Microstructure and texture evolutions in AZ80A magnesium alloy during high-temperature compression

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Abstract
In the process of thermal deformation, the microstructure and texture evolutions of AZ80A magnesium alloy have an important influence on its properties. In order to reveal the evolutions of microstructure and texture of AZ80A magnesium alloy during hot deformation, isothermal compression tests were performed on a Gleeble-3800 thermal simulator at a temperature range of 623 K–723 K, a strain rates range of 0.001 s\textsuperscript{−1}–1 s\textsuperscript{−1}, and a deformation rate of 20%–60%. Based on the analysis of the true stress-true strain curves under the given deformation conditions, the hyperbolic sinusoidal constitutive model and dynamic recrystallization (DRX) kinetics model were established by regression method, respectively. The DRX behavior, microstructure and texture evolutions during the thermal compression were then analyzed by optical microscopy (OM) and electron backscattered diffraction (EBSD). The results show that an increasing deformation temperature and decreasing strain rate greatly support the development of DRX of AZ80A magnesium alloy. As the deformation rate increases, DRX of AZ80A alloy is gradually and more sufficiently developed, and the dominated deformation mechanism is gradually transformed from the \{0001\} base slip to \{1010\} cylinder slip. Moreover, the low-angle grain boundaries (between 5° and 75° in deformation rate of 20%) gradually transform into large-angle grain boundaries (between 50° and 90° in deformation rate of 60%). Besides, the texture is gradually transformed from a base texture paralleled to compression direction to a texture perpendicular to compression direction, where twinning, rather than slip, is the main deformation mechanism. The results in this study provides a guidance in the hot forming process of AZ80A magnesium alloy.

1. Introduction

In recent years, with industrial and economic growth worldwide, CO\textsubscript{2} emissions has been increasing in recent decades, resulting in more serious greenhouse effect and thus, an environmental degradation. In order to reduce CO\textsubscript{2} emissions, the use of lightweight alloys such as, Mg, Al, and Ti, is increasingly being recognized [1]. Magnesium alloys are considered as irreplaceable lightweight alloys, due to the advantages of non-toxic, easy-to-recovery, low density, good damping performance, high specific strength, and specific stiffness. Hence, magnesium alloys are known as ‘green engineering materials in the 21st century’ [2]. Magnesium alloys are considered to have great application prospects in the fields of transportation, aerospace, and electronic 3C industries [3, 4], which has attracted the attention of many researchers worldwide. Among these alloys, AZ80 magnesium alloy has been considered as one of the research hotspots.

In terms of heat treatment, Tang [5] and Xu [6] studied the effect of the heat treatment process of AZ80 magnesium alloy on its mechanical properties, and they believed that \(\beta\) phase greatly reduced the tensile strength of the alloy. Wang et al [7] studied the effect of aging treatment on the microstructure and mechanical properties of AZ80N magnesium alloy. Precipitation hardening of AZ80N alloy was found, due to the precipitations of Al\textsubscript{17}C\textsubscript{3} and Mg\textsubscript{17}Al\textsubscript{12} phases after solution treatment at 400 °C and aging treatment at 170 °C and 250 °C.
In terms of plastic deformation, Su [8], Quan [9, 10] and Li [11] established the constitutive model and DRX model of AZ80 magnesium alloy during isothermal deformation and discussed the microstructure evolution during DRX, and they thought that DRX was the main softening behavior of the alloy during hot deformation. Ming [12] and Honda [13] built an equation with a relation between DRX grain size and the Zener–Hollomon parameter of AZ80 magnesium alloy. Chen et al [3] established a constitutive model of AZ80 magnesium alloy during hot compression based on Arrhenius and Johnson–Cook model. They found that the constitutive relation before the peak strain could be reflected by the modified Arrhenius-type relation and the modified Johnson–Cook model was suitable to predict the stage after peak strain.

In terms of microstructure evolution and texture, Zhang et al [14] used electron backscattered diffraction (EBSD) to study the anisotropy and microstructure evolution of rolled AZ80 magnesium alloy at room temperature, and they believed that, as the deformation increased, the amount of low angle grain boundaries increased. Kim et al [15] studied the formation of texture of rolled AZ80 magnesium alloy sheet during thermal deformation by plane strain compression test. They found that the texture was related to the initial texture. Tang et al [16] discussed the influence of strain path on the evolution of microstructure and texture in the extruded AZ80 magnesium alloy during hot compression. They found that the flow stress obviously depended on changes in the strain path, and the texture components gradually changed from (1210) (0001) and (0110) (0001) in single-pass deformation to (0001) (0110) and (0001) (1110) in four-pass deformation due to DRX and meta-DRX. Dong et al [17] performed the rotating backward extrusion experiments with different rotating revolutions on the AZ80 alloy. They found that, with increasing rotating revolutions, the average grain size decreased and the proportion of dynamic recrystallization (DRX) was gradually increased, whereas, the texture was weakened. Qin et al [18] performed a series of compression experiments along extrusion direction (ED) and radius direction (RD) in the as-extruded AZ80 magnesium alloy. They found that the microstructures of the alloy compressed along ED were more homogeneous than that of the alloy compressed along RD. The texture of the alloy compressed along ED tended to rotate towards the (0001) // ED basal texture but its total intensity weakened. However, the (1010) // ED and (1120) // ED fiber texture of the alloy compressed along RD presented a sharpening phenomenon. Prakash et al [19] investigated the microstructure and texture evolutions of the as-cast AZ80 alloy under different deformation conditions. They found that the thermodynamic stability of the Mg17Al12 phase varied over the test temperature range and significantly impacted the microstructure and texture evolutions at different test temperature.

As mentioned above, extensive research has been carried out, and many constructive conclusions have been presented to guide the hot deformation behavior of AZ80 magnesium alloy. However, few reports on the evolutions of microstructure and texture of AZ80 magnesium alloy during hot compression after homogenization treatment were studied and analyzed by EBSD technology. In this work, AZ80A magnesium alloy after homogenization treatment was chosen as the research material. Hyperbolic sinusoidal constitutive model and DRX kinetics model of AZ80A magnesium alloy were established by calculation from the data of isothermal compression tests. The effects of temperature, strain rate, and deformation rate on the DRX microstructural evolution were investigated. Meanwhile, the DRX behavior and texture evolutions of the alloy during hot compression were revealed by OM and EBSD.

2. Materials and experimental procedures

The chemical composition (wt%) of the as-cast AZ80A magnesium alloy was shown as follows: Al, 8.15; Zn, 0.65; Mn, 0.35; Si, 0.05; Cu, 0.001; Ni, 0.001; Fe, 0.002 and Mg, balance. As shown in figure 1 [20], the microstructures of the as-cast AZ80A magnesium alloy, which consists of α-Mg matrix and β-Mg17Al12 phase, which was distributed in networks. The initial as-cast AZ80A magnesium alloy was homogenized in a box-type resistance furnace at 683 K for 20 h and the average grain size was 203 μm, as shown in figure 2 [20].

Quantitative metallographic analysis technique was required to measure the grain size [21]. The calculation formula was shown as follows:

$$d = \frac{1}{n} \sum_{i=1}^{n} \frac{L_i \times \bar{a}}{N_i \times a'}$$

where, n is the number of measurements, L_i is the measured length of the ith straight line, N_i is the number of grains through which the ith stroke is drawn, a is the measured length of the scale, a’ is the actual length of the scale. In this work, the metallographic photographs were measured five times in different fields of view, and the average value was determined as the grain size.

Cylindrical specimens in dimension of φ10 × 15 mm were cut from the AZ80A magnesium alloy after homogenization treatment. To study the DRX behavior of AZ80A magnesium alloy, Gleeble-3800 thermal simulation machine was used for isothermal compression tests. Before the tests, high-temperature lubricating
oil was sprayed on the end of the anvils and graphite gaskets were coated on the anvils. The hydraulic press was used to clamp the specimen. The tests were classified into two cases. For case 1, similar with our previous research \[20\], the specimens were hot compressed at the temperatures of 623 K–723 K, the strain rates of \(0.001 \text{ s}^{-1} \sim 1 \text{ s}^{-1}\), and the deformation rate of 60\%, respectively. For case 2, the specimens were hot compressed at the temperature of 673 K, the strain rate of 0.1 \text{ s}^{-1}, and the deformation rates of 20\%, 30\%, 40\%, 50\%, and 60\%, respectively. Figure 3 exhibited hot deformation process diagram of the AZ80A magnesium alloy. First, the specimen was heated to the preset temperature at a heating rate of 10 \text{ K s}^{-1}. After kept the preset temperature for 180 s, the hot compression test was conducted. After the test, an immediate water quench was followed to retain the high-temperature deformed microstructure.

The specimen was sectioned and the metallography of the specimen was obtained by etching of a mixed solution, which contains 1 ml nitric acid, 1 ml acetic acid, 1 ml oxalic acid, and 150 ml water. The EBSD sample was electropolished in a 10\% perchloric acid of alcohol solution at room temperature. During the etching
process, a constant current of 1 A was used, where the voltage was in range of 12~15 V, and the corrosion time was 15~20 s.

3. Results and discussion

3.1. Constitutive model

The relationship among flow stress, temperature, and strain rate during isothermal deformation can be expressed by the hyperbolic sinusoidal function proposed by Sellars and Tegart [22], which is shown as follows:

\[
\dot{\varepsilon} = A [\sinh(\alpha \sigma)]^n \exp \left( -\frac{Q}{RT} \right) \quad \text{(for all } \sigma) \tag{2}
\]

At the same time, it can also be expressed as follows:

\[
\dot{\varepsilon} = A_1 \sigma^n \exp \left( -\frac{Q}{RT} \right) \quad (\alpha \sigma < 0.8) \tag{3}
\]

\[
\dot{\varepsilon} = A_2 \exp(\beta \sigma) \exp \left( -\frac{Q}{RT} \right) \quad (\alpha \sigma < 1.2) \tag{4}
\]

where, \( Q \) is the activation energy (KJ mol\(^{-1}\)), \( T \) is the absolute temperature(K), \( R \) is the gas constant (8.314 J/mol/K), \( A \), \( A_1 \), \( A_2 \), \( n \), \( n_1 \), \( \alpha \), \( \beta \) are material constants.

Combining equations (3)–(5) is obtained:

\[
\alpha = \beta / n_1 \tag{5}
\]

At a certain temperature, \( \partial \ln \dot{\varepsilon} \) and \( \partial \ln [\sinh(\alpha \sigma)] \) can be fitted with a slope of \( 1/n \), and \( n \) can be found.

\[
\frac{1}{n} = \left[ \frac{\partial \ln [\sinh(\alpha \sigma)]}{\partial \ln \dot{\varepsilon}} \right]_T \tag{6}
\]

During the isothermal deformation of the alloy, the function between temperatures and strain rates can be expressed by the Zener-Hollomon parameter [23].

\[
Z = A [\sinh(\alpha \sigma)]^n = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) \tag{7}
\]

Combining equations (2) and (4), (8) is obtained:

\[
\frac{Q}{RT} = \ln A - n \ln \dot{\varepsilon} + n [\ln [\sinh(\alpha \sigma)] \tag{8}
\]

Combining equations (3) and (5), (9) is obtained:

\[
Q = -R \left[ \partial \ln [\sinh(\alpha \sigma)] / \partial (1/T) \right] \left[ \partial \ln \dot{\varepsilon} / \partial \ln [\sinh(\alpha \sigma)] \right]_T \tag{9}
\]

Based on the stress-strain curves in our previous research [20] shown in figure 4 and the above equations, the values of \( A \), \( n \), \( \alpha \), \( Q \) can be obtained. According to equations (1)–(3), \( n_1 \), \( \beta \) and \( \alpha \) can be presented as follows:

\[
n_1 = \left[ \frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sigma} \right]_T, \quad \beta = \left[ \frac{\partial \ln \dot{\varepsilon}}{\partial \sigma} \right]_T, \quad \alpha = \beta / n_1.
\]

By linear fitting of the relationship between \( \ln \dot{\varepsilon} \) and \( \ln [\sinh(\alpha \sigma)] \), the values of \( n_1 \) and \( \beta \) can be obtained as shown in figure 5. The value of \( \alpha \) can be estimated by \( \alpha = (\sum_{i=1}^{5} \alpha_i) / 5 \), with the calculation result of \( \alpha = 0.0244 \). From equation (9), \( Q \) can be determined by fitting of the relation between \( \ln \dot{\varepsilon} \) and \( \ln [\sinh(\alpha \sigma)] \) and \( 1000/T - \ln [\sinh(\alpha \sigma)] \) as shown in figure 6, and the value was 173.8 KJ/mol. By substituting different temperatures and different strain rates into equation (7), the \( Z \) values under various deformation conditions can be solved. By linear fitting of the relationship between \( \ln Z \) and \( \ln [\sinh(\alpha \sigma)] \), the values of \( A \) and \( n \) can be obtained, as shown in figure 7, where \( A = 4.9721 \times 10^{11} \) and \( n = 3.8647 \). In summary, the constitutive model of the as-cast AZ80A magnesium alloy after homogenization treatment can be expressed as:

\[
\dot{\varepsilon} = 4.9721 \times 10^{11} \left[ \sinh(0.0244 \sigma_T) \right]^{3.8647} \exp \left( -\frac{173800}{8314T} \right) \tag{10}
\]
3.2. DRX microstructures evolution

3.2.1. Effect of temperature on DRX microstructure

During the deformation process, temperature has a significant effect on energy, dislocation activity, and grain boundary mobility which then affects DRX behavior and deformation microstructures. Figure 8 shows the microstructures of AZ80A magnesium alloy at the strain rate of $1 \text{s}^{-1}$ and the temperature of 623 K, 673 K, and 723 K, respectively. As shown in figure 8, it is observed that, as the deformation temperatures increases, the DRX is gradually and more sufficiently developed, and the DRX grain size increases. The reason is that the increase in temperature provides sufficient energy for dislocation motion. The slip and migration of the dislocations are easier, which provide sufficient driving force for the migration of the grain boundaries and promote the nucleation and growth of DRX grains. Therefore, the increased temperature is beneficial to the occurrence and development of DRX.

Figure 4. Stress-strain curves of homogenized alloy under various deformation conditions: (a) $0.001 \text{s}^{-1}$, (b) $0.01 \text{s}^{-1}$, (c) $0.1 \text{s}^{-1}$, (d) $1 \text{s}^{-1}$.

Figure 5. Linear fitting relationship of alloy between the peak stresses and strain rates (a) $\ln \dot{\varepsilon} - \ln \sigma_p$, (b) $\dot{\varepsilon} - \sigma_p$. 
3.2.2. Effect of strain rate on DRX microstructures

Figure 9 shows the microstructure of AZ80A magnesium alloy at the temperature of 623 K and the strain rates of 0.1 s\(^{-1}\), 0.01 s\(^{-1}\), 0.001 s\(^{-1}\). As shown in figures 9 and 8(a), as the strain rate decreases, the DRX grain size is getting larger. The reason is that as the decrease in strain rate leads to the increase in deformation time, the dislocations in grains of the alloy have sufficient time to slip and climb. On the contrary, a higher strain rate results in a shorter deformation time and thus, a higher dislocation density. As a result, the DRX grains do not have enough time to grow up. At a strain rate of 1 s\(^{-1}\), as shown in figure 8(a), DRX grains are concentrated at the grain boundaries. The reason is that, when the strain rate is large, the strain energy is high, leading to a high nucleation rate. Furthermore, due to the large grain size of the initial microstructure, the amount of grain boundaries is less and the energy is high at the grain boundaries. The new grains nucleate easily at grain boundaries. Therefore, the DRX grains formed are concentrated at the grain boundaries.
3.2.3. Effect of deformation on DRX microstructure

Figure 10 shows the microstructures of AZ80A magnesium alloy at different deformation rate at the strain rate of 0.1 s\(^{-1}\) and the temperature of 673 K. It is observed that the DRX volume fraction increases as the deformation rate increases. Hence, the amount of DRX grains increases reveals a more sufficiently development of DRX with increasing the deformation rate. As shown in figure 10(a), when the deformation rate is 30%, the DRX grains first nucleate at the grain boundaries. The reason is that, at the initial stage of deformation, the dislocation density near the grain boundaries is high, resulting in accumulation and entanglement of dislocations, where stress concentration occurs. This mechanism leads to an easier nucleation of DRX at the grain boundaries. The deformed coarse parent grains and the fine DRX grains are observed, forming a 'necklace-like' DRX microstructure. Moreover, the parent grains are perpendicular to the compression direction. As shown in figure 10(b), when the deformation rate reaches 40%, more DRX grains can be observed around the deformed parent grains. When the deformation rate reaches 50%, as shown in figure 10(c), the DRX volume fraction

![Figure 9. Microstructures at the temperature of 623 K, the deformation rate of 60%, and various strain rates of (a) 0.1 s\(^{-1}\), (b) 0.01 s\(^{-1}\), (c) 0.001 s\(^{-1}\).](image)

![Figure 10. Microstructures at the strain rate of 0.1 s\(^{-1}\), the temperature of 673 K and various deformation rate of (a) 30%, (b) 40%, (c) 50%, (d) 60%.](image)
further increases, which reveals more sufficient development of DRX. When the deformation rate reaches 60%, as shown in figure 10(d), a complete development of DRX and grains growth can be observed, which presents a uniform microstructure.

3.3. DRX kinetics

3.3.1. DRX critical condition

The relationship between the work hardening (WH) rate (θ) and the flow stress based on the thermal compression stress-strain curves can be expressed as [24]:

$$\theta = \frac{d\sigma}{d\varepsilon}$$  \hspace{1cm} (11)

As shown in figure 11, the DRX parameters can be solved by the θ-σ curves. In stage I, the θ rapidly decreases to critical stress ($\sigma_c$), and the θ is greater than zero. The $\sigma_c$ can be obtained from the inflection point of the $\sigma$-curve [25, 26] before the peak stress. In stage II, as the deformation continues, the θ gradually decreases to zero, and the stress reaches $\sigma_p$. During the deformation process, the softening mechanism, such as DRV and DRX, gradually dominates, balancing the HW effect. In stage III, the θ gradually decreases to the minimum value which is corresponding to the maximum value of softening rate stress ($\sigma^*$). Finally, the stress reaches the stable-stress ($\sigma_{ss}$), which is supposed to be a constant.

According to the above method, $\sigma_c$, $\sigma_p$, $\sigma^*$, $\sigma_{ss}$ and the corresponding $\varepsilon_c$, $\varepsilon_p$, $\varepsilon^*$, $\varepsilon_{ss}$ values under different deformation conditions can be obtained.

$\varepsilon_c$, $\varepsilon_p$, $\varepsilon^*$ can usually be expressed as dimensionless parameter ($Z/A$) function expression [27, 28]. By fitting the relationship between $\ln \varepsilon_c$, $\ln \varepsilon_p$, $\ln \varepsilon^*$ and $\ln(Z/A)$, the formula of DRX critical condition can be derived. Their simplified forms can be expressed as:

$$\varepsilon_c = 0.03134(Z/A)^{0.15407}$$  \hspace{1cm} (12)
$$\varepsilon_p = 0.13263(Z/A)^{0.15219}$$  \hspace{1cm} (13)
$$\varepsilon^* = 0.0419(Z/A)^{0.30418}$$  \hspace{1cm} (14)

Where $Z = \dot{\varepsilon} \exp \left( \frac{173800}{8347} \right)$ and $A = 4.9721 \times 10^{11}$. $\varepsilon_c$ and $\varepsilon_p$ can normally be expressed as $\varepsilon_c = k\varepsilon_p$, the range of $k$ value is 0.2128~0.6667 [29]. In this paper, $\varepsilon_c$ and $\varepsilon_p$ are formulated as $\varepsilon_c = 0.2363\varepsilon_p$.

3.3.2. DRX kinetics model

With increasing strain, dislocations are proliferating in the alloy, which increases the WH. When the dislocation density reaches the critical value of DRX, the DRX occurs. The modified Avrami equation can be used to express the DRX kinetics model when the strain rate and temperature are constants [30, 31].

$$X_{DRX} = 1 - \exp \left\{ -p \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right)^m \right\}$$  \hspace{1cm} (15)

where, $X_{DRX}$ is the volume fraction of DRX, $m$ and $p$ are constants.

Take the logarithm of both sides of equation (15),

$$\ln[-\ln(1 - X)] = \ln p + m \ln[(\varepsilon - \varepsilon_c)/\varepsilon_p]$$  \hspace{1cm} (16)

The DRX volume fractions at different temperatures and strain rates can be obtained by equation (15). The relationship between $\ln[-\ln(1 - X)]$ and $\ln[(\varepsilon - \varepsilon_c)/\varepsilon_p]$ was linearly fitted and the slope is $m$ and the intercept is $\ln p$. The values of $m$ and $p$ are calculated to be 3.028 098 and 0.075 756, respectively. Therefore, the DRX kinetics model of AZ80A magnesium alloy can be expressed as:
The DRX kinetics curves of AZ80A magnesium alloy at different temperatures and strain rates are shown in figure 12. From figure 12, the DRX volume fraction curves exhibit typical ‘S-shape’ curves. The DRX volume fraction begins to increase very slowly in small strain. However, it dramatically increases in larger strain. Whereas, it increases slowly again in large strain and finally approaches to 100% volume fraction. When the strain rate are constants, the growth rate of the DRX volume fraction increases with the increase of deformation temperature. The evidence can be found from the slop change at various deformation temperatures. When the strain and the strain rate are all constants, the DRX volume fraction increases with the increase of deformation temperature, which indicates that the strain required to achieve the full DRX decreases with the increase of deformation temperature. When the temperature and strain are all constants, the DRX volume fractions decreases with the increase of strain rate. In sum, the decrease of strain rate and the increase of deformation temperature are beneficial for the development of DRX.

3.4. EBSD analysis
3.4.1. DRX behavior
EBSD technology is used to reveal the DRX process of AZ80A magnesium alloy during isothermal compression. Figure 13 shows EBSD microstructures at different deformation rates at the temperature of 673 K and the strain rate of 0.1 s\(^{-1}\). The green lines represent small angle misorientation, which is also considered as sub-grain boundaries, as shown in figure 13. The black lines represent large angle misorientation, which is also considered as grain boundaries. Figure 14 shows grain orientation micrographics of EBSD at different deformation rates at the temperature of 673 K and the strain rate of 0.1 s\(^{-1}\), where different colors in the figure represent different crystal orientations.

At the initial stage of the deformation, the generation and consumption of dislocations near the initial grain boundaries are competing and forming a balance. The consumption of dislocations leads to the annihilating of the grain boundaries. When the dislocation generation speed is greater than the dislocation consumption speed, dislocation accumulation and stress concentration occur at the grain boundary \[32, 33\]. In order to reduce the stress concentration, a short grain boundary is drawn at the coarse initial grain boundary to form a sawtooth grain boundary. The coarse grains have a sawtooth tendency due to the large grain boundary length \[34, 35\]. At the same time, the dislocations accumulated near the grain boundaries are rearranged to form low-angle grain boundaries and then developed into sub-grains.

Figures 13(a) and 14(a) show the EBSD microstructure at the deformation rate of 20%. It can be observed that fine DRX grains are formed at the grain boundaries. At the same time, a lot of dislocations accumulate near

\[
\begin{cases}
X_{\text{DRX}} = 0 & (\varepsilon \leq \varepsilon_c) \\
X_{\text{DRX}} = 1 - \exp\left\{ -0.075 \, 756 \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right)^{3.028 \, 098} \right\} & (\varepsilon \geq \varepsilon_c)
\end{cases}
\] (17)

The DRX kinetics curves of AZ80A magnesium alloy at different temperatures and strain rates are shown in figure 12. From figure 12, the DRX volume fraction curves exhibit typical ‘S-shape’ curves. The DRX volume fraction begins to increase very slowly in small strain. However, it dramatically increases in larger strain. Whereas, it increases slowly again in large strain and finally approaches to 100% volume fraction. When the strain rate are constants, the growth rate of the DRX volume fraction increases with the increase of deformation temperature. The evidence can be found from the slop change at various deformation temperatures. When the strain and the strain rate are all constants, the DRX volume fraction increases with the increase of deformation temperature, which indicates that the strain required to achieve the full DRX decreases with the increase of deformation temperature. When the temperature and strain are all constants, the DRX volume fractions decreases with the increase of strain rate. In sum, the decrease of strain rate and the increase of deformation temperature are beneficial for the development of DRX.
the grain boundaries, as shown in figure 13(a). Sub-grains are observed inside the deformed grains, and the orientations of the sub-grains are not similar to the orientation of the parent grains, as shown in figure 14(a), indicating that the lattice is deflected during the formation of the sub-grains. The DRX grains are first formed at the trigeminal node of the grain boundaries, as shown in figure 14(a). Due to that the stress at the trigeminal node of the grain boundary is the largest. The deformation is mainly based on the \(\{0001\}\) base slip and \(\{10\bar{1}0\}\) cylinder slip occurs in a low frequency, as shown in figure 14(a). Figures 13(b) and 14(b) show the EBSD microstructure at the deformation rate of 40%. As the deformation increases, huge amount of dislocations accumulates in the sub-grains, as shown in figure 13(b). The slip and climb of these dislocations and the continued deflection of the sub-grain lattice contribute to the formation of the DRX grains. As shown in figure 14(b), the deformation is mainly dominated by the \(\{10\bar{1}0\}\) cylinder slip. Figure 14(c) shows the EBSD microstructure at the deformation rate of 60%. It can be observed that the DRX grains are greatly developed and the microstructure is relatively uniform, which reveals that the DRX process is completely developed.

In general, continuous DRX is a process in which a grain boundary transforms from a low angle to a large angle. Figure 15 shows the orientation angle distribution of the sample at different deformation rates. Low-angle grain boundaries are generally defined as the boundaries with misorientation angle of less than 15°, and the rest angles are defined as large-angle grain boundaries. When the deformation rate is 20%, as shown in figure 15(a), there are relatively more orientation angles of the samples from 0° to 15°, and continuous DRX does not occur in this part. When the deformation rate increases to 40%, as shown in figure 15(b), the fraction of low-angle grain boundary is less than the fraction of the large-angle grain boundary when the deformation is 20%, which indicates that the low-angle grain boundaries gradually transform into the large-angle grain boundaries. At this time, some grains have undergone continuous DRX, and the DRX volume fraction rapidly increases, which indicates that the sub-grains gradually transform into DRX grains. When the deformation rate increases to 60%, as shown in figure 15(c), the low-angle grain boundaries disappear completely, and the misorientation angle is mainly distributed between 50° and 90°, which indicates that the sample have basically undergone complete DRX.
3.4.2. Texture analysis

Magnesium alloys develop texture during high-temperature deformation. Since the lattice of the magnesium alloy is a close-packed hexagonal structure, the slip system that can be activated during the deformation process is fewer than those of cubic lattice. However, the formation of the texture has a significant influence on the mechanical properties of the alloy. Therefore, it is of great significance to study the change of texture during the isothermal deformation of the magnesium alloys.

Figures 16 and 17 show the (0001) pole figures and microstructures at the temperature of 673 K, the strain rate of 0.1 s$^{-1}$ and various deformation rates of (a) 20%, (b) 40%, (c) 60%.

Figure 15. The orientation angle of the temperature of 673 K and the strain rate of 0.1 s$^{-1}$ at various deformations (a) 20%, (b) 40%, (c) 60%.

Figure 16. The (0001) pole figures at the temperature of 673 K, the strain rate of 0.1 s$^{-1}$ and various deformation rates of (a) 20%, (b) 40%, (c) 60%.
20%, two different basal textures paralleled to the compression direction are observed. It is because the initiation of the slip system does not significantly change the grain orientation, however, the initiation of twins significantly change the grain orientation \cite{36}. Although figures 14(a) and 17(a) show that a certain amount of twins is observed in the shear band during the deformation, but the grain orientation is not significantly changed due to their low content. In this process, slip is the main deformation mechanism. When the deformation rate increases to 40%, as shown in figure 16(b), the texture gradually changes from a texture of the base plane that is paralleled to the compression direction to a texture that is perpendicular to the compression direction. The reason is that, as the deformation increases, the \{0001\} basal plane of the region where DRX does not occur gradually rotates to be vertical to the compression direction. In this process, twinning is the main deformation mechanism. As shown in figure 17(a), when the deformation rate is 20%, a small amount of twins is observed. When the deformation rate is 30%, as shown in figure 17(b), lots of twins are generated in the grains. When the deformation rate increases to 40%, as shown in figure 17(c), less twins are observed in the grains. When the deformation rate increases to 60%, as shown in figure 16(c), the texture tendency is weakened and no new texture is formed. This is because when the deformation rate increases to 60%, the development of DRX is substantially completed, and the directional nucleation of the DRX and the preferential growth of the nucleated DRX do not form new textures.

4. Conclusion

In this research, the isothermal deformation behavior and microstructures of AZ80A magnesium alloy was studied by the isothermal compression tests and characterization technologies, i.e., OM and EBSD. The following conclusions were obtained:

(1) Based on the linear regression analysis of the true stress-true strain curves in deformation conditions, the constitutive model, DRX critical conditions, and DRX kinetics model of AZ80A magnesium alloy are obtained.

\[
\dot{\varepsilon} = 4.9721 \times 10^{11} \sinh(0.024 \sigma) \] 

\[
\exp\left(-\frac{173}{8.314T} \right)
\]

\[
\begin{align*}
\varepsilon_p &= 0.13263 (Z/A)^{0.15219} \\
\varepsilon_c &= 0.2363 \varepsilon_p \\
X_{\text{DRX}} &= 0 \\
X_{\text{DRX}} &= 1 - \exp\left(-0.07556 \frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right) \quad \text{for} \quad \varepsilon \leq \varepsilon_c \\
X_{\text{DRX}} &= 1 - \exp\left(-0.028098 \frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right) \quad \text{for} \quad \varepsilon \geq \varepsilon_c
\end{align*}
\]

(2) The increase of deformation temperatures and the decrease of strain rates are conducive to the occurrence and development of DRX. As the deformation rate increases, the low-angle grain boundaries (misorientation angle of less than 15° in deformation rate of 20%) gradually transformed into large-angle grain boundaries (between 50° and 90° in deformation rate of 60%), and the DRX of AZ80A alloy is gradually and more sufficiently developed.

(3) At the beginning of the deformation, the deformation is mainly based on the \{0001\} base slip, and \{10\overline{1}0\} cylinder slip occurs in a low frequency and these two different textures are paralleled to the compression
direction. In this process, slip is the main deformation mechanism. With the increase of deformation, the deformation is mainly dominated by [1010] cylinder slip. The texture is gradually transformed from a base texture that is paralleled to the direction of compression to a texture that is perpendicular to the compression direction. In this process, twinning is the main deformation mechanism.

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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).

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