the existence of non-basal texture. As shown in figure 7, it can be observed that as the extrusion speed increases from 1 m min⁻¹ to 3 m min⁻¹, the maximum polar density value of the basal texture is 16.04, 20.09, and 22.04, respectively, showing a similar trend with increased extrusion temperature. The pole figure analysis suggests the main basal texture components of the hot extruded Mg alloy.

The correlation between basal texture and mechanical properties of Mg alloys has been paid attention. It is reported that the intensive basal texture improves the yield strength [30]. Yu et al [31] suggested that the texture strengthening mechanism can be explained by the Hall-Petch relation. Khdair et al [32] believed that the strong texture leads to a significant increase in the proportion of LAGBs, thereby improving the deformation coordination of adjacent grain boundaries. Sadoun et al [33] reported that the grains with strong texture require more frictional energy to deform further, thus delaying fracture. It is worthy to alter the intensity of basal texture to improve the machanical properties of hot extruded Mg alloys. Although the strong basal texture of extruded Mg alloys has been observed by many works [34–36], the mechanism of basal texture formation and evolution is still unclear. Two main factors are discussed in this work. On the one hand, the change in basal texture intensity is related to the bimodal microstructure features. Although a relatively weaker basal texture is exhibited by the fine DRXed grains, the sharp basal texture of un-DRXed grains can still retain stable. It means that the grain growth with preferred crystallographic orientation helps to increase the texture intensity [37–39]. Therefore, an obvious basal texture can be observed at higher hot extrusion temperature and speed.
On the other hand, the SF is calculated to theoretically analyse the formation mechanism of basal texture of extruded Mg alloys. Generally, the deformation mechanisms of Mg alloys include the \( \langle \alpha \rangle \) slip on basal \{0002\}, prismatic \{10\bar{1}0\} and pyramidal \{10\bar{1}1\} planes, as well as the \( \langle \alpha + c \rangle \) slip on pyramidal \{11\bar{2}2\} planes, and two twinning modes of \{10\bar{1}2\} extension twinning and \{10\bar{1}1\} compression twinning [40, 41]. Figure 8 shows the schematic relation between loading stress and different deformation mechanisms, where loading directions are along \( \{10\bar{1}0\} \). The grains are subjected to compressive stress in different directions. \( \theta \) is the angle between the stress direction and the \( c \) axis, the range of which is \( 0^\circ \) to \( 90^\circ \). It is assumed that the stress direction rotates on a plane parallel to \( \{10\bar{1}0\} \). The selected planes and directions of slips and twins have the largest SF under the compressive stress.

The calculation formula of SF for each deformation mechanism in the extrusion process is as follows:

\[
SF = \cos \lambda \cos \varphi
\]

where \( \varphi \) and \( \lambda \) are the angles between the loading directions and the slip (or twinning) planes normal and slip (or twinning) directions, respectively.

In the process of metal plastic deformation, the start of twins and slip systems usually follows Schmid law, that is, the microscopic deformation mode with large SF starts first. For a specific deformation mode, the
activation depends on the CRSS and SF values. For various microscopic deformation modes of Mg alloys, the CRSS value of the extension twin is the smallest, far below prismatic and pyramidal slip systems. But the twin start exhibits a unidirectional characteristic, that is, under the condition that the grain orientation and load direction are determined, it can only be sheared in a single direction, different from the starting characteristics of a sliding mode system. Therefore, the value range of SF corresponding to twins is $[-0.5, 0.5]$, and the value range of SF corresponding to slip systems is $[0, 0.5]$.

Figure 9 shows the calculated curves of SF with $\theta$ angle. As seen from figure 9(a), along the $\langle 0001 \rangle - \langle 11\bar{2}0 \rangle$ orientation line, for the three slip systems of basal slips, the SF values of $(0001)[\bar{1}\bar{2}10]$ and $(0001)[\bar{1}\bar{2}10]$ are equal, and the SF value of $(0001)[\bar{1}\bar{2}10]$ is close to 0.5 at $\theta = 45^\circ$. For the three slip systems of prismatic slip in figure 9(b), $(10\bar{1}0)[\bar{1}\bar{2}10]$ and $(0\bar{1}\bar{1}0)[\bar{1}\bar{2}10]$ have the same SF values and the maximum is close to 0.433 at $\theta = 90^\circ$, while the SF value of $(11\bar{0}0)[\bar{1}\bar{2}10]$ is always 0. For the six slip systems of pyramidal $\langle a \rangle$ slip shown in figure 9(c), $(1\bar{1}01)[1\bar{1}20]$ and $(1\bar{1}01)[\bar{1}\bar{2}10]$, $(1\bar{1}01)[\bar{1}\bar{2}10]$ and $(01\bar{1}1)[\bar{2}1\bar{1}0]$, $(01\bar{1}1)[\bar{1}\bar{2}10]$ and $(01\bar{1}1)[\bar{2}1\bar{1}0]$ have the same SF values, and the maximum value of $(10\bar{1}1)[\bar{1}\bar{2}10]$ and $(01\bar{1}1)[\bar{2}1\bar{1}0]$ is close to 0.406 when $\theta = 70^\circ$. For the six slip systems of pyramidal $(a + c)$ slip (figure 9(d)), the SF values of $(12\bar{1}2)[12\bar{1}3]$ and $(2\bar{1}12)[\bar{2}1\bar{1}3]$ are respectively equal, the SF values of $(1\bar{1}22)[\bar{1}\bar{2}3]$ and $(\bar{1}\bar{2}22)[1\bar{2}\bar{3}]$ are not equal, and $(12\bar{1}2)$

\[Figure 7. Pole figure of extruded Mg alloys at 380 °C with various extrusion speeds: (a) 1 m min$^{-1}$; (b) 2 m min$^{-1}$; (c) 3 m min$^{-1}$.\]
[1213] of the maximum value is close to 0.472 when $\theta = 10^\circ$. For the six extension twin systems in figure 9(e), (1012)[0111] and (0112)[0111], [1102][1101] and (1102)[1101], (0112)[0111] and (0112)(0011) have the same SF values, and the SF value decreases with the increase of $\theta$, in which the maximum value is 0.499 at $\theta = 0^\circ$. For the six compression twin systems shown in figure 9(f), (1011)[1012] and (1101)[1012] and (0111)[0112], (0111)[0112] and (0111)(0011) have the same SF values, and the SF value of (0111)[1012] and (0111)[0112] first increases with the increase of $\theta$, and then decreases after the maximum value is 0.499 at $\theta = 70^\circ$.

Taking into account the influence of SF as well as an appropriate value of $\theta$, the slip and twinning systems mentioned above can be activated.

According to Taylor model [42], the activation of the Mg alloys slip system is caused by its value of CRSS, and the corresponding slips amount is determined by the dislocation characteristics of the slip planes. Due to the blocking effect of the extrusion wall, the grains on the extrusion surface lack fluidity compared with the internal grains. During plastic deformation, the grains can only flow in the ED and tilt itself. The plastic deformation trend of the extruded Mg alloy is that the lattice spacing along the ED reaches the minimum, meanwhile, the c-axis is perpendicular to the ED to be the most stable state. The deformation changes the angle between the c axis of grain and the stress direction through the pyramidal (a + c) and basal slip, finally reaching $\theta = 90^\circ$. At the same time, the basal plane is parallel to the ED, and the prismatic slip dominates plastic deformation. Although the non-basal slip and twinning have not been studied deeply in this work, the calculated SF and Taylor model can help to further understand the deformation behavior of Mg alloys. Our work provides the theoretical basis for the activation of basal slip and the formation of basal texture.

### 3.3. Residual stress of extruded Mg alloys at different extrusion temperatures

We have measured the residual stress of the compressive planes at elevated temperatures by using XRD with cos$\circ$ method. Figure 10 shows the DSR of extruded Mg alloys at different extrusion temperatures. The diffraction intensity distribution of the circumference of DSR is uneven, indicating the existence of texture. At higher extrusion temperature, the inhomogeneous distribution of diffraction intensity is more obvious, implying that the texture is getting stronger. It is consistent with the results of pole figure analysis. We thus believe that DSR is also an effective method for texture characterization. Figure 11 shows the plot of $\varepsilon_a$ versus cos$\circ$ for extruded Mg alloys. According to the equation (1), the derivative of each point of the $\varepsilon_a$-cos$\circ$ curve is proportional to the value of residual stress. The average value of the derivative is used to calculate the residual stress to improve the validity. The obtained residual stress values are $-58$ MPa ± 12 MPa, $-46$ MPa ± 6 MPa, $-25$ MPa ± 2 MPa for extruded Mg alloys at 360°C, 380°C, 400°C, respectively. It indicates that the compressive residual stress is getting lower with increasing extrusion temperature. The $\varepsilon_a$-cos$\circ$ curve deviates slightly from the ideal linear relation. The origin of the uncertainties associated with the method should be paid attention. Firstly, the residual stress is calculated by measuring the lattice strain. The generated lattice strain depends on the interaction of the macro- and microscopic residual stress and lattice distortion, which is not distinguished by x-ray. This is the major factor. Next, there is a slight deviation in the measuring and conversion of $\Psi_0$ and $\alpha$. Finally, the inhomogeneous DSR can be caused by existence of strong texture and large grains.

We have examined the residual stress distribution of extruded Mg alloys at different extrusion temperatures. Multi-point detection is used to improve the reliability of the results. The scrutiny direction is along ED and TD. The detection interval is 2 cm. The area of 8 cm × 8 cm is tested. As shown in figure 12, the residual stress distribution along ED reduces with increasing temperature. Moreover, we have not found an obvious

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**Figure 8.** The schematic relation between loading stress and different deformation mechanisms, where loading directions are along (112c) (a) projected location of relative loading direction on the basal plane; (b) slip mechanism; (c) twinning mechanism.
Figure 9. The SF of different deformation mechanisms (a) basal; (b) prismatic; (c) pyramidal (a); (d) pyramidal (a + c) slips; (e) (1012) extension twinning; and (f) (1011) compression twinning as a function of the $\theta$ angles.

Figure 10. The DSR of extruded Mg alloys at 3 m min$^{-1}$ with different extrusion temperatures: (a) 360 $^\circ$C; (b) 380 $^\circ$C; (c) 400 $^\circ$C.
dependence of residual stress along TD on extrusion temperature. It implies the anisotropic residual stress distribution of extruded Mg alloys. The anisotropic residual stress causes a strong distortion of the specific crystal plane, which induces anisotropy of the mechanical properties. It is reported that compressive residual stress is beneficial to hinder surface stress corrosion cracking [43]. We believe that the anisotropic compressive residual stress can significantly delay the crack propagation along ED, which improves the service performance of Mg alloys in a specific direction. It is meaningful to adjust the residual stress distribution to alter the mechanical properties of extruded Mg alloys. Wang et al [44] suggested that the anisotropy of residual stress is related to the basal texture that causes the anisotropic flow stress. It has been observed that the decrease in the strength of {0002} basal texture is accompanied by a reduction of residual stress [16]. It not agrees with our result, where the opposite trend is presented. The role of microstructure characteristics should be considered. The inverse relation between grain size and lattice strain has been proposed [45, 46]. High extrusion temperature induces grain coarsening, which is believed to be effective in releasing residual stress [47]. Moreover, a higher extrusion temperature facilitates the diffusion of elements and movement of dislocations, thus reducing the

Figure 11. The plot of $\varepsilon_a$ versus $\cos \alpha$ for extruded Mg alloys at 3 m min$^{-1}$ with different extrusion temperatures: (a) 360 °C; (b) 380 °C; (c) 400 °C.

Figure 12. The residual stress distribution of extruded Mg alloys at 3 m min$^{-1}$ with different extrusion temperatures: (a) ED; (b) TD.
residual stress. Therefore, we believe that the bimodal grain growth is an important reason for the reduction of residual stress. The anisotropic residual stress distribution is related to the coupling effect of grain growth and basal texture strengthening for extruded Mg alloys.

4. Conclusions

In this work, the evolution of microstructure, texture, and residual stress of Mg alloys at several higher extrusion temperatures and speeds has been investigated. The correlation between microstructure, texture, residual stress, and mechanical properties is discussed. This work is expected to provide references for optimizing extrusion parameters, controlling microstructure and texture features, adjusting residual stress distribution, and improving mechanical properties of hot extruded Mg alloys. The conclusions can be stated as follows:

1) All the extruded Mg alloys exhibit the bimodal microstructure. The coexistence of fine grains in DRX zone and coarse grains in non-DRX zone is confirmed. We have found the non-monotonic relation between grain size and extrusion speed. It is attributed to the contradictory effect of higher speed extrusion on the degree of DRX, namely, promoting nucleation and inhibiting growth.

2) The extruded Mg alloys exhibit a strong \{0002\} basal texture at higher extrusion temperature and speed. The calculated SF value suggests the activation of basal slip and the formation of basal texture, which provides a theoretical basis for the formation of basal texture.

3) The extrusion process causes the compressive residual stress in the extruded Mg alloys. The residual stress reduces along ED with increasing extrusion temperature, which is due to the grain growth instead of the strengthening of basal texture. The anisotropy of residual stress distribution has been confirmed and it is related to the coupling effect of grain growth and evolution of basal texture.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Conflicts of interest

The authors declare no competing financial interest.

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