Dynamic tensile behavior of AZ31B magnesium alloy at ultra-high strain rates

Geng Changjian a, Wu Baolin b,*, Liu Fang a, Tong Wenwei a, Han Zhenyu a

a Aviation Industry Corporation of China Engine Design and Research Institute, Shenyang 110015, China
b School of Materials Science and Engineering, Shenyang Aerospace University, Shenyang 110136, China

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AZ31B magnesium alloy; Schmid factor; Slip; Texture; Twinning

Abstract The samples having {0001} parallel to extruding direction (ED) present a typical true stress–true strain curve with concave-down shape under tension at low strain rate. Ultra-rapid tensile tests were conducted at room temperature on a textured AZ31B magnesium alloy. The dynamic tensile behavior was investigated. The results show that at ultra-high strain rates of $1.93 \times 10^2 \text{s}^{-1}$ and $1.70 \times 10^3 \text{s}^{-1}$, the alloy behaves with a linear stress–strain response in most strain range and exhibits a brittle fracture. In this case, $\{10\overline{1}2\} < 10\overline{1}1>$ extension twinning is basic deformation mode. The brittleness is due to the macroscopic viscosity at ultra-high strain rate, for which the external critical shear stress rapidly gets high to result in a cleavage fracture before large amounts of dislocations are activated. Because $\{10\overline{1}2\}$ tension twinning, $\{10\overline{1}1\}$ compressive twinning, basal $<a>$ slip, prismatic $<a>$ slip and pyramidal $<c+a>$ slip have different critical shear stresses (CRSS), their contributions to the degree of deformation are very differential. In addition, Schmid factor plays an important role in the activity of various deformation modes and it is the key factor for the samples with different strain rates exhibit various mechanical behavior under dynamic tensile loading.

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1. Introduction

Magnesium (Mg) alloys, as a kind of the lightest structural materials, have attracted considerable attention in the aerospace and automobile industries. Due to the HCP crystallographic structure with the c/a ratio of 1.624, the deformation of Mg depends mainly on basal $<a>$ slip at room temperature. Although critical resolved shear stress (CRSS) of the basal slip at room temperature is approximately 1/100 of those non-basal slips on prismatic and pyramidal planes, it provides only two independent slip systems, far fewer than the necessary five independent systems for homogeneous deformation according to the von Mises criterion. So a fulfillment of the von-Mises condition would also require the activation of the non-basal slips and mechanical twinning. The information about the deformation mechanism under various conditions is thus fundamental for understanding magnesium mechanical behavior.

* Corresponding author. Tel.: +86 24 89728701.
E-mail address: wubaolin@sau.edu.cn (B. Wu).
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The relative activity of each deformation mode, which depends on many factors such as orientation of deformed grains, deformation temperature and strain rate, etc.\textsuperscript{5,6} is the key accordance for prediction of the deformation behavior. Strain rate sensitivity is an important factor to be considered when the mechanical behavior under dynamic loading is concerned. Carlson\textsuperscript{9,10} found that in excess of $1 \times 10^3$ s\(^{-1}\), strain rate sensitivity for magnesium alloys increases dramatically. Some components suffering from serious damage in real applications. In order to research the dynamic behavior of components, the mechanical behavior of AZ31B magnesium alloy under ultra-rapid tension was investigated and the mechanism of stress–strain response was discussed in the present work. The relation between the yielding strength, fracture strength and total strain with increasing strain rate will be studied. Data sustainment was provided by different mechanical behaviors under static tension. The electron back scattering diffraction and transmission electron microscopy scanning were used to study the microstructure of the deformed grains, orientation of grains, the kinds and number of twinning, distribution of dislocation slip. The deformed mechanism was investigated in-depth. Technology sustainment was provided for the solution of components fracture analysis and the security of components’ working in real condition, so some understanding dynamic behavior of components is important for proper design.

2. Experimental

The as-extruded AZ31B alloy rod with a size of $\varnothing 70$ mm $\times$ 80 mm was annealed at 793 K for 80 min for removing the residual stress and making a complete recrystallization. The AZ31B tensile samples with gauze size of $\varnothing 5$ mm $\times$ 25 mm were machined out of the rod with the tensile axis parallel to the extrusion direction (ED). The nominal composition of AZ31B alloy is listed in Table 1. 

In the present work, the texture was measured with X-ray back diffraction technique on the longitudinal section of the as-extruded and annealed rod. Based on the orientation distribution function (ODF) calculation, the ED inverse pole figure was deduced. The ultra-rapid tensile behavior of the investigated alloy was characterized at ultra-high strain rates of $1.93 \times 10^3$ s\(^{-1}\) and $1.70 \times 10^3$ s\(^{-1}\). To ensure the repeatability and accuracy, each test condition was repeated at two times. Tensile tests were performed on Zwick/Roell HTM5020 test machine in the axial direction of the samples. The deformed microstructures were observed under optical microscope and revealed by EBSD analysis. EBSD and fracture morphology observation were performed on the JEOL-JSM-7001F SEM equipped with the acquisition software (Channel 5).

3. Results and discussion

3.1. Initial microstructure and texture

The microstructure of the as-extruded and annealed rod and the corresponding ED inverse pole figure are shown in Fig. 1. It can be seen that after annealing, the grains of the alloy are equiaxed with the average size of about 15 $\mu$m (Fig. 1(a)). The texture of the rod is the typical fiber consisting of two components. As the tensile samples were cut with their axis parallel to ED, their texture can be considered as the same as that of the rod, i.e., the tensile axis concentrates intensively on $<10-10>$ and $<11-20>$ directions (in Fig. 1(a), $T_{\text{max}}$ the maximum value, no unit). This means that most grains orientate with the c-axis perpendicular to the tensile axis. For grains distributed like this, the basal slip and $<10-12>$ extension twinning are limited because of their very low and negative Schmid factors, meanwhile $<10-11>$ contraction twinning is favored due to the contraction stress along the c-axis.\textsuperscript{11} In addition, prismatic $<a>$ and pyramidal $<c+a>$ slips could also be activated in this hard orientation. Schmid factor is the important parameter to be considered when the deformation modes are predicted. However, CRSS should also be concerned.

3.2. Stress-strain responses

Fig. 2 presents the true stress–strain curves respectively at the strain rates of $2.80 \times 10^{-3}$ s\(^{-1}\), $1.93 \times 10^3$ s\(^{-1}\), and $1.70 \times 10^3$ s\(^{-1}\) at room temperature. It can be found that both the total strain and maximum stress at the ultra-high strain rates ($1.93 \times 10^3$ s\(^{-1}\) and $1.70 \times 10^3$ s\(^{-1}\)) are much larger than those at strain rate of $2.8 \times 10^{-3}$ s\(^{-1}\). The total strain at the strain rate of $1.93 \times 10^3$ s\(^{-1}\) is almost contributed by elastic deformation. At the strain rate of $1.70 \times 10^3$ s\(^{-1}\), the elastic stress–strain curve is interrupted by a platform at about 350 MPa. The above results suggest that the deformation is more elastic at ultra-high strain rate.

<table>
<thead>
<tr>
<th>Table 1</th>
<th>Chemical composition of AZ31B alloy.</th>
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<tbody>
<tr>
<td>Element</td>
<td>Al</td>
</tr>
<tr>
<td>wt%</td>
<td>2.7</td>
</tr>
</tbody>
</table>

![Fig. 1](image_url) Microstructure and inverse pole figure of extruded and annealed rod.
This wide range of elastic strain reflects brittle trend and the effect of the ultra-high strain rate on the tensile behavior of the alloy, which is quite different from that under the rapid tension in our previous research. In that case, the stress–strain curves presented apparent yielding and plastic deforming, on which a “quasi-horizontal step” appeared at the strain rate ranging from $5.6 \times 10^{-3}$ s$^{-1}$ to $1.1 \times 10^{-1}$ s$^{-1}$. Generally, the brittleness of a material has been attributed to intrinsic and environmental factors. Of the intrinsic factors, low mobility of dislocation is the key point. Although the magnesium alloy is not an intrinsic brittle material, the ultra-high strain rate could result in the lack of dislocation slips. This can be explained in terms of dislocation dynamics. As indicated, moving dislocation dissipates its moving energy to the surrounding and such dissipation is equivalent to the exertion on it of a drag resistance force, which depends linearly on the strain rate $\dot{\varepsilon}$, i.e., $\tau_D = \eta \dot{\varepsilon}$, where $\eta$ is macroscopic viscosity; therefore, the shear stress to activate dislocation slip is $\tau = \tau_c + \eta \dot{\varepsilon}$, where $\tau_c$ is critical shear stress for dislocation moving. Due to the ultra-high strain rate in the present study, the external critical shear stress may be so high as to result in a cleavage fracture before large amounts of dislocations activated. The increasing strength and brittleness under dynamic tension condition was also revealed in the TiAl alloy in Ref. [18] at ultra-high strain rate. However, TiAl is an intrinsic brittle material, quite different from magnesium alloys. In addition to slip, twinning is another important deformation mode in the investigated magnesium alloy. As a rapid shear mode, twinning is favored under dynamic loading, but it also needs the plastic relaxation of the internal stress by local dislocation slip as revealed by Barnett et al. so the activation of twinning should also be restricted under ultra-rapid deformation.

3.3. Deformed microstructures and deformation mechanisms

The deformed microstructures after tensile fracture at the two ultra-high strain rates are shown in Fig. 3(a) and (b). Some twin-like crystals can be found in the figures, but they are not in a large number. This is attributed to the ultra-high strain rate effect as discussed above.

In order to identify the types of the twins, the crystallite boundaries with three disorientations of $<11-20>$, $<11-20>$, and $<11-20>$ with $5^\circ$ deviation ambit were set as blue, yellow and red color, respectively, in the EBSD orientation maps (OM) (as shown in Fig. 4). The misorientation distribution profiles at the right of the maps well correspond to the presented twin boundaries. However, accurate identification needs to determine not only the rotation axis and angle between the twin and matrix, but also the habit plane and shear direction of the twinning. With the subset function of Channel 5, the disorientation relationships between the crystallites and the matrix were determined, and as examples, presented in Fig. 5. $F_x$, $F_y$, $F_z$ are the relative values in $x$, $y$, $z$ axes. For the blue boundaries, the crystallographic direction $<11-20>$ of the crystallite and the matrix, as circled, is coincident (Fig. 5(a)). The habit plane $K_1$ and twinning direction $n_1$ of the extension twin are respectively $<10-12>$ and $<10-11>$ which are also coincident for both the crystallite and matrix, as indicated by the circles. This suggests that the crystallites indicated by blue lines are $<10-12>$ extension twins with about $<11-20>$ disorientation with the matrix. Similarly, the crystallites with yellow and red lines were determined as referring to $<10-11>$ contraction twins and $<10-11>$–$<10-12>$ double twins according to the disorientation relationships shown in Fig. 5(b) and (c).

In the deformed microstructure shown in Fig. 4(b), the rapid shear band-like zone can be found as indicated with dot lines. The shear band-like zones seem to correspond to the platform on the curve at the strain rate of $1.70 \times 10^3$ s$^{-1}$ in Fig. 2. The formation process of the zones is yet difficult to be revealed at present, but twinning-related mechanism can be predicted according to the morphology of the deformed microstructure in Fig. 4(b). As discussed, dislocation slip is strongly sensitive to strain rate and it is restricted at the ultra-high strain rate of $1.93 \times 10^2$ s$^{-1}$. At this strain rate, twins
are also found in small number as shown in Figs. 3 and 4(a), which may correlate with the fact that the deformation is almost elastic. At strain rate of $1.70 \times 10^3 \text{s}^{-1}$, the increment of strain rate results in more twins (Fig. 4(b)). Unlike slips accompanied with dislocation proliferations, tangles and interactions, which generally result in the strain hardening, twinning shows less hardening effect. This could be the reason for the platform showing up on the curve at the strain rate of $1.70 \times 10^3 \text{s}^{-1}$ in Fig. 2. In another aspect, because of the fiber texture, the tension twinning in most grains is difficult to be activated. Severe deformation zones could be favorable for the extension twinning to form the twinning-related band-like zones. It should be indicated that different from the microstructure shown in EBSD-OM, which was taken on the longitudinal section of the tensile sample (Fig. 4(b)), the optical microstructure (Fig. 3(b)) was observed on the cross section. So the rapid shear band-like zones were not observed under this condition.

It should be noted that the tensile samples have fiber texture components. In this orientation, the contraction twinning is favored, but the present result exhibits more extension twins than the contraction and the double twins. This tendency could be related to the comparison of the comprehensive function of Schmid factor and CRSS between the extension and contraction twinning. Fig. 6 presents the calculated Schmid factor distributions of the contraction and extension twin variants which have the maximum values in large range in the stereographic projection of the orientation space under uniaxial tension. As shown, for the present texture, Schmid factor of the contraction twin variant is higher than that of the extension twin variant in most grains in the large $\tan(\theta/2)$ range (where $\theta$ is the tilt angle in the inverse pole figure, $\theta_0$ is the initial value). However, CRSS value ($76–153 \text{MPa}^{\text{20}}$) of the contraction twin variant is also much higher than that of the extension twin variant ($2.0–2.8 \text{MPa}^{\text{20}}$). Thus, the contraction twinning is activated only in a small orientation range. Contrarily the extension twinning is activated in a large range. In this case, although a majority of grains have their c-axis perpendicular to tensile loading due to the fiber texture, the grains deviating from the fiber orientation are more easily deformed by extension twinning as discussed above. This could be the main reason that more extension twins have been observed in the deformed microstructures. In addition, it should be noted that even during unloading process, twinning and detwinning can be induced,\textsuperscript{21} which is driven by internal local stress resulting from stress redistribution among the differently oriented grains.

Fig. 7 presents SEM micrographs showing morphologies of the fracture surfaces of the ultra-rapidly tensioned specimens. All the fracture surfaces were covered with cleavage.
Fig. 5 Pole figures with scattered data corresponding to twins and matrix indicated by arrows in Fig. 4(a).

Fig. 6 Schmid factor distribution under uniaxial tension.
morphology, which indicates that the fracture proceeds with less dislocation slips, demonstrating the brittleness of the investigated alloy under the ultra-rapid tension.

4. Conclusions

(1) At ultra-high strain rates of $1.93 \times 10^2$ s$^{-1}$ and $1.70 \times 10^3$ s$^{-1}$, the alloy behaves with a linear stress–strain response in most strain range. Both the total strain and maximum stress at the ultra-high strain rates are much larger than those at the quasi-static strain rate. The fracture exhibits a brittle feature.

(2) $\{10-12\} <10-11>$ extension twinning is dominant deformation mode at the ultra-high strain rate. The shear band-like zone can be predicted to be related to the twinning-related activation, resulting in the platform on the stress–strain curve at the strain rate of $1.70 \times 10^3$ s$^{-1}$. (3) Due to the ultra-high strain rate, the external critical shear stress gets rapidly high to result in a cleavage fracture before large amounts of dislocations activated because of macroscopic viscosity.

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References


**Geng Changjian** received the Ph.D. degree in material Science from Northeastern University in 2012, and then became an engineer there. His main research interest is deformation mechanism of aeroengine materials.