Coupled Influence of Temperature and Strain Rate on Tensile Deformation Characteristics of Hot-Extruded AZ31 Magnesium Alloy

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Abstract To explore the coupled effect of temperature T and strain rate $\dot{\epsilon}$ on the deformation features of AZ31 Mg alloy, mechanical behaviors and microstructural evolutions as well as surface deformation and damage features were systematically examined under uniaxial tension at T spanning from 298 to 523 K and $\dot{\epsilon}$ from 10^{-4} to 10^{-2} s^{-1} . The increase in T or the decrease in $\dot{\epsilon}$ leads to the marked decrease in flow stress, the appearance of a stress quasi-plateau after an initially rapid strain hardening, and even to the occurrence of successive strain softening. Correspondingly, the plastic deformation modes of AZ31 Mg alloy transform from the predominant twinning and a limited amount of dislocation slip into the enhanced non-basal slip and the dynamic recrystallization (DRX) together with the weakened twinning. Meanwhile, the cracking modes also change from along grain boundaries (GBs) and at twin boundaries (TBs) or the end of twins into nearby GBs where the DRX has occurred. The appearance of a stress quasi-plateau, the formation of large-sized cracks nearby GBs, and the occurrence of continuous strain softening, are intimately related to the enhancement of the non-basal slip and the DRX.

KEY WORDS: AZ31 Mg alloy; Uniaxial tension; Temperature; Strain rate; Deformation; Damage; Twinning

1 Introduction

Magnesium and its alloys have great potential for application in automotive and aerospace industries as a substitute of aluminum and other metallic materials due to the

lowest density, high specific strength and stiffness, good castability and recyclability, easy availability and fine cutting performance [1]. Nowadays, most components of commercial magnesium alloys are still prepared by casting, but the wrought magnesium alloys are of considerable interest because they provide higher strength, better ductility and excellent comprehensive mechanical properties [2–4] compared to cast alloys. However, magnesium alloys with hexagonal close packed (HCP) crystal structure always exhibit poor formability at room temperature due to its limited slip systems, which becomes one of the main reasons for hindering their industrial applications. Therefore, a number of investigations on the mechanical behaviors of magnesium alloys at different T and $\dot{\epsilon}$ have been widely conducted [5–13], and many investigated results indicated that the workability of magnesium alloys is intimately related to their microstructures, which strongly depend on the applied T and $\dot{\epsilon}$ conditions.

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According to T , the deformation behavior of magnesium alloys can be divided into three distinct regimes, i.e., low temperature zone (below 473 K), moderate temperature zone (473–573 K) and high temperature zone (623–673 K) [8, 10]. At low temperature zone, the plastic deformation of polycrystalline magnesium is mainly limited to the basal slip and twinning [8, 13]. With increasing T , non-basal slip is activated and the DRX also occurs nearby GBs or TBs [8, 13–16]. As T enters into the high temperature regime, non-basal slip occurs uniformly and the DRX becomes more homogeneous; in addition, other diffusion-assisted mechanisms, such as GB sliding, solute drag and climb-controlled dislocation creep, may become relevant [8, 15–18]. Current commercial forming often requires the elevated temperatures, e.g., 673 K or higher. However, hot forming at over 473 K has some disadvantages including complex high temperature forming tool systems, poor surface quality, high cost, etc. As reported in some literature [19, 20], the forming processes at below 473 K are viable for manufacturing complex magnesium components; therefore, the investigations on the deformation behavior of magnesium alloys at T below 473 K are very important.

However, $\dot{\epsilon}$ has also a notable effect on the deformation behavior of magnesium alloys at low and moderate T zones [5–7, 9–11, 21]. Till now, more attentions are focused on the mechanical behaviors and microstructures, whereas the variations of deformation and damage features with T and $\dot{\epsilon}$ are rarely concerned with **the literature**. To examine deeply the coupled effect of T and $\dot{\epsilon}$ on the deformation features of HCP materials, in the present work, AZ31 Mg alloy was chosen as the experimental material, and the mechanical behavior as well as the surface deformation and damage features and microstructural evolutions at various combinations of T and $\dot{\epsilon}$ were systematically investigated under uniaxial tensile tests.

2 Experimental

The AZ31 Mg alloy ingot with a diameter of 150 mm was chosen as the original material, the chemical composition is (wt%): Al 3.2, Zn 0.86, Mn 0.31, and balanced of Mg. The ingot was then extruded at 673 K into a 7 mm thick sheet. Optical microscopic (OM) images of the microstructure of the as-extruded alloy sheet are given in Fig. 1. It is found that the microstructure consists of the equiaxed α -Mg with an average grain size of $\sim 20 \mu\text{m}$, some $\text{Mg}_{17}\text{Al}_{12}$ precipitates distributed mainly along GBs and a small amount of twins.

The total dimensions of tensile specimens cut from the as-hot-extruded sheet are 10 mm $\times$ 3 mm $\times$ 45 mm with a gauge section of 5 mm $\times$ 3 mm $\times$ 23 mm, and the length direction of specimens is parallel to the extruded

Fig. 1 OM image of microstructure of as-extruded AZ31 Mg alloy

direction of sheet. Before tensile tests, the specimens were polished electrolytically to obtain a strain-free and clean surface for microscopic observations. The tensile tests were conducted to a final fracture using a SANS CMT 5105 testing machine with the initial $\dot{\epsilon}$ of 10^{-4} , 10^{-3} and 10^{-2} s^{-1} , respectively, at T ranging from 298 to 523 K. The deformation and damage features on the specimen surfaces and the corresponding microstructures after tensile rupture were examined carefully by using SSX-55 scanning electron microscope (SEM), Imager A2 m optical microscope and Tecnai G² 20 transmission electron microscope (TEM), respectively.

3 Results

3.1 Stress–Strain Behavior

Figure 2 gives the true stress–strain curves of AZ31 Mg alloy at various T , respectively, at 10^{-4} , 10^{-3} and 10^{-2} s^{-1} . With increasing T at all three strain rate conditions, the flow stress largely decreases, meanwhile, the strain hardening degree becomes weakened after an initially rapid strain hardening and a notable stress quasi-plateau appears at $T \geq 423 \text{ K}$, especially at a low strain rate of 10^{-4} s^{-1} . However, as T is as high as 523 K and $\dot{\epsilon}$ is 10^{-4} s^{-1} , a continuous strain softening stage occurs after the rapid strain hardening stage.

In order to clearly reveal the influence of $\dot{\epsilon}$ on the temperature-dependent mechanical behavior of AZ31 Mg alloy, the true stress–strain curves at various $\dot{\epsilon}$ and under a constant T are given in Fig. 3. At 298 K, the flow stress almost has no change at three $\dot{\epsilon}$, whereas the flow stress at 373 K and two higher $\dot{\epsilon}$ is basically comparable, but lower than that at 298 K, and the obvious decrease in flow stress occurs at 10^{-4} s^{-1} (Fig. 3a). As T is above 373 K, with decreasing $\dot{\epsilon}$, the

Fig. 2 Tensile true stress–strain curves of AZ31 Mg alloy at various temperatures, respectively, at the strain rates of **a** 10^{-4} s^{-1} , **b** 10^{-3} s^{-1} , **c** 10^{-2} s^{-1}

Fig. 3 Tensile true stress–strain curves of AZ31 Mg alloy at different strain rates and different temperatures: **a** 298 K and 373 K; **b** 423 K; **c** 473 K; **d** 523 K

flow stress notably decreases, and the prominent stress quasi-plateau is observed (Fig. 3b–d), especially at a low strain rate of 10^{-4} s^{-1} , with an exception at 523 K and 10^{-4} s^{-1} , which indicates that the strain softening degree increases with lowering $\dot{\epsilon}$ at T above 373 K.

The variation of ultimate tensile strength σ_{UTS} and yield strength σ_{YS} with T at different $\dot{\epsilon}$ is shown in Fig. 4. With

increasing T , the σ_{UTS} and σ_{YS} all continuously decrease at the three strain rate conditions, but the decreasing amplitude is comparatively greater at low $\dot{\epsilon}$. At a certain T , an increase in $\dot{\epsilon}$ raises strength, and the amount of such an effect becomes much greater at increasing temperatures.

In summary, with raising T or decreasing $\dot{\epsilon}$, the flow stress, σ_{UTS} and σ_{YS} of AZ31 Mg alloy all decrease;

Fig. 4 Variations of ultimate tensile strength σ_{UTS} and yield strength σ_{YS} of AZ31 Mg alloy with temperature at different strain rates

meanwhile, a prominent stress quasi-plateau tends to appear after an initially rapid strain hardening, and continuous strain softening even occurs.

3.2 Surface Deformation and Damage Features

Figures 5 and 6 typically show the SEM images of the deformation and damage characteristics on the lateral surfaces near the fracture surfaces after AZ31 Mg alloy was loaded to tensile rupture at different T and $\dot{\epsilon}$. A lot of cracks are formed, and the propagation direction of some cracks is almost normal to the tensile direction at 298 K (Fig. 5a, b). As T is 373 K, the elongated grains are formed

and the amount of cracks notably decreases at 10^{-4} s^{-1} (Fig. 5c), whereas at 10^{-2} s^{-1} , the damage features almost do not change compared to the case at 298 K (Fig. 5d). With continuously raising T , many fine recrystallized grains and cracks are observed nearby GBs at 10^{-4} s^{-1} (Fig. 5e, g); the higher the T , the stronger the recrystallization degree is. In this case, the true stress–strain curves exactly exhibit an obvious stress quasi-plateau (Fig. 3b, c), whereas at 10^{-2} s^{-1} , the critical temperatures, at which the grain elongation and the recrystallization take place, correspond to 423 K (Fig. 5f) and 473 K (Fig. 5h), respectively. Meanwhile, at 473 K, the stress quasi-plateau is also found in the true stress–strain curve (Fig. 3c).

Further observation reveals that the cracks are mainly formed along GBs or at TBs at 298 K (Fig. 6a, b), and the slip bands (SBs) are also observed. As T is 373 K, the cracks are found only along GBs, and the density of SBs increases at 10^{-4} s^{-1} compared to that at 298 K (Fig. 6c); at 10^{-2} s^{-1} , the cracks are also formed at the end of twins, besides along GBs and at TBs, and the density of SBs are still lower (Fig. 6d). With continuously raising T to 423 K, the SBs become denser and the cracks with larger size appear primarily nearby GBs at 10^{-4} s^{-1} (Fig. 6e), where apparent DRX has occurred; however, at 10^{-2} s^{-1} , the surface deformation features are basically similar to that at 373 K and 10^{-4} s^{-1} , namely, the SBs is denser and the cracks are formed only along GBs (Fig. 6f). As T is as high as 473 K, the recrystallization degree becomes enhanced and the size of cracks remarkably increases, especially at 10^{-4} s^{-1} ; in this case, the SBs become sparser compared to the case at 423 K (Fig. 6g, h).

On the whole, the twinning is predominant at 298 K for the plastic deformation of AZ31 Mg alloy. With increasing

Fig. 5 Low-magnification SEM images of the surface deformation features for AZ31 Mg alloy deformed at different temperatures and strain rates: **a** 298 K, 10^{-4} s^{-1} ; **b** 298 K, 10^{-2} s^{-1} ; **c** 373 K, 10^{-4} s^{-1} ; **d** 373 K, 10^{-2} s^{-1} ; **e** 423 K, 10^{-4} s^{-1} ; **f** 423 K, 10^{-2} s^{-1} ; **g** 473 K, 10^{-4} s^{-1} ; **h** 473 K, 10^{-2} s^{-1}

Fig. 6 High-magnification SEM images of the surface deformation features for AZ31 Mg alloy deformed at different temperatures and strain rates: **a** 298 K, 10^{-4} s^{-1} ; **b** 298 K, 10^{-2} s^{-1} ; **c** 373 K, 10^{-4} s^{-1} ; **d** 373 K, 10^{-2} s^{-1} ; **e** 423 K, 10^{-4} s^{-1} ; **f** 423 K, 10^{-2} s^{-1} ; **g** 473 K, 10^{-4} s^{-1} ; **h** 473 K, 10^{-2} s^{-1}

T or decreasing $\dot{\epsilon}$, the twinning deformation becomes weakened, whereas the dislocation slip and the DRX are enhanced. Correspondingly, the cracking modes change from along GBs or at TBs and the end of twins into nearby GBs where the DRX has occurred, and the enhancement of DRX results in the formation of cracks with larger size nearby GBs.

3.3 Microstructures

The microstructural evolutions on the lateral surfaces near fracture surfaces after AZ31 Mg alloy specimens were loaded to tensile rupture are carefully examined, as demonstrated in Figs. 7 and 8. A number of twins are observed at 298 K and three $\dot{\epsilon}$ (Fig. 7a–c). As T is 373 K, the density of twins dramatically decreases and the elongated grains are formed at 10^{-4} s^{-1} (Fig. 7d), whereas at two high $\dot{\epsilon}$, the density of twins still keeps higher (Fig. 7e, f). With increasing T to 423 K, the grain elongating takes place at all three strain rates (Fig. 7g–i), and a small amount of fine DRX grains are formed nearby GBs at 10^{-4} s^{-1} (Fig. 7g), which should be the major reason for the appearance of a stress quasi-plateau in the true stress–strain curve (Fig. 3b); at this temperature, the twins are very few in amount at two low $\dot{\epsilon}$ (Fig. 7g, h), whereas at 10^{-2} s^{-1} , the density of twins is relatively higher (Fig. 7i). With continuously increasing T and decreasing $\dot{\epsilon}$, the DRX becomes more remarkable (Fig. 8), especially at 523 K and 10^{-4} s^{-1} , original grains have been almost completely replaced by fine equiaxed recrystallized grains (Fig. 8d), which leads to the occurrence of successive strain softening (Fig. 3d).

The similar results are also observed by TEM, as shown in Fig. 9. The sizes of twins and grains are larger in the hot-extruded sheet (Fig. 9a). However, for the AZ31 Mg alloy deformed at 10^{-2} s^{-1} , the density of twins is higher and their size is smaller at 298 K (Fig. 9b); with increasing T to 423 K, the amount of twins decreases and their boundaries become indistinct (Fig. 9c). As T is as high as 523 K, more recrystallized grains are observed (Fig. 9d).

In a word, with increasing T or decreasing $\dot{\epsilon}$, the density of twins decreases and the DRX becomes enhanced.

4 Discussion

Based on the above results, the schematic images are depicted in Fig. 10 which shows clearly the major differences in the deformation and damage features and microstructures on the lateral surfaces near fracture surfaces after AZ31 Mg alloy was loaded to tensile rupture at a low (10^{-4} s^{-1}) and a high (10^{-2} s^{-1}) $\dot{\epsilon}$, respectively, at 373, 423 and 473 K. As T is 373 K and $\dot{\epsilon}$ is 10^{-4} s^{-1} , the elongated grains together with a large amount of SBs and a small amount of twins as well as the cracks along GBs are formed (Figs. 5c, 6c, 7d, 10a). However, at a high strain rate of 10^{-2} s^{-1} , a high density of twins, many cracks forming along GBs and at TBs or the end of twins, and sparser SBs are observed (Figs. 5d, 6d, 7f, 10b). Similar phenomena are also presented at 298 K. At 423 K, some recrystallized grains coupled with the cracks are induced nearby GBs at low $\dot{\epsilon}$, and the SB density is still higher, but only very few twins are formed (Figs. 5e, 6e, 7g, 10c). At high $\dot{\epsilon}$, the twin density is still higher, but obviously lower

Fig. 7 OM images of microstructures for AZ31 Mg alloy tensioned at different temperatures and strain rates: **a** 298 K, 10^{-4} s^{-1} ; **b** 298 K, 10^{-3} s^{-1} ; **c** 298 K, 10^{-2} s^{-1} ; **d** 373 K, 10^{-4} s^{-1} ; **e** 373 K, 10^{-3} s^{-1} ; **f** 373 K, 10^{-2} s^{-1} ; **g** 423 K, 10^{-4} s^{-1} ; **h** 423 K, 10^{-3} s^{-1} ; **i** 423 K, 10^{-2} s^{-1}

Fig. 8 OM images of microstructures for AZ31 Mg alloy tensioned at different temperatures and strain rates: **a** 473 K, 10^{-4} s^{-1} ; **b** 473 K, 10^{-3} s^{-1} ; **c** 473 K, 10^{-2} s^{-1} ; **d** 523 K, 10^{-4} s^{-1} ; **e** 523 K, 10^{-2} s^{-1}

Fig. 9 TEM images of microstructures for AZ31 Mg alloy at different states: **a** as-hot-extruded; **b** deformed at 298 K and 10^{-2} s^{-1} ; **c** deformed at 423 K and 10^{-2} s^{-1} ; **d** deformed at 523 K and 10^{-2} s^{-1}

Fig. 10 Schematics of the major differences in the deformation and damage features and microstructures on the lateral surfaces near fractures after AZ31 Mg alloy was loaded to tensile rupture at low (10^{-4} s^{-1}) and high (10^{-2} s^{-1}) strain rates at different temperatures: **a** 373 K, 10^{-4} s^{-1} ; **b** 373 K, 10^{-2} s^{-1} ; **c** 423 K, 10^{-4} s^{-1} ; **d** 423 K, 10^{-2} s^{-1} ; **e** 473 K, 10^{-4} s^{-1} ; **f** 473 K, 10^{-2} s^{-1}

than that at 373 K, and the cracks only along GBs are detected and the SB density decreases (Figs. 5f, 6f, 7i, 10d). As T is as high as 473 K, more recrystallized grains along with large-sized cracks appear nearby GBs at a low $\dot{\epsilon}$ (Figs. 5g, 6g, 8a, 10e), whereas at a high $\dot{\epsilon}$, the recrystallization degree becomes weakened and the SB density is higher than that at 10^{-4} s^{-1} (Figs. 5h, 6h, 8c, 10f).

According to the above experimental results, the deformation features of AZ31 Mg alloy at different $\dot{\epsilon}$, respectively, at $T < 423 \text{ K}$ and $T \geq 423 \text{ K}$ will be discussed in the following Sects. 4.1 and 4.2.

4.1 Deformation Features at Different $\dot{\epsilon}$ Under $T < 423 \text{ K}$

The plastic deformation modes in Mg and Mg alloys include basal ($\{0001\} \langle a \rangle$), prismatic ($\{10\bar{1}0\} \langle a \rangle$), and the first-

and second-order pyramidal ($\{10\bar{1}1\} \langle a \rangle$ and $\{11\bar{2}2\} \langle c + a \rangle$) slip, as well as $\{10\bar{1}2\}$ extension twinning and $\{10\bar{1}1\}$ and $\{10\bar{1}3\}$ contraction twinning [12, 14, 22]. For all these deformation modes, the critical resolved shear stresses (CRSSs) obtained from single crystal experiments at room temperature from low to high are on the order of basal slip, extension twinning, prismatic slip, pyramidal slip and contraction twinning. The five independent slip systems are generally required for the plastic deformation of polycrystalline materials. However, the basal slip with the lowest CRSS can only provide two independent slip systems in Mg alloys having less than five independent slip systems [14, 23]. Thus, the deformation twinning becomes an important deformation mechanism, further leading to the formation of a large number of twins with multiple orientations at 298 K and three $\dot{\epsilon}$. In this case, many cracks are formed at TBs, which is analogous to the phenomenon observed with CP Zn with a

high compressive strain [24]. The formation of such cracks is consequent upon the action of shear stress induced by the twin growth in the surrounding matrix. In the process of twin growth, the shear stress is produced in the matrix surrounding twins. To accommodate the shear deformation of twinning, the plastic deformation in the matrix around twins must be initiated, thus inducing the formation of cracks at TBs or the end of the twins as the matrix cannot well accommodate the twin shearing. Moreover, the intersecting of twins with GBs also aggravates the formation of cracks along GBs. Therefore, some cracks almost normal to tensile direction are observed. The higher CRSS required for deformation twinning compared to that for the basal slip, coupled with the inhibition of numerous twins to dislocation slip [25], leading to the occurrence of successive strain hardening after an initially rapid strain hardening at 298 K.

It has been recognized that the CRSSs for the basal slip and the extension twinning do not change with T , while the CRSSs for the non-basal slip decrease with raising T [22, 26] and that the CRSS ratios of basal to non-basal slip are of order 1:40–1:80 for single crystal Mg and 1:2–1:5 for polycrystalline Mg at room temperature. Consequently, it is expected that the non-basal slip is activated at 373 K and 10^{-4} s^{-1} , as shown in Figs. 6c, 7d and 10a, namely, dense SBs are formed and the density of twins dramatically decreases. Such a phenomenon was also found in uniaxially tensioned CP Ti, for which the number of deformation twins reduces with increasing T [27]. The dramatic decrease in twin density weakens the interactions between twins and dislocations and thus leads to an obvious decrease in flow stress and weakens strain hardening ability. Nevertheless, the decrease in CRSSs for the non-basal slip with increasing T is less distinct at a high $\dot{\epsilon}$ than at a low $\dot{\epsilon}$ [12]. Meanwhile, the stress concentration is readily produced nearby GBs at a high $\dot{\epsilon}$ due to the fact that the dislocation slip is restrained, thus enhancing the tendency for the formation of twins [28, 29]. Therefore, at two higher $\dot{\epsilon}$, the twin density is still higher, which leads to the higher flow stress and stronger strain hardening compared to the case at 10^{-4} s^{-1} .

4.2 Deformation Features at Various $\dot{\epsilon}$ Under $T \geq 423 \text{ K}$

As $T \geq 423 \text{ K}$, the CRSSs for the non-basal slip tend to decrease substantially [22]. Some studies have demonstrated that the activation of the non-basal slip modes can accelerate the DRX [15, 30, 31]. For example, Dudamell *et al.* [15] examined the influence of texture on the recrystallization mechanisms in an AZ31 Mg sheet alloy at dynamic rates and found that the DRX is enhanced by the operation of $\langle c + a \rangle$ slip, since the cross-slip and climb take place more readily, which stems from the following

two facts: the first is that the pyramidal slip has more independent slip systems than prismatic slip [14], and the second is that the stacking fault energy (SFE) of pyramidal dislocation is also significantly higher ($<50 \text{ mJ/m}^2$ for the basal slip, 354 mJ/m^2 for the prismatic slip, 496 and 452 mJ/m^2 for the first and second pyramidal slip [32, 33]). Kaibyshev *et al.* [30] also concluded that, once all three main slip modes (basal, prismatic and pyramidal $\langle a \rangle$) are activated, the DRX are accelerated because the activation of many slip modes promotes the cross-slip of basal dislocations into non-basal planes, and these non-basal dislocations are easier to climb and arrange into new boundaries. Del Valle *et al.* [31] also observed an enhancement of the DRX in AZ31 Mg alloy as the multiple slip occurs compared to the case only for basal slip. Besides, the decrease in $\dot{\epsilon}$ will more easily result in the occurrence of the DRX, which originates from the easiness of diffusion and the more pronounced decrease in the CRSS for the $\langle c + a \rangle$ slip with temperature [12]. For example, at 10^{-4} s^{-1} , dense SBs with multiple orientations are detected at 423 K; meanwhile, slight DRX also occurs nearby GBs; with continuously raising T , the DRX degree is enhanced. Such a DRX mechanism is different from the conventional DRX, and it is based on strain induced boundary migration [16, 31, 34], namely, the operation of the non-basal slip at high T will give rise to the rotated regions in the vicinity of GBs; with further increasing strain, the sub-grains and ultimately high angle boundaries are formed in these regions by dynamic recovery and recrystallization. The DRX region becomes thicker as T increases, as shown in Figs. 7, 8 and 10. This deformation process can only accommodate limited strain; however, once the first recrystallized grains are formed, the subsequently formed recrystallized grains tend to be clustered, together with the first grains, into a large band area around GBs, and the orientations of these grains are more favorable for the development of basal slip, thus leading to the notable decrease in flow stress, the appearance of a stress quasi-plateau or the occurrence of continuous strain softening, and even to the formation of cracks with larger size nearby GB where the DRX has occurred.

However, at a high $\dot{\epsilon}$, the diffusion-assisted mechanisms are limited. In addition, the CRSSs may be still higher for the pyramidal slip than for the prismatic slip [12, 15], and therefore, only the latter and basal slip are active. Therefore, there exists a strong resistance to DRX, as is observed in the current study, e.g., at 423 K and two higher $\dot{\epsilon}$, the DRX does not occur and thus, the flow stress is still higher and no stress quasi-plateau occurs. However, as $T \geq 473 \text{ K}$ which is higher than the recrystallization temperature of commercial pure Mg ($\approx 0.35T_m - 0.4T_m$, where $T_m = 923 \text{ K}$, the melting point of pure Mg [35]), the DRX slightly occurs nearby GBs at 473 K and becomes

enhanced at 523 K. Accordingly, a stress quasi-plateau forms, and even continuous strain hardening occurs. However, for CP Ti, only a quasi-steady flow state after a short-term rapid hardening stage is observed, as T is as high as 873 K, which is above the recrystallization temperature [36], while no fine recrystallized grains were observed.

5 Conclusions

1. With raising T or decreasing $\dot{\epsilon}$, the flow stress of AZ31 Mg alloy largely decreases, meanwhile, strain hardening degree becomes weakened after an initially rapid strain hardening, even an obvious stress quasi-plateau appears or continuous strain softening occurs. The σ_{UTS} and σ_{YS} continuously decrease with raising T at different strain rates, but the decreasing amplitude is comparatively greater at the low $\dot{\epsilon}$. At a certain T , an increase in $\dot{\epsilon}$ raises strength, particularly at increasing temperatures.
2. At the temperatures below 423 K, the plastic deformation of AZ31 Mg alloy is governed by the predominant twinning and dislocation slip with a limited amount; with increasing $\dot{\epsilon}$, twinning is enhanced. As T is over 423 K, twinning becomes weakened; meanwhile, the non-basal slip and the DRX become enhanced, especially at low $\dot{\epsilon}$.
3. With raising T or decreasing $\dot{\epsilon}$, the formation mode of cracks is found to change from along GBs and at TBs or the end of twins into nearby GBs where DRX has obviously occurred, and the enhancement of DRX nearby GBs resulted in the formation of large-sized cracks along GBs.

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