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# Effect of temperature on mechanical properties of continuously cast AZ31 magnesium alloy

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**Abstract**

The flow behaviour of continuously cast AZ31 magnesium was investigated as a function of the temperature and the sample orientation. The microstructure evolution and the acoustic emission signal during the tensile tests were recorded. The results show a significant orientation dependence of mechanical properties. The degradation of mechanical properties with the increasing temperature is discussed in terms of dislocation structure evolution.

Key words: magnesium alloys, acoustic methods, tensile test

## 1. Introduction

It is well-known that magnesium alloys exhibit a relatively high specific strength and good specific stiffness resulting from their low density [1]. These suitable mechanical properties estimated at room temperature are not observed at elevated temperatures. Magnesium alloys possess poor mechanical properties at temperatures above 150 °C [2]. The strength of cast magnesium alloys is strongly decreased by increasing temperature. At higher temperatures, the deformation behaviour of magnesium alloys is influenced by dynamic recovery and/or dynamic recrystallization [3]. However, the strength is significantly dependent on one or more of the well-known hardening (strengthening) mechanisms. The mechanisms may change the temperature dependence of the mechanical properties and deformation behaviour of an alloy [4]. Magnesium alloys have attracted significant interest, due to their specific properties, for a number of structural applications in different branches. Considerable attention has focused on the effect of the testing temperature on the flow properties, which are important for both applications and estimation of forming conditions. In order to explain the flow behaviour of a magnesium alloy, it is important not only to estimate experimental values (yield strength, tensile strength, ductility) but also to understand the main mechanisms of the deformation process.

The aim of the present paper is to estimate the temperature dependence of the flow behaviour of continuously cast AZ31 magnesium alloy and to describe deformation mechanisms that control the strain hardening rate at different temperatures.

## 2. Experimental

Magnesium alloy AZ31 (nominal composition Mg-3wt.%Al-1wt.%Zn) investigated in this work was prepared by continuous casting. Samples for tensile tests having a rectangular cross section of $5 \times 5\ \mathrm{mm}^2$ and a gauge length of 25 mm were deformed using an INSTRON tensile machine at a constant cross head speed giving an initial strain rate of $10^{-3}\ \mathrm{s}^{-1}$. Tensile tests were carried out at various temperatures in a temperature range from 20 to 300 °C. Two types of sheet specimens were used: specimens with load axis parallel (longitudinal direction, hereafter LD samples) and perpendicular (transversal direction, hereafter TD samples) to the casting direction. The temperature in furnace was controlled to within ± 2 °C. The grain size was estimated as 620 μm [5]. During straining at room temperature, acoustic emission (AE) was monitored

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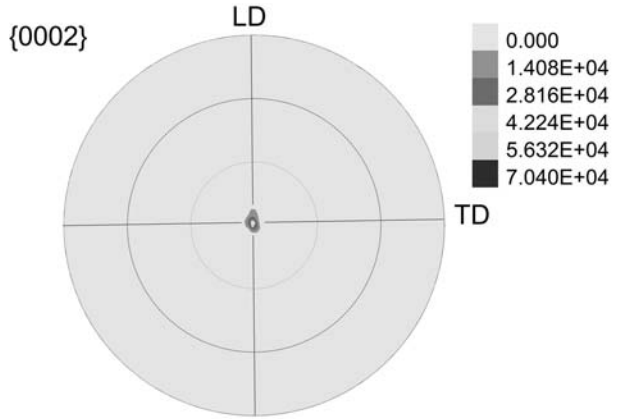


Fig. 1. Initial texture of the specimen represented by (0002) pole figure.

directly using a computer-controlled DAKEL-XEDO--3 facility (DAKEL-ZD Rpety, Czech Republic). The facility incorporated a high temperature S9215 transducer (Physical Acoustic Corporation) with a flat response between 50 and 650 kHz and a built-in preamplifier giving a gain of ∼ 30 dB. The total gain was about 94 dB. The AE analyser detects the AE signals at two settable threshold levels, which corresponded to voltages of 730 mV for the total AE count 1 and 1450 mV for the burst AE count 2 (total range of the A/D converter is from 0 to 2400 mV). The specimens for light microscopy were mechanically polished and finally etched (90 s) in 3 % nital solution.

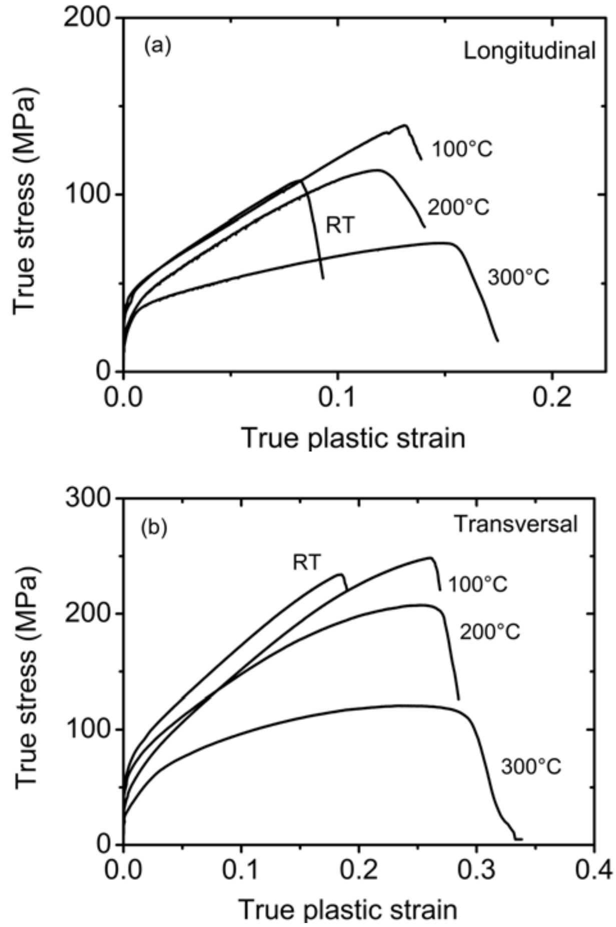


Fig. 2. Temperature dependence of true stress-true strain curves for AZ31 sample. The load direction is parallel to a) longitudinal direction, b) transversal direction.

## 3. Results and discussion

### *3.1. The effect of the orientations*

The initial microstructure exhibits a basal texture (cf. Fig. 1). The basal planes (0002) are nearly parallel to the plane of the sheet (the casting direction lies in the sheet plane). Figures 2a,b show the true stress-true strain curves obtained at various temperatures for longitudinal and transversal samples, respectively. It can be seen that the TD samples exhibit a higher flow stress. Furthermore, the testing temperature influences significantly the shape of the true stress-true strain curves. However, the temperature influence of the flow stress is very similar. Figure 3 shows the true stress-true strain curves obtained at room temperature for samples of both orientations. The flow stress of LD sample is lower than that of TD sample – the maximum stress of the TD sample is about two times higher than that of LD sample. The elongation ($\varepsilon \approx 0.1$) to failure of LD sample is lower than that ($\varepsilon \approx 0.2$) of TD sample. In all cases, higher strain hardening rate is found in the TD sample than in the LD one. A difference in the

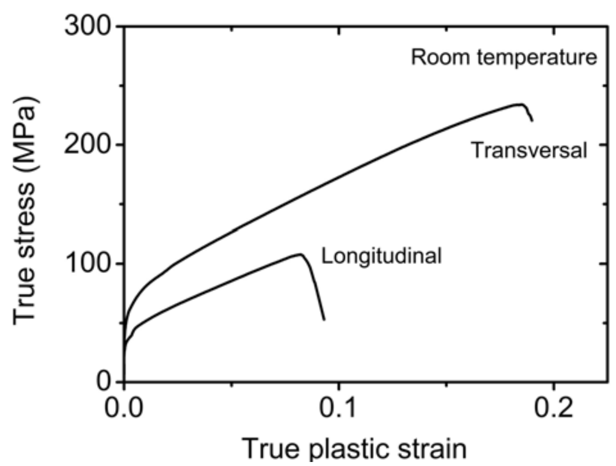


Fig. 3. Orientation dependence of deformation curves at room temperature.

flow stresses of both sample orientations could be explained in terms of different deformation behaviour of textured samples with LD and TD orientation. In

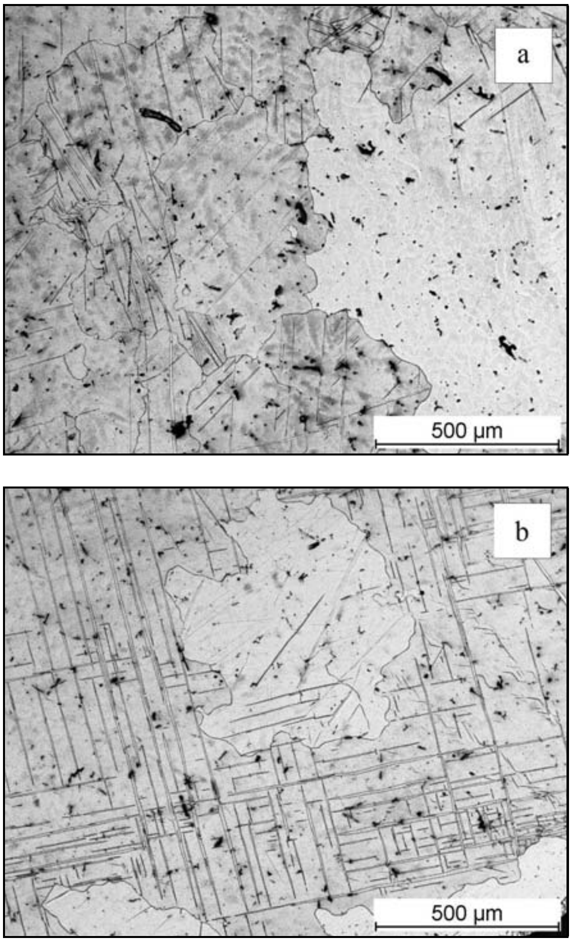


Fig. 4. Micrographs of AZ31 specimens deformed at room temperature. The load direction is parallel to a) longitudinal direction, b) transversal direction.

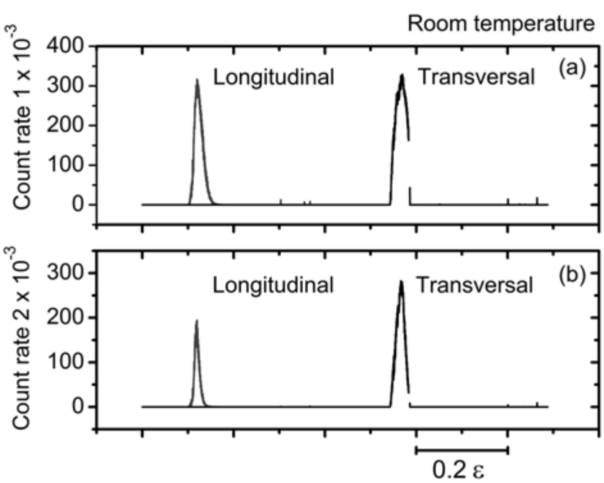


Fig. 5. Orientation dependence of count rate 1 (a) and 2 (b) for samples deformed at room temperature.

the initial state, the basal plane (0001) is nearly parallel to the plane of the sheet. Consequently, slip of basal <a> dislocations is the dominant mechanism for samples with the LD orientation. Since the slip in the basal plane itself does not fulfill the von Mises criterion for plastic deformation of polycrystals [6], $\{10\bar{1}2\}$ twinning takes place as well [7]. The activation of this twinning system causes misorientation of 86.3° between the twinned and untwinned regions. On the other hand, twinning can stimulate dislocation glide because it changes locally the lattice orientation, which can become more favourably oriented for the basal slip [8]. As it has been shown by Yoo [9], the twin boundaries are impenetrable for basal dislocations. Therefore twinning on planes $\{10\bar{1}2\}$ lowers the spacing between the non-dislocation type obstacles causing hardening. In TD direction, where basal slip is not favourable, the twinning activity is more pronounced. It can be seen in the micrographs of Fig. 4, where the microstructures of as-deformed LD and TD samples are shown that the density of twins is much higher for the TD sample. Since in the initial state the grains were elongated in the longitudinal direction [5], influence of grain size on mechanical properties could not be excluded. In TD direction, the stress concentrations developed by dislocation pile-ups at grain boundaries are higher in comparison to the LD direction. Thus, additional strengthening effect could be arisen [10]. Furthermore, the stress concentrations developed by dislocation pile-ups at grain boundaries initiate new slip processes. The shear stress necessary for the activity of a non-basal slip system is lower than the critical resolved shear stress (CRSS) for the system. Therefore, the "effective" CRSS for prismatic slip is lower than the values measured in magnesium single crystals [11]. It means cross slip of large group of screw dislocations may occur. This leads to an increase in the free path of dislocations resulting in higher value of the elongation to failure. The role of twinning in deformation mechanisms is supported also by results of the acoustic emission measurements. In our AE experiments two threshold levels were set. As the first threshold level was set directly above the peak values of the thermal noise, the total AE count (Count rate 1) was assumed as a sum of the response of all deformation mechanisms detectable by AE. The burst AE count (Count rate 2), measured at the second threshold level, was only used for gathering signals coming from strong effects (e.g. twinning). Figure 5 demonstrates the strain dependences of count rates 1 and 2 for both orientations of the samples. There is no difference in the count rate 1 but the value of the count rate 2 is significantly higher for samples with TD direction, which indicates more intensive twinning. Consequently, TD samples have a higher amount of twin boundaries – the free path of dislocations decreases, which leads to a higher hardening rate observed in the experiment. Dislocation pile-ups are formed at the twin boundaries. The stress concentrations at the head of pile-ups contribute to initiations of the activity of the non-basal slip systems [12]. From activities of non-basal slip systems, motion of dislocations with the <c + a> Burgers vector in the second-order pyramidal slip systems is expected. Then the interaction between non-basal and basal dislocations results either to sessile dislocations (obstacles for moving dislocations) leading to hardening or to glissile dislocations that may annihilate leading to softening [13].

### *3.2. The effect of the testing temperature*

The deformation curves for the LD and TD samples behave similarly with increasing temperature. The influence of the testing temperature on the shape of the true stress-true strain curves of continuously cast AZ31 alloy is similar as observed in the case of cast AZ31 alloy [14]. Therefore, we can suppose that the decrease of the internal stress causes a decrease both in the strength and in the strain hardening with increasing temperature. At temperatures below about 200 ℃ a strong hardening is observed. On the other hand, a decrease in the strain hardening above about 200 ℃ indicates significant activity of recovery processes. From the macroscopic point of view the decrease in the strain hardening rate at higher temperatures indicates a dynamic balance between hardening and softening [4]. From the dislocation theory point of view, the observed deformation behaviour may be accounted for assuming a change in deformation micromechanisms. At temperatures below 200 ℃, the strain hardening is caused by multiplication and storage of dislocations [15]. At temperatures above 200 ℃, there is not only storage of dislocation leading to hardening but also annihilation of dislocations (decrease in the dislocation density) leading to softening. At the higher temperatures, the activity of non-basal slip systems is increasing. It has been shown, (using X-ray diffraction), that the density of non-basal dislocations increases with increasing testing temperature [16]. We hence assume the activity of the second-order pyramidal slip systems – the glide of <c + a> dislocations. The motion of <c + a> dislocations has to play an important role in both hardening and recovery processes.

Different reactions between <a> basal dislocations and <c + a> pyramidal dislocations can occur [13, 17]. Glissile (glide) <c + a> dislocations can interact with <a> dislocations, then immobile <c> dislocations may arise within the basal plane according to the following reaction:

$$\frac{1}{3}\langle 2\bar{1}\bar{1}3\rangle + \frac{1}{3}\langle \bar{2}110\rangle \rightarrow \langle 0001\rangle . \qquad (1)$$

Another reaction may yield a sessile <c + a> dislocation:

$$\frac{1}{3}\langle 2\bar{1}\bar{1}3\rangle + \frac{1}{3}\langle \bar{1}2\bar{1}0\rangle \rightarrow \frac{1}{3}\langle 11\bar{2}3\rangle . \qquad (2)$$

A combination of two glissile <c + a> dislocations gives rise to a sessile dislocation of <a> type that lies along the intersection of the second-order pyramidal planes according to the following reaction:

$$\frac{1}{3}\langle 2\bar{1}\bar{1}3\rangle + \frac{1}{3}\langle \bar{1}2\bar{1}\bar{3}\rangle \rightarrow \frac{1}{3}\langle 11\bar{2}0\rangle . \qquad (3)$$

It can be seen that different dislocation reactions may produce both sessile and glissile dislocations. Production of sessile dislocations increases the density of the forest dislocations that are obstacles for moving dislocations. Therefore, an increase in the flow stress with straining is caused by an increase of obstacles, which is observed in the experiment. On the other hand, screw components of <c + a> (and also <a>) dislocation may move to the parallel slip planes by double cross slip and after cross slip they can annihilate – the dislocation density decreases, which leads to softening. It is clear that the intensity of cross slip of dislocations increases with an increase in temperature. Máthis et al. [18] have reported that an analysis of the stress dependence of the strain hardening rate gives information on an increase of cross slip in AM60 alloy with increasing deformation temperature. The influence of the texture on the mechanical properties sustains at high temperatures as well. Due to the above mentioned high stress concentration at grain boundaries of the TD specimens, the probability of the double-cross slip mechanism is also higher. Thus, a higher strain to failure is observed for TD specimen at 300 ℃.

## 4. Conclusions

The flow behaviour of continuously cast AZ31 magnesium alloy was investigated as a function of the testing temperature and the sample orientation. The following conclusions can be drawn:

(i) The twinning activity of the RD samples is higher due to their unfavourable orientation for basal slip.

(ii) The twin boundaries cause pile-ups of basal dislocations. The stress field of pile-ups facilitates the activation of non-basal slip systems.

(iii) The softening process at higher temperatures is governed by cross-slip of basal and pyramidal dislocations.

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