TEMPERATURE DEPENDENCE OF TENSILE PROPERTIES OF AZ31 Mg ALLOY SHEETS DEFORMED AT A HIGHER STRAIN RATE

ALEŠ JÄGER, PAVEL LUKÁČ*, VIERA GÄRTNEROVÁ

Tensile properties of commercial wrought AZ31B (Mg-3Al-0.9Zn-0.15Mn in wt.%) magnesium alloy sheets were investigated at various temperatures between room temperature and 400 ℃ at a strain rate of $4.5 \times 10^{-2}\,\mathrm{s}^{-1}$. The yield stress and the maximum stress decrease with increasing temperature. The strain to fracture is constant up to 100 ℃ and increases very rapidly above this temperature to reach approx. 100 % at 400 ℃. Yield point was observed at temperature range of 250–400 ℃, which indicates the interaction of alloying elements with dislocations. The results are explained by the dislocation glide in the basal and non-basal slip systems. Possible softening mechanisms are discussed.

Key words: magnesium alloy, mechanical properties, tension test, high temperature deformation, dislocations

TEPLOTNÍ ZÁVISLOST MECHANICKÝCH VLASTNOSTÍ VÁLCOVANÉ HOŘČÍKOVÉ SLITINY AZ31 DEFORMOVANÉ V TAHU PŘI VYŠŠÍ POČÁTEČNÍ RYCHLOSTI

Mechanické vlastnosti komerční válcované hořčíkové slitiny AZ31B (Mg-3Al-0,9Zn-0,15Mn v hm.%) byly studovány v tahu v intervalu teplot 20–400 ℃ při počáteční rychlosti deformace $4{,}5 \cdot 10^{-2}\,\mathrm{s}^{-1}$. Napětí na mezi kluzu a maximální napětí klesá s rostoucí teplotou. Deformace do lomu je konstantní až do 100 ℃ a velmi rychle roste nad tuto teplotu; dosahuje 100 % při 400 ℃. V teplotním intervalu 250–400 ℃ byla pozorována ostrá mez kluzu, indikující interakci příměsových atomů s dislokacemi. Výsledky jsou interpretovány aktivitou bazálního a nebazálního kluzu. Možné odpevňující mechanismy jsou diskutovány.

Department of Metal Physics, Charles University, Ke Karlovu 5, 121 16 Prague 2, Czech Republic

* corresponding author, e-mail: lukac@met.mff.cuni.cz

1. Introduction

Magnesium alloys exhibit high specific strength at ambient temperature. Therefore they are considered as promising light structural materials. On the other hand, magnesium alloys have poor ductility and poor cold workability. When deformed at room temperature, they exhibit a tensile elongation of only a few percent. An increase in temperature may cause an increase in elongation. The poor ductility of Mg alloys may be due to difficulty of the dislocation activity in the non-basal slip systems. According to the von Mises criterion [1], five independent slip systems are required for homogeneous deformation of polycrystalline materials. The basal slip system, dominant slip system for Mg at ambient temperature, provides only two independent slip systems, which is not sufficient for satisfying the von Mises criterion [1]. The activity of non-basal slip systems is required. For commercial cast Mg alloy, an elongation to fracture is low, in most cases less than 10 % and increases with temperature up to about 30 % at 300 °C [2]. In a previous paper [3], we investigated tensile properties of AZ31 sheets deformed at a strain rate of 1.3×10^{-4} s^{-1}. We found that elongation to failure at room temperature was 22 % and it increased with increasing temperature, reaching 420 % at 400 °C indicating superplastic-like deformation. Some achievements of superplasticity of Mg alloys have been reported, e.g. in [4, 5]. Very recently Wei et al. [6] have reported that an excellent superplasticity with the maximum elongation to failure of 455 % was obtained at 350 °C in the rolled AZ91 magnesium alloys. Superplasticity is defined formally as the ability of a polycrystalline material to exhibit very high tensile elongations to failure [7, 8]. Industrial superplastic forming operations require high strain rate superplasticity, i.e. the occurrence of superplastic elongations at strain rates at and above 10^{-2} s^{-1} [8, 9].

The aim of the present work is to study the influence of temperature on tensile properties of AZ31 Mg alloy sheets deformed at 4.5×10^{-2} s^{-1} in the temperature range of 20 to 400 °C and to compare the results with those obtained at 1.3×10^{-4} s^{-1}. The results could help to estimate whether superplastic deformation of AZ31 sheets at higher strain rates is possible.

2. Experimental procedure

Commercial magnesium alloy AZ31B sheets (with nominal composition Mg-3Al-0.9Zn-0.15Mn in wt.%) in the stress-relieved (H24) temper were studied in the present work. Tensile specimen with 25 mm in gauge length, 5 mm width and 1.6 mm thickness were machined. Tensile tests were performed in an Instron tensile machine at an initial strain rate of 4.5×10^{-2} s^{-1} at various temperatures between room temperature and 400 °C. Before testing, each specimen was held at the test temperature for 30 min. The tensile direction was parallel to the extrusion direction. In previous papers [10, 11], it has been shown that the yield stress of AZ31B sheet depends on the orientation of the sheet. The yield stress is lower for the specimens that were deformed in the rolling direction than for those stressed in the transversal direction. Other mechanical properties (maximum stress $\sigma_{\max}$ and ductility) exhibited no significant dependence on the direction of the applied stress [10, 11]. Metallographic specimens were cut from the sheets, mechanically ground on progressively finer grades of SiC impregnated paper and then mechanically polished. Microstructural features were examined in an optical microscope Olympus IX 70. Specimens were etched in a solution: 5 ml acetic acid, 6 g picric acid, 100 ml ethanol and 10 ml H_2O.

3. Experimental results

The microstructure of as received sheets is shown in Fig. 1. The microstructure was investigated in detail in [11, 12]. The observations revealed inhomogeneous grain size distribution with the mean grain size of about 40 μm ($1.74 \times \overline{d}$, where $\overline{d}$ is mean linear intercept). Twins are present in many grains.

Figure 2 shows the true stress-true strain curves obtained at various temperatures. It can be seen that the flow stress decreases with increasing test temperature. At temperatures 350–400 ℃ the flow stress is practically independent of strain; no significant work hardening occurs in any specimen. The results indicate that softening can take place. At room temperature, the values of σ_{02} determined as the flow stress at 0.2 % offset strain and $\sigma_{\max}$ determined as the maximum flow stress are 256 and 330 MPa, respectively in comparison with 230 and 343 MPa, respectively, determined at a strain rate of 1.3×10^{-4} s^{-1} [3]. Insert in Fig. 2 shows detail of the transition from elastic to plastic deformation characterized by clear yield point observed in the temperature range of 250–400 ℃.

Figure 3 shows the temperature dependence of σ_{02} and $\sigma_{\max}$. The values of σ_{02} and $\sigma_{\max}$ decrease very rapidly with increasing temperature for temperatures above about 100 ℃. The values σ_{02} and $\sigma_{\max}$ are practically the same at 350 and

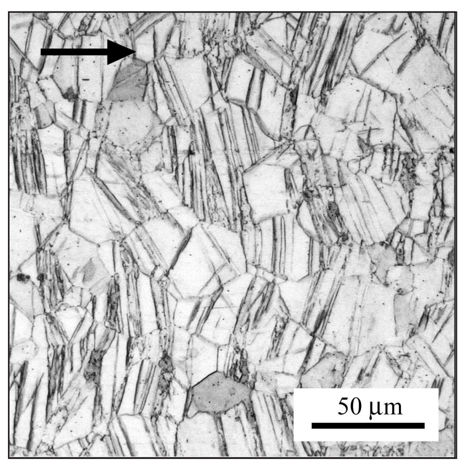

Fig. 1. Typical microstructure of as received alloy. A lot of twins as a result of rolling are clearly visible. Arrow indicates rolling direction.

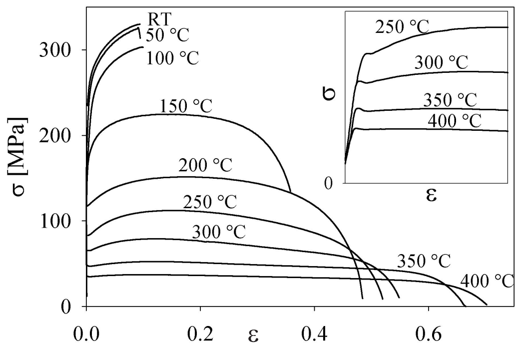

Fig. 2. Typical true stress-true strain curves for the hot rolled AZ31 alloy sheets at various temperatures and at a constant initial strain rate of 4.5×10^{-2} s^{-1}. The insert shows detail from the beginning of deformation in the temperature range of 250–400 ℃.

400 ℃, which corresponds to nonsignificant work hardening identified by a nearly constant flow stress. It should be noted that both characteristic stresses of specimens deformed at 1.3×10^{-4} s^{-1} have the same values at and above 250 ℃. The yield stress decreases from 256 MPa at room temperature to 47 MPa and 35 MPa at 350℃ and 400 ℃, respectively. A reduction of the maximum stress is from 330 MPa to 52 MPa and 37 MPa at 350℃ and 400 ℃, respectively.

Comparing the stress-strain curves in Fig. 2, it is evident that the ductility, in general, increases with increasing test temperature. Figure 4 shows the temperature dependence of the elongation to fracture A. The elongation at room temperature is 10 %, which is lower than that of 22 % obtained for specimens deformed at 1.3×10^{-4} s^{-1}. An increase of the elongation to fracture occurs at temperatures above 100 ℃. The ductility at 350 ℃ and 400 ℃ is 90 % and 100 %, respectively. The values are significantly lower than those of 294 % and 421 % determined at 350 ℃ and 400 ℃, respectively, for the specimens deformed at 1.3×10^{-4} s^{-1}. The ductility values indicate that one cannot expect high strain rate superplasticity of AZ31 sheets under conditions used in this work. Either the grains should be finer, the test temperature has to be higher or previous different thermomechanical processing is necessary. In this connection it is interesting to note that the recent investigation by Wu and Liu [13] indicates that hot rolled AZ31 magnesium alloy with coarse grains (about 300 μm) exhibits superplasticity at 500 ℃; an elongation of 320 % was obtained at 500 ℃ and at a strain rate of 1×10^{-3} s^{-1}. When deformed at 400 ℃, an elongation of 120 % was observed [13]. Further experiments are needed to explain superplasticity in AZ31 sheets.

4. Discussion

Changes in the shape of the stress-strain curves at temperatures at and above 150 ℃ can be attributed to the activity of some softening processes. The work

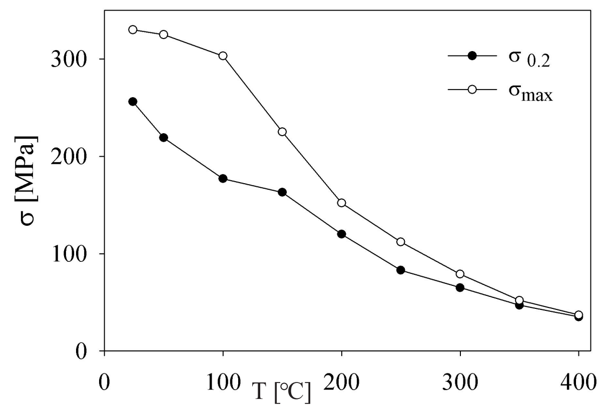

Fig. 3. Temperature dependence of the yield stress (full circles) and the maximum stress (empty circles).

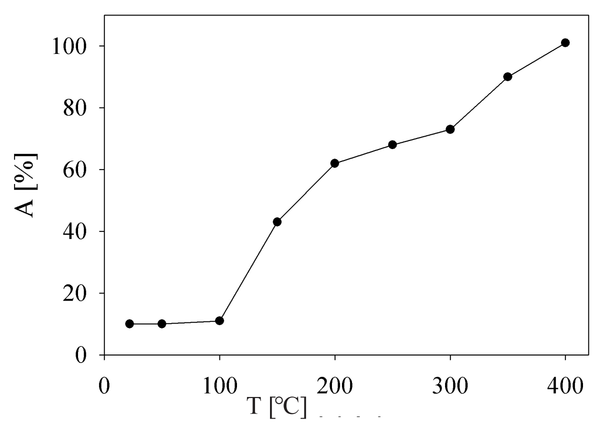

Fig. 4. The variation of elongation to fracture as a function of temperature. The line serves as a guide for eyes.

hardening rate of specimens deformed above 300 ℃ is very close to zero. Such a steady-state deformation occurring at higher temperatures may be a dynamic balance between hardening and softening processes. In another words, there is a dynamic balance between storage of dislocations leading to hardening and annihilation of dislocations leading to softening. In magnesium alloys, the activity of non-basal slip systems, which may cause softening, can occur at about 200 ℃. The activity may be influenced by the orientation of grains in respect to the loading direction, mutual orientation of the nearest grains, alloying elements, the grain size and previous processing. The stress concentration at grain boundary caused e.g.

by accumulation of dislocation in initial stage of deformation can contribute to the stress necessary to activate non-basal slip even at room temperature.

As found by Bohlen et al. [12], the AZ31 alloy sheets exhibit a non-random texture with a scattering of basal planes that is wider in the rolling direction than in the transversal direction. Most grains have basal planes nearly parallel with tensile axis, so that c-axis of hexagonal structure is nearly perpendicular to the plane of sheet. Bohlen et al. [10], using acoustic emission measurements during plastic deformation of the AZ31 alloy sheets, have deduced that deformation twinning and dislocation glide play an important role in controlling plastic deformation of the AZ31 alloy at room temperature. It is known that deformation twinning in hexagonal metals may stimulate dislocation glide because it changes the lattice orientation, which may become more favorably oriented for the basal slip. On the other hand, twin boundary is impenetrable for moving dislocation. The activity of non-basal slip systems may result in twinning, primarily on $(10\bar{1}\bar{2})$ planes. The intensity of deformation twinning depends on the grain orientation and on the testing mode (tension, compression).

As mentioned above, the observed work hardening rate is a sum of hardening and softening mechanisms. The contribution of the mechanisms to work hardening rate depends sensitively on temperature. Dynamic recovery processes cause softening during deformation. The activity of non-basal slip systems plays an important role in dynamic recovery. During deformation of the AZ31 alloy polycrystalline sheets, the motion of not only $\boldsymbol{a}$ (basal) dislocations but also $\boldsymbol{c}+\boldsymbol{a}$ (pyramidal) dislocations is assumed. When a slip system with $\boldsymbol{c}+\boldsymbol{a}$ Burgers vector is active, the von Mises [1] criterion is obeyed. Different reactions between $\boldsymbol{a}$ basal and $\boldsymbol{c}+\boldsymbol{a}$ pyramidal dislocations can occur [14–16]. Dislocation reactions between the $\boldsymbol{a}$ dislocations and the $\boldsymbol{c}+\boldsymbol{a}$ dislocations may produce obstacles of the dislocation type (an increase in the density of forest dislocations), which cause an increase in hardening. Finally, an interaction among the pyramidal $\boldsymbol{c}+\boldsymbol{a}$ dislocations may result in annihilation of dislocations, which causes the softening. Edge dislocations of $\boldsymbol{c}+\boldsymbol{a}$ type can decompose into single $\boldsymbol{a}$ and $\boldsymbol{c}$ dislocations. Because $\boldsymbol{c}$ dislocations cannot move, $\boldsymbol{c}+\boldsymbol{a}$ edge dislocations become immobile. It means storage of the dislocations occurs and, therefore, the flow stress increases. On the contrary, screw dislocations of $\boldsymbol{c}+\boldsymbol{a}$ type can move to the parallel slip planes by double cross slip and then annihilate, which causes a decrease in the work hardening rate. Local cross slip and/or dislocation climb can be taken into account as the mechanisms responsible for softening. Koike et al. [17] have shown that a major deformation mechanism from all non-basal slip systems in AZ31 alloy is glide of $\boldsymbol{c}+\boldsymbol{a}$ dislocations in pyramidal slip system. Their investigations are supporting evidence for the variation of the work hardening with the temperature.

An increase in the non-basal slip systems activity with increasing temperature may lead to annihilation of dislocations. The result is an increase in the ductility of the AZ31 alloy sheets, which is experimentally observed (Fig. 4).

The yield point in the temperature range of 250–400 ℃ (Fig. 2) ((0.56–0.72) T_m, where T_m is the melting temperature) can be explained by diffusion of alloying elements to dislocations. The force (stress) necessary for break away of the dislocations from pinning points is higher than the force (stress) necessary for the movement of the dislocations freed from their atmospheres of foreign atoms. At lower temperatures, one can assume that diffusion along high diffusivity paths like along dislocations and grain boundaries will be preferred. At higher temperatures lattice diffusion can take place more easily.

Additionally, grain boundary sliding (GBS) cannot be excluded. According to Koike et al. [18], GBS can play role even in room temperature deformation of rolled AZ31 alloy.

5. Conclusions

The work hardening behaviour of AZ31 magnesium alloy sheets was investigated between room temperature and 400 ℃ at a strain rate of 4.5×10^{-2} s^{-1} and compared with that obtained for specimens deformed at 1.3×10^{-4} s^{-1} published previously [3]. The yield stress and the maximum flow stress decrease with increasing temperature; from 256 MPa at room temperature to about 47 MPa at 350 ℃ for the yield stress and from 330 to about 52 MPa at 350 ℃ for the maximum stress. The ductility of the alloy increases with temperature above 100 ℃, reaching about 100 % at 400 ℃. The work hardening rate also decreases with increasing temperature; it becomes very close to zero at temperatures between 350 and 400 ℃, which can be attributed to an increase in the activity of non-basal slip systems. The yield point in the temperature range of 250–400 ℃ was observed as a result of interaction of moving dislocation with alloying elements.

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REFERENCES

[1] VON MISES, R.: Z. Angew. Math. Mech., *8*, 1928, p. 161.
[2] ASM Specialty Handbook Magnesium and Magnesium Alloys. Eds.: Avedesian, M. M., Baker, B. Materials Park, OH, ASM International 1999.
[3] JÄGER, A.—LUKÁČ, P.—GÄRTNEROVÁ, V.—BOHLEN, J.—KAINER, K. U.: J. Alloys Comp. (in press).
[4] MABUCHI, M.—HIGASHI, K.: Mater. Trans., *40*, 1999, p. 787.
[5] MUKAI, T.—WATANABE, H.—HIGASHI, K.: Mater. Sci. Tech., *16*, 2000, p. 1314.
[6] WEI, Y. H.—WANG, Q. D.—ZHU, Y. P.—ZHOU, H. T.—DING, W. J.—CHINO, Y.—MABUCHI, M.: Mater. Sci. Eng., *A360*, 2003, p. 107.
[7] LANGDON, T. G.: Metall. Trans., *13A*, 1982, p. 689.

[8] XU, C.—FURUKAWA, M.—HORITA, Z.—LANGDON, T. G.: Acta Mater., *51*, 2003, p. 6139.
[9] HIGASHI, K.—MABUCHI, M.—LANGDON, T. G.: ISIJ Int., *36*, 1996, p. 1423.
[10] BOHLEN, J.—CHMELÍK, F.—KAISER, F.—LETZIG, D.—LUKÁČ, P.—KAINER, K. U.: Kovove Mater., *40*, 2002, p. 290.
[11] KAISER, F.—LETZIG, D.—BOHLEN, J.—STYCZYNSKI, A.—HARTIG, CH.—KAINER, K. U.: Mater. Sci. Forum, *419–422*, 2003, p. 315.
[12] BOHLEN, J.—HORTSMANN, A.—KAISER, F.—STYCZYNSKI, A.—LETZIG, D.—KAINER, K. U.: In: Magnesium Technology 2003. Ed.: Kaplan, H. J. Warrendale, PA, TMS 2003, p. 253.
[13] WU, X.—LIU, Y.: Scripta Mater., *46*, 2002, p. 269.
[14] LUKÁČ, P: Czech. J. Phys., *B31*, 1981, p. 135.
[15] LUKÁČ, P.: Czech. J. Phys., *B35*, 1985, p. 275.
[16] LUKÁČ, P.—MÁTHIS, K.: Kovove Mater., *40*, 2002, p. 281.
[17] KOIKE, J.—KOBAYASHI, T.—MUKAI, T.—WATANABE, H.—SUZUKI, M.—MARUYAMA, K.—HIGASHI, K.: Acta Mater., *51*, 2003, p. 2055.
[18] KOIKE, J.—OHYAMA, R.—KOBAYASHI, T.—SUZUKI, M.—MARUYAMA, K.: Mater. Trans., *44*, 2003, p. 445.

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