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Hot tensile deformation behavior and microstructure evolution of Mg-1Al-6Y alloy

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Abstract

In this study, the hot tensile test was carried out using the extruded and annealed Mg-1Al-6Y alloy. The effect of temperature and strain rate on the hot tensile deformation behavior of the alloy was systematically studied at different temperatures (200 °C ~ 350 °C) and different strain rates (8 × 10⁻³ s⁻¹ ~ 1.6 × 10⁻³ s⁻¹). In addition, the effect of temperature on the evolution of microstructure when the strain rate is 1.6 × 10⁻³ s⁻¹ was investigated. The results showed that as the temperature increased or the strain rate decreased, the peak stress decreased and the elongation increased. Hot tensile at different temperatures all increased the texture intensity, and the microstructure after deformation showed obvious characteristics of basal fiber texture ([0001]⊥ED). Correspondingly, the weaker [−15−40]//ED texture before deformation transformed into a stronger [01−10]//ED fiber texture. After deformation, the average Schmid factor (SF) of each non-basal slip was significantly increased compared with the average SF before deformation, indicating that abundant non-basal slip was activated during the deformation. When the deformation temperature was 300 °C, dynamic recrystallization (DRX) occurred significantly, and the DRXed grains accounted for 15.9%. DRX was a combination of continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX). Furthermore, the calculated activation energy of the alloy was about 98.8 kJ mol⁻¹. Comprehensive research showed that the hot tensile deformation mechanism mainly included intragranular slip, grain boundary slip (GBS) and DRX.

1. Introduction

Magnesium (Mg) alloys are widely used in aerospace, transportation, automation, 3 C products and other fields because of their low density, high specific strength and specific stiffness, good damping performance and easy recycling [1–3]. However, as a metal with hexagonal close-packed (HCP) structure, only a limited slip system can be activated at room temperature, which leads to poor plastic deformation ability and seriously hinders the wider application of Mg alloys [4]. Researches have shown that when Mg alloys are stretched at elevated temperature, they can exhibit excellent plasticity and even superplasticity [5–7]. At present, the Mg alloys that have achieved large-scale commercial applications are mainly Mg–Al alloys, but the melting point of the β-Mg17Al12 phase in the Mg–Al alloys is only 437 °C [8]. When the service temperature exceeds 125 °C, the strength of Mg–Al alloys will be greatly reduced, which extremely limits the application of Mg–Al alloys [3, 9, 10]. It has been found that adding rare earth elements (Gd, Nd, Er, Y, Sc, etc.) to Mg–Al alloys can not only produce strong solution strengthening effect, but also form Al–RE intermetallic compounds with high melting point, which could effectively improve the elevated temperature mechanical properties of Mg–Al alloys [11, 12].

It is well known that the elevated temperature deformation mechanisms of Mg alloys mainly include intragranular slip, grain boundary sliding (GBS) and dynamic recrystallization (DRX) [13–16]. Research indicates that when the deformation temperature is greater than 200 °C, the critical resolved shear stress (CRSS) of non-basal slip systems (including prismatic (a) slip, pyramidal (a) slip and pyramidal (c+a) slip) will be
GBS provides a possible channel for Mg alloys to achieve excellent plasticity and even superplasticity [18]. However, GBS is often based on fine grain size, and the grain size of Mg alloys usually needs to be less than 10 μm [19]. At present, alloying and severe plastic deformation (SPD) are the most effective means to refine the grain of Mg alloys [20]. For instance, Figueiredo et al. [21] prepared AZ31 alloy with an average grain size of 6.0 μm through equal channel angular pressing (ECAP), when the strain rate is $10^{-4}$ s$^{-1}$ or below and the temperature range is 623–723 K, the elongation of AZ31 alloy is as high as 1000%. The results are in good agreement with the theoretical model for superplasticity based on GBS by calculating the strain rate sensitivity coefficient (m) and activation energy. In addition, Miyahara et al. [22] developed AZ61 alloy with an elongation of 1320% at a strain rate of $3.3 \times 10^{-4}$ s$^{-1}$ at 473 K through ECAP. R. Lapovok et al. [23] also prepared ZK60 alloy with an elongation of 2040% at a strain rate $3 \times 10^{-4}$ s$^{-1}$ at 220 °C through ECAP. These researches all show that GBS plays an important role in promoting the plastic deformation of the Mg alloys at elevated temperature.

DRX is one of the important research topics during the hot deformation of Mg alloys. Numerous studies have shown that DRX can refine grains, weaken texture, promote GBS and facilitate dynamic recovery (DRV) during hot deformation [24–27]. DRX mechanism contains two different modes: continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX) [28]. CDRX is mainly related to the movement of dislocations. During hot deformation, the accumulation of dislocations at grain boundaries will induce the formation of sub grain boundaries and substructures. Due to the small size of substructures, they can rotate by continuously absorbing dislocations at the sub grain boundaries in the subsequent deformation. With the rotation of the substructures, these low angle grain boundaries (LAGBs) gradually transform into high angle grain boundaries (HAGBs), and new DRXed grains are formed as a result [24]. Differently, DDRX involves two processes of nucleation and growth [29]. Generally, the second phase particles, deformation twins and grain boundary protrusions can all play the role of DDRX nucleation sites [30–32]. However, most of the existing researches on DRX mechanism are about hot compression deformation or hot extrusion deformation [32, 33], while the research on DRX mechanism during hot tensile deformation is extremely limited [34].

Hence, in this study, the extruded and annealed Mg-1Al-6Y alloy was used to study the elevated temperature mechanical properties and the effect of temperature and strain rate on the hot tensile deformation behavior. Meanwhile, the effect of temperature on the evolution of the deformed microstructure of the alloy was also investigated, and the DRX mechanism of the alloy during the hot tensile deformation was analyzed as well.

2. Experimental procedure

Mg-1Al-6Y alloy ingots were prepared by melting and casting using pure magnesium (99.99%, wt%), Mg-30Y (wt%) master alloy, and pure aluminum (99.99%, wt%). During the melting, argon and sulfur hexafluoride (99:1) were introduced to protect the Mg alloy melt. After the Mg alloy melt was completely melted, the impurities on the melt surface were removed, the temperature was raised to 740 °C and kept for 20 min, and then the melt was casted into the mold preheated at 200 °C. Subsequently, the ingot was turned into a cylinder with a size of 644 mm × 40 mm. The extrusion temperature is 500 °C, and the strain rate is 131:1 with an extrusion speed of 1 mm s$^{-1}$. During extrusion, the graphite lubricant was first coated on the inside of the mold to reduce friction, and then the alloy and the mold were preheated in the furnace, and the temperature was raised to 500 °C and kept for 1.5 h before the extrusion. A bar with a diameter of 12 mm was obtained by extrusion and then air-cooled to room temperature. After that, the obtained bar was taken out after holding at 500 °C for 2 h, immediately water quenched to eliminate the residual stress of the extruded bar, that is, annealing treatment.

In this study, the phase of the alloy was determined by x-ray diffractometer (Bruker D8 advance). The annealed bar was processed to the sample shown in figure 1 for the hot tensile test. The hot tensile test was carried out using the universal testing machine Instron-5982 equipped with a heating device, and the loading direction was parallel to the extrusion direction (ED). The temperature of tensile test was set at 200°C, 250°C, 300°C and 350°C. The strain rate of tensile test is $8 \times 10^{-5}$ s$^{-1}$, $1.6 \times 10^{-4}$ s$^{-1}$, $8 \times 10^{-4}$ s$^{-1}$ and $1.6 \times 10^{-3}$ s$^{-1}$. During the hot tensile test, the heating rate is $10^\circ$ C s$^{-1}$, and the temperature was kept for 3 min after being heated to the target temperature to heat the whole sample evenly. After hot tensile, the sample was taken out and water quenched immediately to retain the deformed microstructure. The EBSD observation surfaces of all tensile specimens are located at the center of the section parallel to the ED near the fracture, as shown in the red box in figure 1. The EBSD specimen was first mechanically polished to 7000#, then electrolytic polishing and argon ion polishing were carried out. The parameters of argon ion polishing are 2° and 3 keV, and the polishing time is 60 min. Electron backscatter diffraction (EBSD) analysis was performed using Zeiss field emission scanning electron microscope (SEM), and Oxford HKL Channel 5 software was used to process the data.
3. Results

3.1. Microstructure of the extruded and annealed Mg-1Al-6Y alloy

Figure 2(a) is the XRD pattern of the extruded and annealed Mg-1Al-6Y alloy, indicating that the alloy contains two phases, α-Mg and Al2Y. The stress-strain curve of the alloy at room temperature is shown in figure 2(b). The ultimate tensile strength (UTS), tensile yield strength (TYS) and elongation are 202.67 MPa, 104.76 MPa and 26.25%, respectively. As shown in the EBSD map of the alloy in figure 3(a), except for individual coarse grain, most of the grains are small-sized equiaxed grains. The grain size distribution map is shown in figure 3(f), and the average grain size is 13.46 μm. The color of the grains is random, and there is no obvious preferred orientation. Figure 3(b) visually shows the 3D crystal orientation of the four grains with different sizes in figure 3(a), and also shows the random orientation of the grains. Moreover, the pole density points in the {0001} pole figure (PF) and {10−10} PF in figure 3(e) are randomly distributed, showing no obvious texture characteristics, and the maximum pole intensity is 4.25. In the inverse pole figure (IPF) of the normal direction (ND), as shown in figure 3(e), the maximum pole is near [0001], indicating that there is a c-axis//ND of most grains. In the ED IPF, the maximum pole intensity is 2.43, and the maximum pole exists near [−15−40], implying that the [−15−40] of most grains are nearly parallel to the ED. In conclusion, the c-axis of more grains in the alloy is perpendicular to ED, but the texture intensity is weaker, which is mainly due to the occurrence of DRX during hot extrusion and annealing, and the random orientation of DRXed grains weakens the extruded texture [26]. Meanwhile, it can also be seen from the distribution map of misorientation angle in figure 3(c), the average misorientation angle is 40.5°, and the HAGBs greater than 10° accounts for about 75.4%, further indicating that the proportion of DRXed grains of the alloy is larger. The distribution of DRXed grains is shown in figure 3(d), in which the blue grains represent DRXed grains and the yellow grains represent substructures. Most of the microstructures are DRXed grains with an area fraction of 88.7%.

Figure 1. Specimen dimension and schematic diagram of EBSD observation surface.

Figure 2. (a) XRD pattern of the extruded and annealed Mg-1Al-6Y alloy. (b) Stress-strain curve of the alloy at room temperature.
3.2. Hot tensile deformation behavior

Figure 4 shows the stress-strain curves of extruded and annealed Mg-1Al-6Y alloy at different temperatures and different strain rates. All stress-strain curves show typical three different stages as shown in an example specially marked in figure 4(a), which is consistent with the results of previous hot tensile studies [13, 18, 34, 35]. In the first stage of the curve, the stress rises sharply until the critical stress is reached. In this case, dislocation proliferation occurs, which leads to the rapid increase of dislocation number and density, and the curve shows obvious work hardening characteristics. In the second stage, with the continuous increase of strain, the flow stress value increases from the critical stress to the peak stress, and the work hardening rate of the alloy is significantly reduced compared to the first stage. During hot tensile, DRX will occur, and the dislocations accumulated at the grain boundaries can be used as the driving force for DRX, causing some dislocations to be consumed and causing a certain degree of softening. At this stage, both softening and work hardening occur simultaneously, and softening will offset part of the work hardening effect. As the strain continues to increase, when the softening rate is equal to the hardening rate, the flow stress reaches the peak stress. In the third stage, as the deformation continues, DRX continues to occur, and the softening effect caused by DRX always occupies a dominant position, resulting in the continuous reduction of flow stress until the alloy fractures.

Figure 5 shows the changes of elongation and peak stress at different temperatures and different strain rates. Except for the abnormality of individual results, the overall tendency of the change is consistent with the conventional law. At the same temperature, with the increase of strain rate, the elongation decreases and the peak stress increases. When the strain rate is slow, there is sufficient time for DRV and DRX during the deformation, and the stress concentration is released, which is conducive to plastic deformation and aggravates the softening effect of the alloy, resulting in lower peak stress. In addition, at the same strain rate, the strength of the alloy decreases and the elongation increases with the increase of temperature. The increase of temperature is more conducive to the occurrence of DRX, which aggravates the softening degree of the alloy as well, resulting in higher elongation and lower strength. When the hot stretching temperature is 200 °C and 250 °C, the UTS of the alloy is still high. When the temperature increases to 300 °C, the peak stress of the alloy decreases greatly, indicating that DRX occurs a lot during the deformation at 300 °C. With the further increase of temperature to 350 °C, DRX is more likely to occur, so the softening caused by DRX seriously affects the strength of the alloy.

Figures 6(a)–(c) are the EBSD maps near the fracture of the tensile specimens at different temperatures when the strain rate is $1.6 \times 10^{-3}$ s$^{-1}$. The corresponding grain size distribution map is shown in figure 6(d). These coarse grains in the IPF map are all flat along the loading direction, which is due to the deformation of the grains caused by tensile loading at elevated temperatures. Meanwhile, it is observed that there is an obvious color
gradient inside the coarse grains, indicating that a large number of slips occur in the coarse grain. In addition, when the deformation temperature is 300 °C, as shown in figure 6 (c), extensive fine equiaxed DRXed grains appear around the coarse grains, and these grains are evenly distributed at the grain boundaries of the coarse grains, showing a typical necklace structure [36]. The corresponding average grain size is 7.2 μm, and the grain size is obviously refined.

Figures 7(a)–(c) are the grain boundary maps of the tensile specimens. The green line represents the LAGBs of 2–10°, the black line represents the HAGBs of >10°, and the red line represents \{10–12\} extension twin boundaries. With the increase of deformation temperature, the content of LAGBs increases slightly and then decreases significantly, as shown in figure 7(e). During the hot tensile, dislocations move to the grain boundary and accumulate in large quantities. The entanglement and accumulation of dislocations will cause a part of the grain near the grain boundary to separate from the grain and form substructure, resulting in the increase of LAGBs. Therefore, the content of LAGBs in the tensile specimen at 200 °C and 250 °C is as high as 63.1% and 65.8% respectively. When the deformation temperature is 300 °C, the content of LAGBs is dramatically reduced to 37%. This is due to the increase of temperature promotes the occurrence of CDRX, which makes the LAGBs
gradually change into HAGBs. As shown in Figure 7(c), the LAGBs are mainly concentrated at the grain boundaries of coarse grains, while there are almost no LAGBs inside the fine DRXed grains. Correspondingly, as shown in Figure 7(d), the average misorientation angle of the tensile specimen at 300 °C is the largest, 18.1°. The increase of the misorientation angle further illustrates the occurrence of DRX. In addition, with the increase of deformation temperature, the content of {10-12} extension twin boundary in the deformed microstructure decreases significantly, which is mainly due to the increase of temperature accelerates the movement of atoms, rearranges dislocations and forms recrystallized microstructure, which inhibits the nucleation of twins.

Figure 6. EBSD maps of tensile specimens at different temperatures at a strain rate of $1.6 \times 10^{-3}$ s$^{-1}$, (a) 200 °C, (b) 250 °C, (c) 300 °C; (d) Grain size distribution map.

Figure 7. Grain boundary maps of tensile specimens at different temperatures at a strain rate of $1.6 \times 10^{-3}$ s$^{-1}$, (a) 200 °C, (b) 250 °C, (c) 300 °C; (d) Misorientation angle distribution map; (e) Statistics of various grain boundary contents.
Figure 8 is the microstructure distribution map and the statistical diagram of the proportion of each microstructure of the tensile specimen at different temperatures when the strain rate is $1.6 \times 10^{-3}$ s$^{-1}$. The blue, yellow and red regions in these maps represent DRXed grain, substructure and deformed grain, respectively. As shown in figures 8(a) and (b), when the deformation temperature is 200 °C and 250 °C, the microstructure is mainly deformed grains, accounting for 85.7% and 90.2% respectively, while there are almost no DRXed grains. When the deformation temperature increases to 300 °C, as shown in figure 8(c), the number of DRXed grains increases evidently, and the area fraction is 15.9%. Figure 9 is Kernel average misorientation (KAM) maps of tensile specimens. The KAM value reflects the dislocation density in the deformed alloy to a certain extent. The larger the KAM value, the greater the dislocation density. As shown in figures 9(a) and (b), when the deformation temperature is 200 °C and 250 °C, the average KAM value is 1.47 and 1.66 respectively, and considerable number of green lines gather in the grain, especially the green lines with a darker color and higher density in the grain in figure 9(b), indicating that extensive slip deformation has occurred in the grain. As shown in figure 9(c), when the deformation temperature is 300 °C, the density of green lines in the microstructure decreases significantly and is mainly distributed at the grain boundary of coarse grains. The corresponding average KAM value is 1.12, indicating that the dislocation density is dramatically reduced.

Figure 10 shows the PFs and IPFs of the tensile specimens at different temperatures when the strain rate is $1.6 \times 10^{-3}$ s$^{-1}$. It can be found that the texture intensity of the alloys after hot tensile at different temperatures is all higher than that of the initial state alloy (figure 3(e)). In the {0001} PFs of all deformed alloys, the pole density points are concentrated along the TD. In the {10–10} PFs and {11–20} PFs, the pole density points are concentrated on both sides of the ED, showing the characteristics of typical basal fiber texture ([0001] ⊥ ED). However, the texture intensity of the alloy before deformation is weak, and there is no obvious texture, indicating that when the alloy is subjected to tensile stress parallel to ED, the grains rotate and form a clear basal fiber texture. In addition, after hot tensile, the weaker [−15–40]//ED texture before stretching is transformed into a stronger fiber texture ([01–10]//ED), as shown in figure 10. Although the random orientation of DRXed grains can weaken the texture to some extent [14, 36], the recrystallization proportion of the alloy is small, and the DRXed grains nucleated during hot tensile are small, which is prone to GBS in the subsequent deformation to form a certain preferred orientation. Researches have shown that the coarse grains are not conducive to the coordination of the deformation between the grains, which makes it difficult for the GBS deformation to occur. Meanwhile, extensive non-basal slip will be activated in the grains, resulting in a certain preferred orientation of coarse grains [34]. In this study, when the strain rate is $1.6 \times 10^{-3}$ s$^{-1}$ and the deformation temperature is 300 °C, the engineering strain of the alloy is 101%, which is much greater than 48% at 250 °C and 44% at 200 °C.
Therefore, the alloy undertakes greater plastic deformation, there are more dislocation movements in the coarse grains, and more than 80% of the microstructure in the deformed alloy is coarse deformed grains and substructures (figure 8(c)), resulting in a stronger texture with a maximum intensity of 6.42 mrd.

4. Discussion

4.1. DRX mechanism

As shown in figure 11(a), it is the R1 area marked by the black dashed ellipse in figure 6(c). Combined with the corresponding grain boundary map in figure 11(b), it can be seen that many smaller substructures s2–s10 are distributed around the coarse parent grain p1. There are green LAGBs between these substructures and the adjacent matrix. In figure 9(c), it can also be seen that these LAGBs have high KAM values, representing large dislocation density, which shows that there is massive dislocation accumulation. In addition, eighteen DRXed grains (d1–d18) with different sizes are also marked near parent grain p1 as shown in figure 11(b). There are almost no LAGBs in these DRXed grains, and the high dislocation density is not reflected in the corresponding KAM map (figure 9(c)), indicating that there is no extensive slip activation in these DRXed grains. In conclusion, the microstructure in this region shows an obvious characteristic of CDRX. In figure 11(e), the 3D crystal orientations of eight DRXed grains that are significantly different from the orientation of parent grain p1 are marked in the {0001} PF, indicating that DRX can weaken the texture to a certain extent [38]. However, other DRXed grains and all substructures s2–s10 have an orientation similar to parent grain p1, which further illustrates the CDRX process from substructures to DRXed grains. The misorientation angle between the two points A and B in figure 11(a) presents different degrees of fluctuation as shown in figure 11(c). However, the fluctuations are all within 10°, and as the distance from point A increases, the cumulative misorientation angle increases to 21° at the highest, indicating that there is considerable slip in the parent grain p1. Accordingly, the three-dimensional hexagons at five different positions between points A and B also indicate that prismatic slip was activated [14]. Therefore, the [01–10]//ED texture formed after hot tensile in figure 10 is closely related to the activation of prismatic slip in deformed alloys. For comparison, figure 11(d) shows the misorientation angle along the arrow CD in figure 11(a). The misorientation angle along the arrow CD hardly fluctuates within 10°, indicating that there is no significant dislocation movement in these DRXed grains.

It is well known that DDRX includes explicit nucleation and growth processes [29]. Under elevated temperature deformation conditions, the nucleation of DDRX usually occurs at the jagged grain boundary of the parent grain and the triple junctions of the parent crystal [24, 39]. Subsequently, further growth of DRXed grains is achieved through grain boundary migration under continuous strain [29]. As an important feature of the DDRX mechanism, grain boundary bulging is mainly due to the difference of dislocation density at grain boundary caused by uneven strain during deformation [31]. The dislocations with different densities at the grain boundaries continue to slip, resulting in different degrees of grain boundary migration, and ultimately leading to...
local grain boundaries protruding toward adjacent grains. Then, under the action of other slip systems, LAGBs is generated between the protruding part of the parent crystal and the parent crystal, and substructure composed of the bulged original HAGBs and the new LAGBs is formed. In the subsequent deformation, these LAGBs could

Figure 10. PFs and IPFs of tensile specimens at different temperatures at a strain rate of $1.6 \times 10^{-3} \text{s}^{-1}$, (a) (b) 200°C, (c) (d) 250°C, (e) (f) 300°C.

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transform into HAGBs through substructure rotation and grain boundary migration, eventually changing substructure into new DRXed grains.

The EBSD maps shown in figures 12(a) and (e) are R2 and R3 areas marked by black dashed ellipses in figure 6(c), respectively. In the R2 region, as shown in figure 12(a), numerous fine DRXed grains marked by black arrows, these DRXed grains have completely different orientations from the surrounding grains and are located at the triangle confluences of parent grains. In the R3 region, as shown in figure 12(e), fine DRXed grains (d1–d8) are formed at the serrated grain boundary of the parent crystal p1 due to stress concentration and uneven deformation. These fine DRXed grains in the R2 and R3 regions show typical DDRX characteristics [40, 41]. More specifically, in the R3 region, the grains s9 and s10 are all composed of the bulging initial grain boundaries and the newly formed LAGBs (the white lines in the EBSD maps represent the LAGBs), that is, the substructure, and their orientation is approximately consistent with the parent grain p1. Grains s9 and s10 are in the nucleation stage of DRX, which provides evidence for the DDRX mechanism of grain boundary bulging nucleation [36]. Furthermore, in the R3 region, the grains d6 and d7 are completely composed of HAGBs, and the crystal orientation is significantly different from the parent grain p1, indicating that they have completely become DRXed grains. For the grains d7 and s9 in the R3 region, figure 12(h) shows the misorientation angle distribution along the arrows AB and CD passing through the interior of the grain, further indicating that there is LAGBs between the grain s9 and the parent crystal, while HAGBs exists between the grain d7 and the parent crystal. And there is almost no change in the misorientation angle inside the grains. As shown in figure 12(d), similar results can also be seen in the misorientation angle distribution along the arrow EF in figure 12(a), indicating that there is no obvious dislocation movement in these fine DRXed grains. However, some LAGBs exist in the large DRXed grains d4, d8, d11 and d21, indicating that the grown DRXed grains can continue to be refined through slip deformation [24]. In addition, figures 12(b) and (f) are the {0001} PFs of the R2 region and the R3 region, respectively; and figures 12(c) and (g) are the ED IPFs of the R2 region and the R3 region, respectively. It can be seen that the DRXed grains are randomly distributed in the {0001} PFs and the ED IPFs, indicating that DRX has a certain weakening effect on the texture.
4.2. Deformation mechanism

Figure 13 is the histogram of the Schmid factor (SF) distribution before and after hot tensile at different temperatures (including basal \(\langle a \rangle\) slip, prismatic \(\langle a \rangle\) slip, pyramidal \(\langle a \rangle\) slip and pyramidal \(\langle c + a \rangle\) slip). The inset is the SF map of the grains with the SF greater than 0.35 in the corresponding slip system. The color changes from blue to red, indicating that the SF increases from 0.35 to 0.5. For the alloy before hot tensile, the proportion of grains with basal \(\langle a \rangle\) slip SF greater than 0.35 is only 42.5%, which is much smaller than that of grains with prismatic \(\langle a \rangle\) slip, pyramidal \(\langle a \rangle\) slip and pyramidal \(\langle c + a \rangle\) slip SF greater than 0.35, which are 65.2%, 88.7% and 72.9% respectively. The average SF of basal \(\langle a \rangle\) slip is 0.287, which is also much smaller than that of prismatic \(\langle a \rangle\) slip of 0.375, pyramidal \(\langle a \rangle\) slip of 0.425 and pyramidal \(\langle c + a \rangle\) slip of 0.398. It is well known that under elevated temperature deformation conditions, the CRSS of non-basal slip will be substantially reduced [17], and it is easy to be activated. After hot tensile at 200 °C, the proportion of grains with basal \(\langle a \rangle\) slip SF greater than 0.35 is slightly increased. However, when the deformation temperature exceeds 200 °C, the proportion of grains with basal \(\langle a \rangle\) slip SF greater than 0.35 is significantly reduced. Moreover, the grains with large basal \(\langle a \rangle\) slip SF are mainly smaller grains, as shown in figures 13 (b1), (c1) and (d1). Basal \(\langle a \rangle\) slip has the smallest CRSS and is the easiest to operate [17, 42]. In the initial deformation stage, the activation of a large number of basal \(\langle a \rangle\) slip changes the orientation of relevant grains and their surrounding grains to an orientation more conducive to the operation of other non-basal slip systems [34]. These small grains with high basal \(\langle a \rangle\) slip SF are DRXed grains or substructures, and most of them have different orientations from the coarse deformed grains. In addition, it is worth noting that after hot tensile at different temperatures, the proportion of grains with non-basal slip SF greater than 0.35 has a remarkable increase compared with the alloy before deformation, and the average SF of non-basal slip also increases correspondingly, indicating that abundant non-basal slip is activated during the hot tensile.

In general, the strain rate sensitivity coefficient \((m)\) can reflect the plastic deformation ability of metal materials [43]. When the deformation temperature is constant, the value of \(m\) is the slope of the relation \(\ln \sigma - \ln \dot{\varepsilon}\) through linear fitting, and the specific definition is as follows [44]:

\[
m = \left. \frac{\partial \ln \sigma}{\partial \ln \dot{\varepsilon}} \right|_{T, \dot{\varepsilon}}
\]

In the equation, \(\sigma\) represents the peak stress, and \(\dot{\varepsilon}\) represents the strain rate. Through equation (1), according to the tensile test results of the alloy at different temperatures and strain rates, the results obtained by calculation and fitting are shown in figure 14. The value of \(m\) gradually increases from 0.005 to 0.13 as the deformation temperature increases from 200 °C to 350 °C. Generally, the value of \(m\) reflects the transfer and diffusion
Figure 13. Schmid factor distribution histogram, (a1)–(a4) initial alloy, (b1)–(b4) 200 °C, (c1)–(c4) 250 °C, (d1)–(d4) 300 °C; (a1)–(d1) Basal (a), (a2)–(d2) Prismatic (a), (a3)–(d3) Pyramidal (a), (a4)–(d4) Pyramidal (c + a); the inset is the Schmid factor map of the corresponding slip system with SFs greater than 0.35.

Figure 14. Strain rate sensitivity coefficient (m) calculated by fitting $\ln \sigma - \ln \dot{\varepsilon}$. 
capacity of local necking [44, 45]. It is universally believed that the greater the value of m, the stronger the inhibition of necking, and the higher the elongation of the alloy [46]. Therefore, in this study, the alloy has the largest m value when the deformation temperature is 350 °C and exhibits good plastic deformation ability.

To further study the hot tensile deformation behavior of the alloy, the Arrhenius model was used to investigate the relationship between the deformation factors and the flow stress. The specific expression is as follows [47]:

$$\dot{\varepsilon} = A \sigma^n \exp \left(-\frac{Q}{RT}\right)$$  \hspace{1cm} (2)

where $\sigma$ and $\dot{\varepsilon}$ are the peak stress and strain rate, respectively; A and n are material constant ($n = 1/m$, m is strain rate sensitivity coefficient); R and Q are molar gas constant (8.314 J mol$^{-1}$ K$^{-1}$) and the activation energy, respectively; T is deformation temperature. Thus, equation (2) is transformed into the following equation:

$$\dot{\varepsilon} \cdot \exp \left(\frac{Q}{RT}\right) = A \sigma^n$$  \hspace{1cm} (3)

To find the value of activation energy Q, take the natural logarithm on both sides of equation (3), which can be obtained as:

$$\ln \sigma = \frac{\ln \dot{\varepsilon} - \ln A}{n} + \frac{1}{T} \cdot \frac{Q}{R \cdot n}$$  \hspace{1cm} (4)

Equation (4) can be further expressed as:

$$Q = R \cdot n \cdot \left|\frac{\partial \ln \sigma}{\partial(1/T)}\right| = R \cdot \frac{1}{m} \cdot \left|\frac{\partial \ln \sigma}{\partial(1/T)}\right|$$  \hspace{1cm} (5)

As shown in figure 15, by linearly fitting the $\ln \sigma - 1/T$ relation, the corresponding slope can be obtained, and then the average slope $Q/(R \cdot n)$ can be obtained. Finally, after calculation, the corresponding activation energy Q value of the alloy is about 98.8 kJ mol$^{-1}$, which is higher than the grain boundary diffusion activation energy of pure Mg (92 kJ mol$^{-1}$) [48]. Hence, in this study, the elevated temperature deformation mechanism of the alloy also includes GBS.

4.3. Fracture morphology and elevated temperature mechanical properties

Figure 16 shows the fracture morphology of the tensile specimens at different temperatures when the strain rate is $1.6 \times 10^{-3}$ s$^{-1}$. In the SEM micrographs of the fracture of the tensile specimens at 200 °C and 250 °C, as shown in figures 16 (a) and (b), a large number of dimples, cleavage steps and tearing edges co-exist in the fracture, and the fracture shows the characteristics of ductile-brittle mixed fracture. With the increase of deformation temperature to 300 °C and 350 °C, as shown in figures 16 (c) and (d), the number of dimples in the
fracture gradually increases, the diameter of dimples gradually decreases, the color of dimples gradually deepens, showing a typical characteristic of ductile fracture.

When the strain rate is $1.6 \times 10^{-3} \text{s}^{-1}$ at 200°C, the UTS of the alloy is 190.21 MPa. Compared with the UTS of 202.67 MPa at room temperature, the UTS is reduced by about 6.15%, indicating that the alloy has excellent elevated temperature mechanical properties. During hot tensile at 200°C, due to the low temperature, the atomic diffusion is slow, the nucleation of DRX is difficult, and DRX hardly occurs (figure 8(a)). In the second stage of deformation, the softening effect caused by DRX is weak, and it takes longer for the alloy to enter the third stage where the softening rate is greater than the hardening rate, so the alloy shows higher peak stress, as shown in figure 4. In addition, the high melting point Al₂Y phase dispersed at grain boundaries and in the grains [49] has excellent thermostability and is not easy to decompose [50, 51]. During the hot tensile deformation, the Al₂Y phase can inhibit the diffusion of grain boundary, make the microstructure near grain boundary relatively stable, and effectively hinder GBS. And Al₂Y phase has a good orientation relationship with Mg matrix: (0002)_{Mg}//[2–22]Al₂Y; [2–1–10]_{Mg}//[−2–11] Al₂Y and [01–10]_{Mg}//[011] Al₂Y [52]. Therefore, the Al₂Y phase can still play an excellent second phase strengthening role during hot tensile.

5. Conclusion

(1) As the temperature increases or the strain rate decreases, the peak stress of the extruded and annealed Mg-1Al-6Y alloy decreases and the elongation increases. Under the condition of hot tensile at 200°C, the UTS of the alloy is reduced by about 6.15%, and the alloy has excellent elevated temperature mechanical properties.

(2) When the deformation temperature is 300°C, DRX occurs significantly, and the proportion of DRXed grains in the microstructure after hot tensile is 15.9%. DRX is mainly composed of CDRX and DDRX.

(3) The hot tensile at different temperatures all increase the intensity of the texture, and the microstructure after deformation shows obvious characteristics of the basal fiber texture ((0001)⊥ED), and the weaker [−15–40]/ED texture before deformation transform into a stronger [01–10]/ED fiber texture.

(4) The activation energy of the extruded and annealed Mg-1Al-6Y alloy is about 98.8 kJ mol⁻¹. The hot tensile deformation mechanism is the synergistic effect of intragranular slip, GBS and DRX.
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Data availability statement

The data generated and/or analysed during the current study are not publicly available for legal/ethical reasons but are available from the corresponding author on reasonable request.

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