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# Influence of temperature and strain rate on flow stress behavior of twin-roll cast, rolled and heat-treated AZ31 magnesium alloys

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**Abstract:** The effects of temperature and strain rate on the flow stress behavior of twin-roll cast, rolled and heat-treated AZ31 magnesium alloys were investigated under uniaxial tension. At high temperatures, dynamic recovery, continuous dynamic recrystallization, grain boundary sliding and the activation of additional slip systems lead to an improvement of the ductility of the alloys. The elongation to failure is nearly independent of the strain rate between 473 and 523 K at  $10^{-2}$  s<sup>-1</sup> and  $10^{-1}$  s<sup>-1</sup>, which is related to the strain rate dependence of the critical resolved shear stress (CRSS) for nonbasal slip. Despite the high temperature, twins are even observed at 573 K and  $10^{-3}$  s<sup>-1</sup> because they have a low CRSS.

Key words: AZ31 magnesium alloy; twin-roll casting; strain rate; temperature; dynamic recrystallization; flow stress

# **1** Introduction

In the last decade magnesium alloys were brought into focus by the automotive and aircraft industries because of their high lightweight potential. In contrast with other structural materials, magnesium alloy has high specific strength, low density (1.74 g/cm<sup>3</sup>), good damping quality, respectable machinability, and excellent recyclability [1].

Because of the hexagonal close packing, magnesium and its alloys have a low formability at low temperatures. Magnesium does not exhibit five independent slip systems at room temperature, and according to von Mises criterion, it cannot produce uniform polycrystalline plastic deformation [2]. Therefore, dislocation glide on the basal plane and twinning are the most common deformation mechanisms in wrought magnesium alloys at low temperatures and strain rates. Furthermore, prismatic and pyramidal glide is also possible for hexagonal metals depending on the c/a axial ratio and the stacking fault energy [3,4]. Depending on the temperature, different deformation mechanisms are active because of their critical resolved shear stress (CRSS). In general, at low temperatures, this assumption is made: CRSS<sub>basal</sub> < CRSS<sub>twinning</sub> < CRSS<sub>prismatic</sub> < CRSS<sub>pyramidal</sub> [5]. It is assumed that the CRSS of basal slip is nearly independent of the temperature. In contrast to this, with rising temperature the CRSS of prismatic and pyramidal slip decreases. This results in an activation of additional slip systems above 498 K depending on alloying elements, and grain size, which causes an improvement of ductility and a decrease of flow stress [6]. ULACIA et al [7] suggested that the decrease of CRSS<sub>nonbasal</sub> with increasing temperatures is less distinct at high strain rates than at low strain rates.

Beside the dislocation motion on the basal plane, mechanical twinning plays an important role because of its low CRSS [8,9]. The CRSS for twinning is not a function of the temperature and twins are even active at high temperatures due to their low CRSS compared with nonbasal slip [5,10]. Twinning influences the texture development by changing the orientation of twinned grains which can cause hardening or softening [11]. In addition, twins can act as barriers for the dislocation motion because they interrupt the slip length [12,13]. Depending on the c/a axial ratio, tensile twinning is enhanced because of the decrease of individual shear. A compression of the *c*-axis results in the formation of

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contraction twins. Tension twins are formed when the *c*-axis is extended by external loading [9]. With decreasing the grain size, twinning is increasingly inhibited [14]. Below a critical grain size twins do no longer occur. BARNETT [5] found that the  $CRSS_{twinning}$  is not a function of the strain rate.

Other mechanisms during high temperature deformation of magnesium alloys are dynamic recovery (DRV) and dynamic recrystallization (DRX) [15]. DRV is represented by the rearrangement and the annihilation of moving dislocations. DRX can be divided into continuous (CDRX) and discontinuous dynamic recrystallization (DDRX) [16]. CDRX leads to the formation of high-angle grain boundaries out of subgrains. Therefore, a strong movement of low-angle grain boundaries is required. Consequently, CDRX is actually a distinct recovery process [17]. DDRX induces the nucleation and the growth of new undeformed fine grains which also cause an enhancement of the ductility. All DRX mechanisms are influenced by the temperature, strain rate, predeformation and dislocation density, grain size, and texture [7,17]. For AZ31 magnesium alloy, DRX starts above 473 K [18].

As noted by KOIKE [19], a fine grained microstructure can improve the ductility of magnesium alloys as well. Grain refinement and texture control in magnesium alloys can be achieved by alloy design [20,21], equal channel angular pressing (ECAP) also known as equal channel angular extrusion (ECAE) [22–25], conventional extrusion [26], accumulative roll-bonding (ARB) [27] or conventional hot rolling [28].

As described above, because of the existing deformation mechanisms, drawing of magnesium parts is only useful at elevated temperatures [29]. This results in high energy costs for the producers. The main goal of the manufacturing industry is to establish a cost-effective production of these lightweight alloys. In contrast to conventional hot rolling, twin-roll casting (TRC) does not need the roughing rolling of billets in several steps [30]. It is a combination of casting and rolling and influences the microstructure, e. g. the grain size, by rapid solidification on the sheet surface and Scheil solidification in the sheet thickness [30,31]. It has the

advantage that alloying elements with limited solid solubilities, which are prevented in ingot-casting, can be used [32]. In comparison to cast magnesium alloys, wrought magnesium alloys show a strong basal texture, which can influence the anisotropy and formability [33,34]. This can be reduced by alloy design or specific rolling and subsequent heat-treatment after TRC processing [35–37].

Previous studies determined the deformation behavior of the conventional rolled AZ31 magnesium alloy. However, there are no investigations on the influence of temperature and strain rate on the flow stress behavior of TRC+R+HT AZ31 magnesium alloy. These results are necessary for modeling and simulation of material behavior regarding to prospective applications.

The main objective of this study was to investigate the flow stress behavior according to the microstructural evolution of TRC, rolled and heat-treated AZ31 magnesium alloy in a wide range of temperatures and strain rates. The occurrence of the different deformation mechanisms was discussed.

## 2 Experimental

AZ31 (Mg-3Al-1Zn-0.3Mn, mass fraction, %) sheets with 720 mm in width and 4.5 mm in thickness were produced by twin-roll casting technology) (Fig. 1 [38]). In this process, AZ31 sheet was melted in a gas-fired furnace and passed the water-cooled vertical rotating rolls through a casting nozzle. Subsequently, the material was solidified in the roll gap. In order to dissolve existing segregations and obtain a recrystallized microstructure, the sheets were homogenized at 743 K for 2 h. After cutting to 600 mm×100 mm (length×width) and rolling to 2 mm in thickness in three roll passes at a roll speed of 5 m/s, a final heat-treatment was carried out at 603 K for 45 min. Previous investigations show that a roll speed of 5 m/s tended to enhance dynamic recrystallization, resulting in better ductility in comparison to 1 m/s [37]. Out of these sheets tensile specimens were fabricated along the rolling direction  $(RD, 0^{\circ})$  via water jet cutting. The specimen geometry is presented in Fig. 2(a). Quasi-static tensile tests were



Fig. 1 Principle of twin-roll casting process [38]



**Fig. 2** Characterization of initial material (referring to Web version of this article): (a) Tensile specimen geometry (mm); (b) Grain size distribution chart of initial material AZ31 (determined by linear intercept method); (b) Starting texture represented by  $\{0002\}$ ,  $\{10\overline{1}0\}$ , and  $\{10\overline{1}1\}$  pole figures; (c) EBSD image showing color-coded crystal orientation and grain boundaries of twin-roll cast, rolled and heat-treated AZ31 magnesium alloy (after noise reduction; step size of 1 µm)

performed in air with the electromechanical universal testing devices Zwick 1476 and Zwick Z020 with 100 kN and 20 kN loading cells, respectively. For validation of the strain rate to  $0.76 \times 10^{-3}$  s<sup>-1</sup>,  $0.73 \times 10^{-2}$  s<sup>-1</sup> and  $0.72 \times 10^{-1}$  s<sup>-1</sup>, strain gauge measurements were conducted at room temperature. In common production processes magnesium alloys are formed below 573 K. Due to this fact, the flow stress behavior under tensile loading was investigated in a temperature range of 293–573 K. Immediately after the tests, all specimens were quenched in water.

For qualitative and quantitative microstructure evaluations, the deformed gauge lengths were studied

with optical microscopy (OM), scanning electron microscopy (SEM) including electron backscatter diffraction (EBSD), and X-ray diffraction analysis (XRD) with a Seifert-FPM URD 6 in the area of uniform elongation. For this procedure, the samples were ground under water with SiC paper using 500, 800, 1200, 2400, and 4000 grit size at a pressure of 0.08 MPa for 60 s, respectively. Finally, a polishing with OP-S suspension (130 mL OP-S, 130 mL H<sub>2</sub>O, 10 mL H<sub>2</sub>O<sub>2</sub>, 1 mL 25% NH<sub>3</sub>·H<sub>2</sub>O solution) followed using 0.31 MPa pressure for 3.5 min. Etching was performed with picric acid for 15 s. EBSD measurements were conducted with a FEG SEM LEO 1530 Gemini (acceleration voltage of 20 kV) at step size of 0.2 and 1  $\mu$ m with additionally ion-beam polished samples [39]. Grain size measurements were accomplished with the linear intercept method on 500–700 grains in optical micrographs in the area of uniform elongation.

# **3** Results and discussion

#### 3.1 Material characterization

The initial sheet has a recrystallized and twin-free grain structure with an average grain size of  $7 \,\mu m$ . Figure 2(b) presents the grain size distribution of the initial material. It can be seen that the fraction of fine grains ( $\leq 2 \mu m$ ) is 9 %. It must be noted, that there is a difference in the grain size distribution between the middle section of the sheet and the sheet surface, which was also reported by SCHWARZ et al [40] for TRC AZ31 magnesium alloy. Larger amount of coarse grains  $(d>10 \,\mu\text{m})$  in the middle section of the sheet are ascribed to the lower cooling rate in comparison to the sheet surface in the TRC, rolling and heat-treatment process [41]. Figure 2(c) presents an EBSD map showing the orientation of the microstructure. This map was performed with an indexing rate of 85 %. Al<sub>8</sub>Mn<sub>5</sub> particles were detected with transmission electron microscopy (TEM), which are in agreement with the findings of CAO et al [42] and PARK et al [32]. In addition, Mg-Si-based particles were also observed by SEM. There are no known data available which explain the influence of these Mg-Si particles on the mechanical properties of AZ31 magnesium alloy [43]. As shown in Fig. 2(d), the entire AZ31 material has a strong basal texture, which means that most of the unit cells in the c-axis are orientated roughly parallel to the sheet normal direction (ND).

#### 3.2 Influence of temperature on flow stress behavior

True stress-true strain curves of the deformed AZ31 magnesium alloy at a strain rate of  $0.76 \times 10^{-3}$  s<sup>-1</sup> and various temperatures are displayed in Fig. 3. With increasing temperature, the material shows lower stresses and higher deformability. At 293 K strong hardening occurs in the early stages of deformation below the ultimate tensile strength. This is related to dislocation hardening in combination with a rising degree of twinning (Figs. 4(a) and (b). Twins act as barriers for dislocation motion on the {0002} basal planes [12]. Because of the initial basal texture and the probable contraction of the c-axis during tensile deformation,  $\{10\overline{1}1\}$  compression twins are expected. However, {1012} tensile twins were mainly observed with EBSD measurements (Fig. 5). This is also related to the c/aaxial ratio, which is higher than the ideal value [3]. This type of twins actually need an extension of the *c*-axis.

KOIKE [19] suggested that the twin type did not depend on the loading direction in relation to the HCP unit cell. Twins are rather formed by internal stresses which act in various directions [3]. Furthermore, the low CRSS of  $\{10\overline{1}2\}$  tensile twinning (2 MPa) is responsible for their formation [44]. In contrast to this,  $\{10\overline{1}1\}$  compression twins are inhibited because of their high CRSS at 76–153 MPa [45].



Fig. 3 True stress–true strain curves at strain rate of  $0.76 \times 10^{-3}$  s<sup>-1</sup> and various temperatures

As a result of DRV, which provides dislocation annihilation and rearrangement above 373 K, the hardening rate decreases significantly. Another reason for the rising deformability is the increasing dislocation motion with rising temperature. KOIKE et al [46] investigated the deformed ECAE AZ31 magnesium alloy by TEM and assumed that the occurrence of DRV is also possible at room temperature. Further examinations are needed for proof on wrought alloys. With rising temperatures, the value difference between 0.2 % yield stress ( $\sigma_{0,2}$ ) and ultimate tensile strength decreases [47]. As can be seen in Fig. 4(c), incomplete DRX takes place at 473 K. Above 498 K fully recrystallized grains are present (Figs. 4(d)-(f)). The current DRX mechanism can be identified as CDRX because of the formation of low-angle boundaries within the grains (Fig. 6).

In contrast to low temperatures (293 and 373 K), the ultimate tensile strength is not related to the beginning of the specimens necking above 473 K. Interrupted tensile tests show that there is a uniform extension of the specimens. HUANG et al [48] suggested that the stress softening is related to dynamic recrystallization. Because of the dramatic decrease in the CRSS of the nonbasal slip systems, the activation of additional slip systems takes place above 498 K. GALIYEV et al [16] implied that cross slip of  $\langle a \rangle$  dislocations on the nonbasal planes is the dominating deformation mechanism between 473 K and 523 K. Cross slip at grain boundaries with high stress concentration leads to a transformation of screw



**Fig. 4** Optical micrographs of fractured AZ31 magnesium alloy at  $0.76 \times 10^{-3}$  s<sup>-1</sup> and various temperatures in area of uniform elongation (arrows showing rolling direction): (a) 293 K; (b) 373 K; (c) 473 K; (d) 498 K; (e) 523 K; (f) 573 K



**Fig. 5** EBSD image of AZ31 magnesium alloys showing formation of  $\{10\overline{1}2\}$  tensile twins (red color) and high-angle grain boundaries (black color), grey color corresponding to Kikuchi band contrast at  $\dot{\varepsilon} = 0.76 \times 10^{-3} \text{ s}^{-1}$ , *T*=293 K and  $\varepsilon_{\text{pl}} = 20$  % (after noise reduction, step size of 0.2 µm) (referring to Web version of this article)

dislocations to edge dislocations. Because of the high stacking fault energy of the nonbasal planes, climbing and cross slip are promoted. The consequence is the formation of low-angle grain boundaries which result in the CDRX. Furthermore, the improved dislocation movement causes a dramatic reduction of twinning which results in less hardening. However, because of their low CRSS, twins are still active at 573 K and  $0.76 \times 10^{-3}$  s<sup>-1</sup> (Fig. 7). Above 523 K, bulging of grains is slightly visible, which was also found by GALIYEV et al [16] and YI et al [49]. TAN and TAN [50] suggested that the occurrence of precipitations is not inevitably necessary for DRX of AZ31 magnesium alloy. This is caused by the hexagonal structure and the lack of active slip systems. Furthermore, the low stacking fault energy and the high grain boundary diffusion rate are



**Fig. 6** Microstructures of AZ31 magnesium alloys measured by EBSD in area of uniform elongation at 473 K (a) and 573 K (b) and strain rate of  $0.76 \times 10^{-3}$  s<sup>-1</sup> (high-angle boundaries (black), low-angle boundaries (yellow), tension twins (red), compression twins (green), double twinning (pink), referring to Web version of this article)



**Fig. 7** Optical micrograph of fractured AZ31 magnesium alloy at  $0.76 \times 10^{-3}$  s<sup>-1</sup> and 573 K in area of uniform elongation (arrows showing rolling direction)

more important than the occurrence of precipitation. YI et al [49] investigated the formability as a function of the precipitation structure in the extruded AZ31 magnesium alloy. At 473 K, the deformation takes place in the two-phase area (Mg+ $\gamma$ ). In contrast to this, precipitates are not visible at 523 K because the deformation was performed in the single-phase region (Mg). This results in an enhancement of dislocation slip at 523 K because of the missing precipitate structure cannot be observed in the Al<sub>8</sub>Mn<sub>5</sub> precipitate structure cannot be observed in the current investigation. In addition, Mg<sub>17</sub>Al<sub>12</sub> precipitate is not detected with EBSD.

With decreasing the grain size, the driving force for grain boundary sliding (GBS) is increased, but the driving force for twinning is reduced [49,51]. The combination of DRV, DRX, GBS, and the activation of additional slip systems above 498 K leads to a high deformability of  $\varepsilon_{pl, 498 \text{ K}}=0.82$ ,  $\varepsilon_{pl, 523 \text{ K}}=0.95$ , and  $\varepsilon_{pl, 573 \text{ K}}=1.14$ . GBS can result in the formation of voids (Fig. 4(f), Fig. 7) because of the high stresses, which are induced at the triple points of three moving grains [52].

In combination with the formation of recrystallized grains, these stress concentrations can cause crack initiation. This phenomenon is also a function of temperature. High temperatures provide a large fraction of DRX (small grains), which activates large amount of GBS. Fine grains (d<10 µm) and high misorientations are the requirement for GBS [53]. KOIKE et al [19,54] reported that GBS is also visible under tensile loading at room temperature where recrystallization does not occur.

The grain size distribution of fractured AZ31 samples is shown in Fig. 8. In comparison to Fig. 2(c), the relative fraction of fine grains ( $\leq 2 \mu m$ ) increases from 9% (initial state) to 69% (473 K) and to 30% (498 K). At 523 and 573 K, CDRX takes place combined with large grain growth. This results in a lower fraction of fine grains and a higher fraction of coarse grains ( $\geq 4 \mu m$ ).

The measured textures (Fig. 9) show differences to the initial texture (Fig. 2(d)). At each deformation



**Fig. 8** Grain size distribution chart of AZ31 magnesium alloys (fracture) at strain rate of  $0.76 \times 10^{-3} \text{ s}^{-1}$  and different temperatures (determined by linear intercept method)



Fig. 9 Pole figures of AZ31 magnesium alloys corresponding to failure at different temperatures and strain rate of  $0.76 \times 10^{-3}$  s<sup>-1</sup>

temperature the maximum intensities of the {0002} pole figure are stretched in the transversal direction (TD) that is orthogonal to the deformation direction. This phenomenon is probably related to prismatic slip which was also reported by JAIN et al [55] and ULACIA et al [7]. At 573 K, the maximum intensity of the basal plane is reduced, which is implied by grain rotation as a consequence of GBS and CDRX [56]. As can be seen below 498 K, the orientation of  $\{10\overline{1}1\}$  pyramidal planes reflects the hexagonal symmetry and the  $\{11\overline{2}0\}$ Burgers vector is always in TD. Above 498 K, the hexagonal regularity of the {1011} planes disappears. The basal planes are mostly oriented roughly parallel to the deformation direction, resulting in a strong basal texture. Therefore, the Schmid factor of basal slip and the effective shear stress are nearly zero. Despite its low CRSS, basal slip is prevented. Combined with fine grains, which have an inhibitory effect on twinning, only insufficient deformation mechanisms are active [57]. Consequently, the activation of prismatic slip systems is possible [46], so that the rotation of the grains takes place [48,58]. This was confirmed by in-situ measurements of KESHAVARZ and BARNETT [59]. As a result of CDRX, grain growth, GBS, and activation of additional slip systems, the pyramidal planes are rotationally symmetric at 523 and 573 K as in the initial sheet material. DUDAMELL et al [60] showed that additional basal texture components can be formed by extension twinning. This cannot be confirmed at any temperature or strain rate in this investigation. The reason is that  $\{10\overline{1}2\}$  twinning is more pronounced at dynamic strain rate of 10<sup>3</sup> s<sup>-1</sup>. Larger amount of twins lead to higher degree of grain reorientation and texture modification.

#### 3.3 Influence of strain rate

The effect of strain rate on the flow stress behavior of AZ31 magnesium alloy at room temperature is presented in Fig. 10. With increasing strain rate, the flow



**Fig. 10** True stress-true strain curves of AZ31 magnesium alloys at various strain rates and 293 K

stress rises. Below plastic strains of 0.1, the hardening rate, which is the slope of the  $\sigma$ - $\varepsilon$  curve, is equal for every strain rate. Above plastic strain of 0.17, the curves intersect, which is not related to measurement errors or value spreading. This is related to the energy conversion during plastic deformation at various strain rates [61]. The result is that an increase of strain rate leads to a temperature rise  $\Delta T$  during deformation depending on plastic work ( $\int \sigma d\varepsilon$ ) as a function of strain, density  $\rho$ , specific heat capacity  $c_V$  and the factor  $\eta$  which describes the strain rate dependent degree of energy conversion. With the thermodynamic equation  $\Delta T = (\eta \int \sigma d\varepsilon)/(\rho c_V)$ , the temperature rises of 4 K ( $10^{-3}$  s<sup>-1</sup>), 8 K ( $10^{-2}$  s<sup>-1</sup>), and 13 K  $(10^{-1} \text{ s}^{-1})$  were calculated for room temperature deformation. This increase of temperature leads to softening the flow stress at higher deformation degrees as a result of reduced twinning as a consequence of enhanced dislocation motion.

For analysis of the deformation mechanisms, Fig. 11 shows the influence of strain rate on the yield stress ( $\sigma_{0,2}$ ) of AZ31 magnesium alloys at various temperatures. With increasing strain rate, the  $\sigma_{0,2}$  rises. The strain rate does not influence the CRSS of basal



**Fig. 11** Effect of strain rate on yield stress ( $\sigma_{0.2}$ ) of AZ31 magnesium alloys at various temperatures

slip [5]. However, it affects the CRSS of the nonbasal planes. Therefore, an activation of nonbasal slip systems requires different stresses at various strain rates so that the flow stress increases with increasing the strain rate. Higher strain rates also lead to larger amount of twins due to the stress concentration at the grain boundary, where dislocation accumulation occurs [60]. In contrast to  $10^{-3} \text{ s}^{-1}$ ,  $\{10\overline{1}1\}$  compression twins (green) and  $\{10\overline{1}1\} - \{10\overline{1}2\}$  double twins (pink) are additionally found at  $10^{-1} \text{ s}^{-1}$  because of the high stresses (Fig. 12).

Twins interrupt the free path length of the dislocations so that the number of piled-up dislocations at one boundary decreases at constant loading, which results in lower local stresses [12,13,62]. The smaller the stresses are, the lower the activation of neighboring grains is. According to this, fewer grains were deformed by dislocation motion on the slip systems so that the external stress for plastic deformation must rise [63]. In addition, twins reorient the parent grains which can lead

to softening or hardening. BARNETT [64] suggested that hardening as a result of grain reorientation is much more pronounced than for grain refining. Higher strain rates enhance dislocation hardening in terms of dislocation interactions and dislocation-twin interactions. Furthermore, DRV, DRX and GBS are diffusion-controlled mechanisms, and therefore time and inversely strain rate dependent [54]. High strain rates tend to enhance the dislocation generation, which induces a distinct formation of subgrains at  $10^{-1}$  s<sup>-1</sup>.

To present the influence of strain rate on the flow stress of AZ31 magnesium alloys, the flow stress at a plastic strain of 0.1 is plotted as a function of the strain rate (Fig. 13(a)). As described above, the flow stress decreases with rising temperature. The slope of the double logarithmic flow stress vs strain curve is defined as the strain rate sensitivity (*m*) and shown in Fig. 13(b). The parameter m increases with decreasing plastic strain [65]. At 293 K, the true stress is nearly independent of the strain rate. With increasing temperature, m rises, which indicates a change of the deformation mechanisms [66,67]. From 293 K (*m*=0.009) to 573 K (*m*=0.170), *m* rises by a factor of nearly 19, which is responsible for the enhanced ductility at high temperatures [67]. An average stress exponent n (n = 1/m) was calculated to be 7.87 (m = 0.127) for hot deformation between 473 K and 573 K. BARNETT [64] (486 K  $\leq T \leq$  677 K,  $10^{-3} \text{ s}^{-1} \leq \dot{\varepsilon} \leq 10^{-1} \text{ s}^{-1}$ , and MAKSOUD et al [66] (486 K  $\leq T \leq$  677 K,  $10^{-4} \text{ s}^{-1} \leq \dot{\varepsilon} \leq 10^{-2} \text{ s}^{-1}$ ) determined stress exponent to be approximately 7. ISHIKAWA et al [10] (523 K  $\leq T \leq$ 723 K,  $10^{-3}$  s<sup>-1</sup>  $\leq \dot{\varepsilon} \leq 10^{-1}$  s<sup>-1</sup>) calculated n values of 5–7 and recommended climbing-controlled dislocation creep as a possible deformation mechanism.

The influence of strain rate on the ductility is presented in Fig. 14(a). Below 323 K, the elongation to failure reaches approximately 20% and is not influenced by the strain rate as it was found in Ref. [48]. As



**Fig. 12** Microstructures measured by EBSD in area of uniform elongation at 293 K, plastic strain of 20 % and strain rates of  $0.76 \times 10^{-3} \text{ s}^{-1}$  (a) and  $0.72 \times 10^{-1} \text{ s}^{-1}$  (b) (high-angle boundaries (black), low-angle boundaries (yellow), tension twins (red), compression twins (green), double twinning (pink), referring to Web version of this article)



**Fig. 13** Relationship between  $\ln \sigma$  and  $\ln \dot{\varepsilon}$  (a) and strain rate sensitivity *m* vs temperature (b) at  $\varepsilon_{pl} = 0.1$  for AZ31 magnesium alloys



**Fig. 14** Elongation to failure vs strain rate at various temperatures (a) and deformed specimens fractured at 573 K and various strain rates (b)

described above, with increasing temperature the formability is significantly improved. Between 473 K and 523 K, it seems that the elongation to failure is less affected by temperature at  $10^{-2}$  and  $10^{-1}$  s<sup>-1</sup> [48]. Compared with the investigations of JÄGER et al [47], this phenomenon should be also a function of the grain size because it occurred already at  $10^{-4}$  s<sup>-1</sup> and a grain

size of 53 µm in their study. This has not been investigated yet, but it may be related to the dependence on strain rate for dislocation glide. The activation of nonbasal slip is more pronounced at low strain rates because of the decrease of the CRSS<sub>nonbasal</sub> at elevated temperatures [7]. Therefore, high temperatures and high strain rates are opposed factors for dislocation motion. At 573 K, the formability rises at  $10^{-3}$  s<sup>-1</sup> and  $10^{-2}$  s<sup>-1</sup>, but stagnates at  $10^{-1}$  s<sup>-1</sup>. The influence of strain rate on the ductility at 573 K can be observed on the deformed specimen (Fig. 14(b)). As noted above, the specimen failed by local necking after exhibiting a uniform elongation [68].

Below 473 K, a fracture surface at a  $45^{\circ}$  angle is visible independent of the strain rate [65]. This can be explained by the formation of twins which trigger the crack initiation. Above 293 K, an increasing ductile behavior is recognizable because of the activation of additional slip systems. Therefore, the fraction of twins is reduced, which causes a ductile fracture with distinct necking.

In comparison to Fig. 8, the grain size distribution at strain rates of  $10^{-2}$  s<sup>-1</sup> and  $10^{-1}$  s<sup>-1</sup> and different temperatures is presented in Figs. 15 and 16, respectively. At these strain rates, DRX is visible below 523 K because of larger amount of fine recrystallized grains ( $\leq 2 \mu m$ ). Nevertheless, grain growth takes place at 523 K and particularly at 573 K. Especially with rising strain rate, grain growth is increasingly inhibited.

Compared with Fig. 9  $(10^{-3} \text{ s}^{-1})$ , pole figures at different temperatures and a strain rate of  $10^{-1} \text{ s}^{-1}$  are shown in Fig. 17. The maximum intensities are independent of the strain rate [60]. In addition, the rotational symmetry of pyramidal planes is a function of strain rate because of the time dependence of DRX and the activation of additional slip systems. With increasing strain rate to  $10^{-1} \text{ s}^{-1}$ , the rotational symmetry can be

observed only at 573 K. This is related to the high thermal energy which reduces the CRSS of nonbasal slip and increases the extent of grain growth. Therefore, a reorientation of the unit cell is not necessary for plastic deformation.



**Fig. 15** Grain size distribution of fractured AZ31 magnesium alloys after deformation at different temperatures and  $0.73 \times 10^{-2} \text{ s}^{-1}$  (determined by linear intercept method)



**Fig. 16** Grain size distribution of fractured AZ31 magnesium alloys after deformation at different temperatures and  $0.72 \times 10^{-1} \text{ s}^{-1}$  (determined by linear intercept method)

#### **4** Conclusions

1) With increasing temperature and decreasing strain rate, the flow stress of TRC+R+HT AZ31 decreases and the ductility increases. This is related to the occurrence of DRV, CDRX, GBS and the enhanced reduction of the CRSS for dislocation motion. The change of the deformation mechanisms with increasing temperature is indicated by the rise of the strain rate sensitivity.

2) CDRX-induced GBS enhances the formation of voids.

3) With increasing strain rate, dislocation hardening and mechanical  $\{10\overline{1}2\}$  tensile twinning take place which cause an increase of the yield stress. Compared with  $10^{-3} \text{ s}^{-1}$ ,  $\{10\overline{1}1\}$  compression twins and  $\{10\overline{1}1\} - \{10\overline{1}2\}$  double twins are additionally found at  $10^{-1} \text{ s}^{-1}$  because of the high stresses.

4) At  $10^{-2}$  and  $10^{-1}$  s<sup>-1</sup>, the elongation to failure is nearly independent of the strain rate between 473 and 523 K. This is related to the strain rate dependence of CRSS for nonbasal slip.

5) Under each deformation condition, the maximum intensities of the {0002} pole figure are stretched in the transversal direction (TD) that is orthogonal to the deformation direction. This is probably related to the prismatic slip.

6) Below 498 K  $(10^{-3} \text{ s}^{-1})$  and 573 K  $(10^{-1} \text{ s}^{-1})$ , the orientation of  $\{10\overline{1}1\}$  pyramidal planes reflects the hexagonal symmetry which is probably related to the prismatic slip. Above this strain rate dependent transition temperature, the hexagonal regularity of the  $\{10\overline{1}1\}$  planes disappears because CDRX, grain growth, GBS, and the activation of additional slip systems take place.



Fig. 17 Pole figures of fractured AZ31 magnesium alloy at different temperatures and  $0.72 \times 10^{-1} \text{ s}^{-1}$ 

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# 温度和应变速率对双辊铸轧、轧制和 热处理态 AZ31 镁合金流动应力行为的影响

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摘 要: 在单轴拉伸条件下研究温度和应变速率对双辊铸轧、轧制和热处理 AZ31 镁合金流动应力行为的影响。 结果表明,在高温下,动态回复、连续动态再结晶、晶界滑移和附加滑移系的活化使合金的延展得到改善。在 473~523 K和 10<sup>-2</sup>~10<sup>-1</sup> s<sup>-1</sup>条件下,合金的断裂伸长率几乎与应变速率无关,而与应变速率依赖于临界剪切应的 非基面滑移有关。由于经上述工艺加工的 AZ31 镁合金具有较低的临界剪切应力,因此在 573 K和 10<sup>-3</sup> s<sup>-1</sup>条件下 出现了孪晶。

关键词: AZ31 镁合金; 双辊铸轧; 应变速率; 温度; 动态再结晶; 流动应力

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