Mechanical Behavior and Microstructural Analysis of Extruded AZ31B Magnesium Alloy Processed by Backward Extrusion

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This study investigates the mechanical behavior of an extruded AZ31B magnesium alloy profile at various strain rates from 0.001 to 375/s. The electron backscatter diffraction analysis revealed that the profile has $\{0001\}\langle 10\bar{1}0\rangle$ and $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ textures. Due to the textures, the profile exhibits pronounced anisotropy in mechanical properties. In the extrusion direction (ED), the profile shows the highest yield strength (YS) but the lowest total elongation at fracture (TE) due to a hard activation of non-basal slip and $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning; in the diagonal direction (DD), it shows the lowest ultimate tensile strength (UTS) but the highest TE due to an easy activation of basal slip; in the transverse direction (TD), it shows the lowest YS due to an easy activation of $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ twinning. Moreover, the number of twins increases with the increasing strain rate. This indicates that deformation twinning becomes prevalent to accommodate high-rate deformation. Due to the different deformation mechanisms, the profile exhibits an orientation-dependent effect of strain rate on the mechanical properties. A positive effect of strain rate on the YS and UTS was found in the ED, while the effect of strain rate on the YS is negligible in the DD and TD. The TE in the ED, DD, and TD decreases in general as the strain rate increases. Fractographic analysis under a scanning electron microscope revealed that the fracture is a mixed mode of ductile and brittle fracture, and the magnesium oxide inclusions could be the origins of the fracture.

Keywords automotive and transportation, dynamic tensile test, fractographic analysis, magnesium, strain-rate sensitivity, texture

1. Introduction

Automotive weight reduction has been a trend to meet increasingly strict fuel efficiency and emission standards and help solve environmental issues. In this regard, magnesium (Mg) alloys have attracted increasing attention due to their low density (~ 1.77 g/cm³) and high strength-to-weight ratio (Ref 1, 2). As far as vehicle crashworthiness is concerned, however, the introduction of Mg alloys in automotive body structures is still a challenge (Ref 3). In crash events, impacted structures generally experience severe plastic deformation within the local zones at room temperature and intermediate strain rates ranging up to 500/s (Ref 4). In such circumstances, Mg alloys exhibit low ductility and fracture toughness due to the hexagonal closed-packed (HCP) structure, consequently causing premature failures of impacted structures. Therefore, structural applications of wrought Mg alloys are restricted although some non-structural casting and forging products have been used in automotive structures.

Recently, the current authors (Ref 5) reported that thin-walled Mg beams can significantly outperform mild steel beams in terms of energy absorption capacity when the beams were filled with polyurethane foams and subjected to dynamic bending loads. It therefore indicates that wrought Mg alloys have potential applications in automotive body structures, such as rocker rails, door beams, and bumper beams, if suitable structural design methods are developed to control the deformation modes and suppress the failures of Mg structures. To support the design of Mg crashworthy structures, it is therefore worthwhile to study the dynamic mechanical behavior of wrought Mg alloys.

Extensive studies have reported that wrought Mg alloys exhibit significant plastic anisotropy, tension-compression asymmetry, and temperature sensitivity under quasi-static loading (Ref 6–8); however, the studies on the effect of strain rate on the mechanical properties are relatively few (Ref 9–20). Mukai et al. (Ref 9) observed that the tensile strength and ductility of extruded AZ31 bar with grain refinement via equal channel angular extrusion increased with increasing the strain rate even higher than 10³/s. El-Magd and Abouridouane (Ref 10) presented that the ductility of extruded AZ80 bar decreased with increasing the strain rate lower than 10³/s; while it increased at strain rates higher than 10³/s. Lin and Chen (Ref 11) performed a microscopic examination of the fracture morphologies for extruded AZ31B and observed that the ductility decreases and the brittle fracture characteristics increase with increasing the tensile strain rate. Khan et al. (Ref 12) reported that the strain-rate sensitivity of rolled AZ31 was more pronounced at higher temperatures and lower strain rates. Kurukuri et al. (Ref 14) observed that the tensile ductility of rolled AZ31B sheet at strain rates higher than 5×10^2 /s was lower than that at quasi-static strain rates. Asgari et al. (Ref 15)

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presented that the strength and ductility of rolled AZ31B under high strain-rate compressive loading increased with decreasing grain size, while twinning fraction and strain hardening rate were found to decrease with decreasing grain size. Feng et al. (Ref 13) found that the tensile fracture pattern of rolled AZ31B bar was mainly quasi-cleavage pattern at room temperature, while mainly ductile fracture pattern at high strain rates and temperatures. Geng et al. (Ref 16) observed that extruded AZ31B bar had less dimple patterns but more cleavage patterns on the fracture surfaces of ultra-rapidly tensioned specimens. As reviewed above, it can be concluded that Mg alloys have a

Fig. 1 Cross section of the extruded AZ31B profile and the orientation designations. ED: extrusion direction, DD: diagonal direction, TD: transverse direction, ND: normal (thickness) direction

positive effect of strain rate on the tensile strength, but the effect of strain rate on the fracture behavior is scattered and depends upon the microstructures.

Compared with the research on the dynamic mechanical behavior of Mg alloys in the forms of rolled sheets and extruded bars/plates/solids, relatively little attention has been paid to that of extruded Mg profiles in the open literature. Therefore, to study the mechanical behavior of extruded Mg profiles is of significance since they have potential applications in automotive body structures. This study characterizes the tensile behavior of an extruded AZ31B Mg profile at room temperature and intermediate strain rates. The plastic anisotropy and the effect of strain rate on the mechanical properties are analyzed. The microscopic analyses are performed to study the deformation and fracture mechanisms. Finally, various constitutive models are examined to describe the strain hardening behavior and strain-rate sensitivity.

2. Experimental Procedure

2.1 Material and Microstructure

This study investigated an extruded AZ31B (Mg-3Al-1Zn, wt.%) Mg alloy profile with 3 mm wall thickness. Fig. 1 illustrates the cross section of the extruded profile and the orientation designations. The cross section was designed by simplifying a typical geometry of rocker rails. The profile was processed by backward extrusion and then quenched in air (F temper). The reason to employ backward extrusion rather than advanced processes such as equal channel angular extrusion and hydrostatic extrusion is that backward extrusion is

Fig. 2 (a) Initial microstructure, (b) orientation map in the TD, and (c) pole figures of some important texture components for the extruded AZ31B profile

beneficial to extrude larger billets with lower forces and it remains one of the primary options for mainstream automotive products.

The microstructure of the profile sectioned normal to the TD (TD section) was analyzed by electron backscatter diffraction (EBSD), as shown in Fig. 2(a). To improve analysis accuracy, the measured data at the four adjacent regions were merged together to involve enough grains in microstructural analysis. It

Fig. 3 Schematic diagram of the ND and TD components in the specimen coordinate system

Fig. 4 Dynamic uniaxial tensile test setup: (a) Instron VHS 100/20 testing system; (b) specimen attached with strain gages

Fig. 5 (a) Custom specimen geometry for dynamic tensile testing and (b) ASTM standard specimen geometry for quasi-static tensile testing

reveals that the grains have a preferential alignment with a longer axis in the ED and have a broad grain size distribution with an average intercept size of about 52.9 μm . As shown in Fig. 2(b) and (c), the intensity distribution of the poles reveals that there are $\{0001\}\langle 10\bar{1}0 \rangle$ and $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$ textures. Some c -axes of the grains are preferentially aligned parallel to the ND, resulting in a $\{0001\}\langle 10\bar{1}0 \rangle$ texture (called the ND component, see Fig. 3); while the other c -axes are preferentially aligned parallel to the TD, resulting in a $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$ texture (called the TD component). The ND component predominates. Moreover, it reveals that the intensity distribution of the basal $\{0001\}$ pole is distinctly broader between the ND and TD than between the ND and ED. The c -axis of the ND component is tilted about $\pm 0 - 30^\circ$ away from the ND towards the TD. The c -axis of the TD component is tilted about $\pm 0 - 15^\circ$ away from the TD towards the ND. Similar textures were also observed by Yu et al. (Ref 21) and Krajewski et al. (Ref 22) for extruded AZ31 plates, while different textures were reported by Xu et al. (Ref 23, 24) who observed that the TD component predominates in extruded AZ31 and AM30 sections.

2.2 Quasi-Static and Dynamic Tensile Tests

For dynamic tensile testing, no international testing standards are available. This work employed a servo-hydraulic testing machine Instron VHS 100/20. As shown in Fig. 4(a) and (b), the testing system mainly comprised a Kistler 9361B piezo-electric load cell, an upper grip, a lower grip, a hydraulic actuator, a high speed camera Photon FASTCAM SA-Z, and two high-power lamps. As shown in Fig. 5(a), the custom specimen geometry for dynamic tensile testing was carefully designed using finite element simulation and experimental verification. The function of the holes at two ends is to clamp and connect the specimen to the upper and lower grips by screws, nuts, and pins. The dynamic tensile tests along the ED were conducted at room temperature and velocities of 0.035, 0.2, 3, 6.3, and 14 m/s, and it obtained averaged strain rates of 1, 5, 93, 186, and 375/s, respectively.

As shown in Fig. 4(b), the elongation in the gage section was measured by a high-elongation strain gage, Vishay Measurement EP-08-125AD-120. Force measurement is a crucial procedure in dynamic tensile testing. The tensile force measured by the Kistler load cell generally exhibits significant oscillations due to the “ringing” of the loading system when the testing velocity is higher than 1.0 m/s. To reduce the oscillations, the tensile force was indirectly measured by measuring the elastic tensile strain in the upper grip section. To compensate the bending effect, the elastic tensile strain was measured by two general-used strain gages CAE-06-125UN-120, which were attached to the both side surfaces of the upper grip section. To guarantee uniform elastic deformation on the cross section where the CAE strain gages were attached, the position of the CAE strain gages was optimized using finite element simulation and experimental verification.

Quasi-static tensile tests were conducted on a universal testing machine Zwick/Roell Z250. In the ED, the custom specimens were employed. In the DD and TD, the standard specimens shown in Fig. 5(b) were employed due to the limited dimensions of the profile shown in Fig. 1. To obtain averaged strain rates of 0.001, 0.01, and 0.1/s, the custom specimens were tested at crosshead velocities of 0.035, 0.35, and 3.5 mm/s, respectively, while the standard specimens were tested at crosshead velocities of 0.036, 0.36, and 3.6 mm/s. In order to

Fig. 6 Engineering stress-strain curves of the extruded AZ31B profile at various strain rates: (a) in the three directions, (b) in the ED, (c) in the DD, and (d) in the TD

Table 1 Mechanical properties of the extruded AZ31B profile at various strain rates

Orientation	Strain rate, 1/s	YS, MPa	UTS, MPa	TE, %
ED	0.001	182	227	15.3
	0.01	190	229	14.2
	0.1	196	230	12.8
	1	198	230	12.0
	5	201	234	11.8
	93	211	259	11.4
	186	216	272	11.2
	375	224	284	11.8
DD	0.001	69	218	20.1
	0.01	72	220	18.4
	0.1	75	221	16.2
TD	0.001	64	237	15.8
	0.01	65	241	15.2
	0.1	65	244	14.8

prevent initial damages such as cracks or overheating zones on the edges, all the specimens were prepared by water-jet cutting and then sanded along the cutting edges.

2.3 Microscopic Examinations

The microstructures of some selected specimens were observed via an optical metallographic microscope Zeiss

Axiovert 200MAT after etching with an acetic-picral solution. To analyze the texture, the samples were carefully prepared via grinding with 2000 SiC paper, mechanical polishing with 0.05 μm silica suspension, and final electro-chemical polishing for 90 s at 33 V and -30°C using a 10% perchloric acid and 90% methanol. The prepared sample was measured by EBSD using a Hitachi scanning electron microscope (SEM) and AZtecHKL software. The fracture surface morphologies of the broken specimens were analyzed using a SEM (Shimadzu Superscan SSX-550) equipped with an energy dispersive x-ray spectroscopy (EDS) analyzer.

3. Experimental Results

3.1 Tensile Flow Behavior

3.1.1 Plastic Anisotropy. The engineering stress-strain curves at various strain rates are plotted in Fig. 6. The basic mechanical properties are summarized in Table 1. It is seen that the material exhibits pronounced anisotropy of strength and ductility. At a strain rate of 0.001/s, the 0.2% offset yield strength (YS) in the ED is the highest (182 MPa). The YS in the DD and TD is less than half of the value in the ED. The ultimate tensile strength (UTS) is highest (237 MPa) in the TD and lowest (218 MPa) in the DD. But the difference of the UTS

among the three directions is relatively small. The total elongation (TE) at fracture is also anisotropic—the TE (20.1%) in the DD is significantly higher than the TE (15.3%) in the ED and the TE (15.8%) in the TD. The TE is

Fig. 7 Total elongation at fracture of the extruded AZ31B profile with respect to true strain rate

slightly lower in the ED than in the TD. Moreover, the stress-strain curves show different strain hardening rate depending on the loading direction.

3.1.2 Strain-Rate Sensitivity. For the ED specimens, as shown in Fig. 6(b) and Table 1, the YS and UTS increase as the strain rate increases from 0.001 to 375/s. The differences of the YS and UTS between strain rates of 0.001/s and 375/s are about 42 and 57 MPa, respectively. It indicates a significant positive effect of strain rate on the YS and UTS in the ED. In the DD and TD, as shown in Fig. 6(c) and (d) and Table 1, the effect of strain rate on the YS is relatively insignificant or negligible; however, the effect of strain rate on the subsequent flow stress is not negligible. Fig. 7 shows the TE with respect to true strain rate. In all the three directions, the TE decreases in general as the strain rate increases in the tested range of strain rate.

3.2 Twinning Mechanisms

The deformed microstructures at the fracture zones (TD sections) of the deformed ED specimens tested at strain rates of 0.001, 1, and 186/s are shown in Fig. 8. A considerable number of twins are visible under an optical microscope, being arranged in needle-like narrow bands. Moreover, an increase of the

Fig. 8 Microstructures at the fracture zones (TD sections) of the deformed ED specimens tested at various strain rates: (a) 0.001/s, (b) 1/s, and (c) 186/s

Fig. 9 EBSD orientation maps of the microstructure in the deformed ED specimen at a strain rate of 1/s: (a) pole map; (b) $\{10\bar{1}1\} - \{10\bar{1}2\}$ double twins, 37.5° ; (c) $\{10\bar{1}1\}$ contraction twins, 56° ; (d) $\{10\bar{1}3\}$ contraction twins, 64° ; (e) $\{10\bar{1}2\}$ extension twins, 86.3°

Fig. 10 Image series of the ED specimens in tension at strain rates of (a) 1/s and (b) 186/s, (c) macroscopic appearance of a typical fracture area of an ED specimen

number of twins was also observed when the strain rate increases from 0.001 to 186/s, although the increase is small possibly due to the relatively small jump of strain rate. Similar results were also reported by Ulacia et al. (Ref 18) and Jiang et al. (Ref 25). It indicates that deformation twinning becomes prevalent to accommodate high-rate deformation.

To identify the twins, the microstructure at a strain rate of 1/s was analyzed by EBSD, as shown in Fig. 9. The twinning boundaries with four reorientations of 37.5° , 56° , 64° , and 86.3° with 5° deviation were set as red lines in separate EBSD orientation maps. Herein the angles refer to the angle through which the basal planes are rotated around the direction $\langle \bar{1}2\bar{1}0 \rangle$. The twins with 86.3° , 56° , and 64° reorientations are $\{10\bar{1}2\}$ extension twins, $\{10\bar{1}1\}$ contraction twins, and $\{10\bar{1}3\}$ contraction twins, respectively, while the twins with 37.5° reorientations are primary $\{10\bar{1}1\}$ – $\{10\bar{1}2\}$ double twins, respectively (Ref 26). The $\{10\bar{1}1\}$ contraction twins have the largest volume fraction, while the $\{10\bar{1}2\}$ extension twins have the smallest volume fraction. It indicates that $\{10\bar{1}1\}$ contraction twinning is a predominant twinning mode in the ED specimens. Moreover, the presence of double twins might be beneficial for plastic deformation as it reorients the basal planes so that it is more favorably aligned for slip.

3.3 Fractographic Analysis

As analyzed above, the tensile fracture behavior of the extruded profile intensively depends upon the strain rate and specimen orientation. The deformation process of the ED specimens tested at strain rates of 1 and 186/s are manifested in Fig. 10(a) and (b), respectively. At each strain rate, the four images show the specimen at initial state, uniform plastic deformation, necking, and the state after the fracture, respec-

tively. Unlike ductile steels and aluminum alloys, the extruded AZ31B specimens have no significant necking in the width direction and are quickly broken after necking. It can be seen from Fig. 10(c) that the ED specimen has a rough fracture surface with some brightly reflecting facets. Like in the width direction, no significant necking was observed in the ND. Similar macro-characteristics of fracture surface were also observed in the DD and TD specimens. Generally, the fracture of the extruded AZ31B profile can be characterized as brittle since no significant necking was observed prior to fracturing; however, the fracture on a macro-level can also be characterized as ductile since the TE exceeds 10% and even reaches about 24%, as shown in Fig. 7.

On a micro-level, the fracture process in metals is generally associated with the growth of micro-voids and local failure of second phase particles such as precipitates, oxide, and inter-metallic inclusions/particles. Fig. 11 shows the typical SEM fracture morphologies of the ED specimens tested at strain rates of 0.001 and 186/s. A rough fracture surface with some cleavage-like bands (indicated by the arrows) was observed at low magnification in Fig. 11(a). At both strain rates, the fracture surfaces shown in Fig. 11(b) and (e) are mainly composed of smooth cleavage facets without river patterns (indicated by A), rough cleavage facets with thin river patterns (indicated by B), and plastic deformation zones with dimples (indicated by C). Many second phase particles were also observed on the fracture surfaces, as indicated by ellipses in Fig. 11(c) and (e), which could be the origins of the fracture. These features reveal that the fracture of the ED specimens is a mixed mode of ductile and brittle fracture. In Fig. 11(b), a considerable number of dimples were observed on the fracture surfaces, which indicates that the material in these regions experienced considerable plastic deformation at a strain rate of 0.001/s. In contrast, the volume fraction of the dimples at a strain rate of 375/s is smaller than that at a strain rate of 0.001/s. This tendency is consistent with the results of the TE shown in Fig. 7.

In the TD and DD specimens, a mixed fracture mode of ductile and brittle fracture was also observed after tensile tests at a strain rate of 0.001/s, as shown in Fig. 12. Compared with the fracture morphologies of the ED specimen tested at a strain rate of 0.001/s, fewer cleavage facets but more dimples were observed on the fracture surface of the DD specimen shown in Fig. 12(a) and (b), which is consistent with the results of the TE shown in Fig. 7. It is noted that the TE is slightly lower in the ED than in the TD; however, more features of brittle fracture, i.e., more cleavage facets but fewer dimples, were observed on the fracture surface of the TD specimen shown in Fig. 12(c) and (d). It is difficult to explain the reason at present, while the difference of deformation mechanisms between the ED and TD specimens can be anticipated to be one of the major factors. The responsible factors should be investigated further.

Second phase particles in the forms of lump, lamellar, and granular particles were observed on the fracture surfaces of the analyzed samples, which could be the origins of the fracture. As shown in Fig. 13, backscattered electron (BSE) images were obtained, which provide atomic number contrast wherein regions of higher average atomic number appear brighter. The chemical composition tables obtained from the EDS spectrum analysis are embedded in the BSE images. Fig. 13(a) and (b) show the BSE images of the fracture surfaces corresponding to the SEM images shown in Fig. 11(c) and (e). The EDS

spectrum analysis data indicate that the second phase particles in the forms of lump (indicated by A) and lamellar particles (indicated by B) near the cleavage facets are magnesium oxide inclusions. Fig. 13(c) shows the BSE image of the fracture surface corresponding to the SEM image shown in Fig. 12(d). The second phase particles are in the forms of small granular particles (indicated by C and D) at the bottom of the dimples. The EDS spectrum analysis data indicate that the particle C is rich in Fe, Mg, and O with a smaller amount of other elements including Cr and Ni; therefore, the particle C is mainly ferric oxide inclusion with a smaller amount of intermetallic compounds. In general, the majority of the second phase particles are composed of magnesium oxide inclusions, which may be the main origins of the brittle fracture. It can be seen that the presence of magnesium oxide inclusions is a key factor that restricts the tensile ductility of the extruded AZ31B profile.

4. Constitutive Modeling

In order to apply the experimental stress-strain data to numerical simulations, various empirical constitutive models are examined. First, the Ludwig model was used to describe the strain hardening behavior due to its simplicity and robustness. It can be expressed as

$$\sigma = f_L = A + B \cdot (\bar{\epsilon}_p)^n, \quad (\text{Eq 1})$$

where A is the yield strength; B and n are the two strain hardening parameters; σ and $\bar{\epsilon}_p$ are the true stress and effective plastic strain, respectively. The best-fit curves by the Ludwig model are shown in Fig. 14(a) and the best-fit parameters are listed in Table 2. The approximation degree is measured by the adjusted R^2 value. It shows that the best fitting results are

Fig. 11 Typical SEM fracture morphologies of the ED specimens after tensile tests at different strain rates: (a) low magnification image of a typical fracture surface; (b, c) at a strain rate of 0.001/s; (d, e) at a strain rate of 375/s

Fig. 12 Typical SEM fracture morphologies of the TD and DD specimens after tensile tests at a strain rate of 0.001/s: (a, b) in the DD; (c, d) in the TD

Fig. 13 Typical BSE images and EDS spectrum analysis of the AZ31B specimens after tensile tests

Fig. 14 Comparison of experimental data and best-fit constitutive models for tensile testing of the extruded AZ31B profile: (a) Ludwig model; (b) biphasic dose-response (BDR) model

Table 2 Best-fit material parameters of various models for the extruded AZ31B profile

	Ludwig	Biphasic dose-response		Cowper-Symonds	
ED	$A: 175.63 \text{ MPa}$ $B: 225.98 \text{ MPa}$ $n: 0.483$ $\bar{R}^2: 0.998$	$A_1: 32.61 \text{ MPa}$ $B_1: 4.16\text{E}-3$ $h_1: 105.11$ $p: 0.1047$	$A_2: 289.13 \text{ MPa}$ $B_2: -0.041$ $h_2: 4.88$ $\bar{R}^2: 1.0$	$A: 179.74 \text{ MPa}$ $n: 0.520$ $p: 2.97$	$B: 233.42 \text{ MPa}$ $C: 2.65\text{E} + 4/\text{s}$ $\bar{R}^2: 0.986$
DD	$A: 52.09 \text{ MPa}$ $B: 496.19 \text{ MPa}$ $n: 0.475$ $\bar{R}^2: 0.995$	$A_1: -1.46\text{E}+3 \text{ MPa}$ $B_1: 4.94\text{E}-4$ $h_1: 203.98$ $p: 0.0128$	$A_2: 282.79 \text{ MPa}$ $B_2: -0.155$ $h_2: 5.68$ $\bar{R}^2: 1.0$	...	
TD	$A: 53.58 \text{ MPa}$ $B: 807.81 \text{ MPa}$ $n: 0.630$ $\bar{R}^2: 0.993$	$A_1: -789.18 \text{ MPa}$ $B_1: -0.0284$ $h_1: 52.85$ $p: 0.770$	$A_2: 283.20 \text{ MPa}$ $B_2: 0.0443$ $h_2: 14.14$ $\bar{R}^2: 1.0$	...	

Fig. 15 Measured true stress-effective plastic strain curves (symbols) at various strain rates for the extruded AZ31B profile in the ED and their fitted curves (lines) by the Cowper-Symonds model

satisfactory, moderately satisfactory, and unsatisfactory for the curves in the ED, DD, and TD, respectively. Other empirical models such as the Voce model and mixed H/V model (Ref 27) were also examined, but the results are still unsatisfactory for the curve in the TD. It is noted that the curve in the TD shows a degraded biphasic sigmoid shape. Therefore, a biphasic dose-response (BDR) function is proposed as follows

$$\sigma = A_1 + (A_2 - A_1) \left[\frac{p}{1 + 10^{(B_1 - \bar{\epsilon}_p)h_1}} + \frac{1 - p}{1 + 10^{(B_2 - \bar{\epsilon}_p)h_2}} \right] \quad (\text{Eq } 2)$$

where A_1 , A_2 , B_1 , B_2 , h_1 , h_2 , and p are the material parameters. As can be seen from Fig. 14(b) and Table 2, the model prediction agrees very well with the experimental data in all the three directions. Although the BDR model is complex due to too many parameters, it can be an alternative if high accuracy modeling is required in some applications.

To consider the effect of strain rate on the flow stress, the Cowper-Symonds model (Ref 28) was used to describe the rate-dependent stress-strain behavior in the ED. The model can be expressed as

$$\sigma = f_L \left[1 + \left(\frac{\dot{\bar{\epsilon}}_p}{C} \right)^{1/p} \right], \quad (\text{Eq } 3)$$

where f_L is defined in Eq 1; C and p are the material parameters for the strain-rate sensitivity; $\dot{\bar{\epsilon}}_p$ are the effective plastic strain rate. The best-fit curves by the Cowper-Symonds model are shown in Fig. 15 and the best-fit parameters are listed in Table 2. The model prediction agrees well with the experimental data in general. The Cowper-Symonds model was not applied to fit the data in the DD and TD due to the lack of experimental data at intermediate strain rates in these two directions.

5. Discussion

Besides basal a slip and prismatic a slip, the main deformation mechanisms in Mg alloys also include two deformation twinning mechanisms, i.e., $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning (extension twinning) and $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning (contraction twinning). Pyramidal $a + c$ slip can accommodate only a limited amount of deformation because of its high critical resolved shear stress (CRSS) value (Ref 29, 30). It is generally accepted that $CRSS_{\text{basal}} < CRSS_{\text{extension twinning}} < CRSS_{\text{prismatic}} \leq CRSS_{\text{pyramidal}}$ at room temperature (Ref 31–33). However, $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning behaves different from $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning and it has much higher CRSS value than $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning at room temperature (Ref 34). Furthermore, the deformation mechanisms are highly related to the orientation of deformed grains with respect to the direction of applied stresses, i.e., the so-called Schmidt factor. In Mg alloys, $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning can only be activated by applying a tensile stress parallel to the c -axis or a compressive stress perpendicular to the c -axis. On the contrary, $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning can only be activated by applying a compressive stress parallel to the c -axis or a tensile stress perpendicular to the c -axis.

In a rolling process for sheets, the metal billet is inevitably deformed in compression along the sheet ND and in tension along the rolling direction (RD). In such condition, $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning is easily activated if the c -axis is aligned parallel to the RD. When $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning occurs, the basal plane is rotated about 86.3° so that the basal plane is aligned parallel to the sheet plane. On the contrary, $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning is hardly activated if the c -axis is aligned perpendicular to the RD. Consequently, a rolled Mg sheet exhibits a strong basal texture (ND component). Different from rolling, during extrusion the metal billet is deformed in compression in all the three principle directions, as illustrated in Fig. 16. For example, in zone B of the metal billet, the stress in the ED is the first principal stress σ_1 (min. absolute value); the stress in the TD is the second principal stress σ_2 ; the stress in the ND is the third principal stress σ_3 (max. absolute value). Therefore, besides the ND component, the TD component produced by the compressive stress in the TD appears in the extruded Mg profile during backward extrusion.

Fig. 16 Schematic diagram of the stress state for a thin-walled profile during backward extrusion

The developed texture in the extruded Mg profile is responsible for its anisotropic mechanical properties. For the ED specimens, the tensile axis is nearly perpendicular to the c -axes of both the ND and TD components, resulting in a near-zero Schmidt factor for basal slip. It is not beneficial to activate basal slip and $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning, but it is beneficial to activate $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning. Therefore, as shown in Fig. 8 and 9, a considerable number of contraction twins were observed. Meanwhile, non-basal slip has to be activated in the favorably oriented grains to promote plastic deformation. The CRSS of non-basal slip and $\{10\bar{1}1\}\langle 10\bar{1}2\rangle$ twinning are higher than that of $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning; moreover, the plastic deformation contributed by $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning is limited since the magnitude of the twinning shear is small. This leads to the highest YS and the lowest TE in the ED.

For the DD specimens, the grains with the ND component have near-zero Schmidt factors for basal slip. However, the grains with the TD component have high Schmidt factors for basal slip, allowing an easy activation of basal slip during the early stage of deformation. The CRSS of basal slip is low, resulting in the relatively low YS, the lowest UTS, and the highest TE in the DD.

For the TD specimens, the tensile axis is nearly perpendicular/parallel to the c -axes of the ND/TD components, respectively, resulting in a near-zero Schmidt factor for basal slip in the both components and a high Schmidt factor for $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning in the TD component. The CRSS of $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning for AZ31B alloy was roughly equal to that of basal slip (Ref 29). Therefore, $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning is responsible for the lowest YS in the TD. As twinning has been exhausted, non-basal slip has to be activated in the favorably oriented grains. The basal planes of the twinned grains are typically rotated by 86.3° (Ref 35), so that the twinned grains are still in a hard orientation for basal slip. Furthermore, the twin boundaries are effective in inhibiting dislocation slip. Generally, the dislocation-dislocation and twin-dislocation interactions will increase UTS and strain hardening rate, but reduce the TE. To be noted, the TE is slightly lower in the ED than in the TD, but more features of brittle fracture were observed on the fracture surface of the TD specimens. It is difficult to explain the reason, while this might be caused by two factors: (i) the relatively small number of the grains with the TD component lead to a limited number of extension twins; (ii) the c -axes of the ND/TD components are not perfectly aligned parallel to the ND/TD, respectively, but tilt towards TD and ND, respectively, which results in the activity of basal slip.

The effect of strain rate on the YS can also be interpreted by the developed texture. It is noted from Fig. 6 that the profile in the ED has a positive effect of strain rate on the YS, while the effect in the DD and TD is insignificant or negligible. It indicates that the CRSS of basal slip and twinning are strain-rate independent, while the CRSS of non-basal slip have a positive strain-rate dependence. This is in agreement with the results from previous studies (Ref 17, 24, 31).

As discussed above, it is seen that the texture and deformation mechanisms play an essential role in the mechanical properties of the extruded AZ31B Mg profile. Therefore, to study further the deformation mechanisms under different conditions are of significance for material design and applications.

6. Conclusions

This study conducted uniaxial tensile tests at room temperature and various strain rates to characterize the mechanical behavior of an extruded AZ31B Mg profile. The main conclusions are listed as follows:

- (1) EBSD analysis reveals that the profile has $\{0001\}\langle 10\bar{1}0\rangle$ and $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ textures.
- (2) A considerable number of twins (mainly $\{10\bar{1}1\}\langle 10\bar{1}\bar{2}\rangle$ twins) are visible in the deformed ED specimens. The number of twins increases with increasing the strain rate, which indicates that deformation twinning becomes prevalent to accommodate high-rate deformation.
- (3) Pronounced anisotropy in the mechanical properties was observed. The ED specimens show the highest YS but the lowest TE due to a hard activation of non-basal slip and $\{10\bar{1}1\}\langle 10\bar{1}\bar{2}\rangle$ twinning. The DD specimens show the lowest UTS but the highest TE due to an easy activation of basal slip. The TD specimens show the lowest YS due to an easy activation of $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning.
- (4) A positive effect of strain rate on the YS and UTS was found in the ED, while the effect of strain rate on the YS is negligible in the DD and TD.
- (5) The TE of the extruded profile in the ED, DD, and TD decreases in general as the strain rate increases from 0.001 to 375/s.
- (6) SEM fractographic analysis revealed that the fracture is a mixed mode of ductile and brittle fracture. EDS spectrum analysis observed that the main second phase particles are magnesium oxide inclusions, which could be the origins of the fracture.

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