APPLICATION OF A MODEL FOR THE WORK HARDENING BEHAVIOUR TO AN AM20 ALLOY

PAVEL LUKÁČ[1]*, KRISTIÁN MÁTHIS[2]

A new model describing the dislocation density evolution in alloys with hexagonal structure is proposed. Forest dislocations and non-dislocation obstacles are considered as obstacles for primary dislocations. Annihilation of dislocations due to recovery processes is assumed. The stress dependence of the work hardening rate is derived. Experimental results obtained for magnesium alloy AM20 (Mg-2wt.%Al-0.1wt.%Mn) deformed at various temperatures are analysed. The proposed model describes satisfactorily the shape of the stress dependence of the work hardening rate.

Key words: Mg alloy, work hardening behaviour, dislocation density evolution

POUŽITÍ MODELU DEFORMAČNÍHO ZPEVNĚNÍ NA POPIS DEFORMACE SLITINY AM20

Je navržen nový model popisující vývoj hustoty dislokací ve slitinách s hexagonální strukturou. Jako překážky pro pohyb dislokací uvažujeme dislokační les a nedislokační překážky. Předpokládáme, že zotavovací procesy způsobí anihilaci dislokací. Je odvozená napěťová závislost koeficientu zpevnění. Experimentální výsledky získané při deformaci hořčíkové slitiny AM20 (Mg-2hm.%Al-0.1hm.%Mn) za různých teplot jsou podrobeny analýze. Navržený model dostatečně dobře popisuje tvar napěťové závislosti koeficientu zpevnění.

1. Introduction

Magnesium alloys, as the lightest structural materials, are very attractive in large amount of applications. Mg-based alloys exhibit high specific strength (the ratio of the yield stress to density), superior damping capacity and high thermal conductivity. On the other hand, they possess low cold forming capability, low creep resistance and low corrosion resistance.

[1] Department of Metal Physics, Charles University, Ke Karlovu 5, 121 16 Praha 2, Czech Republic

[2] Department of General Physics, Eötvös University, H-1117 Budapest, Pazmany Peter setany 1/A, Hungary

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

In order to improve plasticity of Mg-based alloys it is important to understand their deformation behaviour. Whereas the deformation mechanisms responsible for yielding have been studied from both experimental and theoretical point of view, there are limited investigations of the work hardening mechanisms. In comparison to fcc metals, many fundamental problems remain open in the case of hcp metals. The deformation behaviour of Mg alloys has not been investigated in detail in the view of the evolution of dislocation structure. In crystalline materials the flow stress σ depends on the dislocation structure that may be represented by the average dislocation density ρ for which holds $\sigma = \alpha G b \rho^{1/2}$ where α is a numerical constant, G is the shear modulus and b is the magnitude of the Burgers vector. Very recently, Máthis et al. [1] have reported an analysis of the deformation behaviour of three commercial magnesium alloys AZ91, AS21 and AE42. The stress dependence of the work hardening rate (coefficient) of the alloys was evaluated and compared with predictions of models of Malygin [2] and Lukáč and Balík [3]. Description of the deformation behaviour of fcc alloys, especially Al-based alloys, by both models is satisfactory [4]. On the other hand, both these models do not describe satisfactorily all the experimental curves of Mg alloys. Malygin [2] has assumed that the mean free path of dislocations is determined by forest dislocations, non--dislocation (impenetrable) obstacles and process of dislocation annihilation due to cross slip. Lukáč and Balík [3] assumed dislocation climb to be an additional recovery process. In magnesium commercial alloys the presence of non-dislocation obstacles, such as precipitates and grain boundaries, should be considered. Forest dislocations are also obstacles for dislocation motion in hexagonal alloys [5, 6]. In some cases dislocations can cross slip in magnesium alloys [6] and local climb of dislocations due to pipe diffusion of vacancies cannot be excluded.

The true stress-true strain curves of hexagonal polycrystals are very similar to those for polycrystals with fcc structure. But there are differences in the activity of slip systems. According to von Mises [7] the activity of five independent slip systems is required for plastic deformation of polycrystals. In fcc metals there are five independent crystallographically equivalent slip systems. In contrast, hcp metals do not possess five independent crystallographically equivalent slip systems. The directions for easy slip in hcp single crystals are three $\boldsymbol{a}$ type directions. The directions lie in the basal plane and in three prismatic planes. In hexagonal single crystal, crystallographic slip is commonly observed to occur on basal slip or prismatic slip systems. In magnesium the main slip system is basal plane (0001) and three close packed directions:

$$\boldsymbol{a}_1 = \frac{1}{3}[\bar{1}\bar{1}20],\ \boldsymbol{a}_2 = \frac{1}{3}[2\bar{1}\bar{1}0] \text{ and } \boldsymbol{a}_3 = \frac{1}{3}[\bar{1}2\bar{1}0].$$

These vectors are perpendicular to c axis ($\boldsymbol{c}$ direction) and slip in $\boldsymbol{a}$ directions cannot produce strain parallel to the c axis. It is clear that another non-basal slip systems must be activated to deform polycrystals of Mg and of Mg-based alloys. Slip in $(\boldsymbol{c}+\boldsymbol{a})$ direction, and/or twinning, is necessary. Slip in $\langle 11\bar{2}3\rangle$ directions and on the second order pyramidal $\{11\bar{2}2\}$ planes are very probable. This indicates that dislocations are moving not only in basal slip system but also in non-basal slip systems. The prismatic and pyramidal slip systems as well as twinning can be considered as non-basal slip systems required for deformation of Mg polycrystals.

In our previous papers [1, 8–10] we have reported the stress-strain curves of some Mg alloys deformed at a constant strain rate at temperatures between 293 and 673 K. The variations of the stress-strain curves, therefore also the stress dependences of work hardening rate at various temperatures, may be accounted for by the activity of non-basal slip systems. The deformation mechanisms for Mg polycrystals are similar to those that take place in stage B of the work hardening curve of Mg single crystals [11].

The aim of the present paper is to analyse the motion of dislocations in basal and non-basal slip systems and to propose dislocation reactions and processes responsible for the work hardening behaviour of magnesium alloys deformed at various temperatures above room temperature.

2. Outline of a model – deformation mechanisms

The work hardening behaviour of Mg, Cd and Zn single crystals has been already investigated in a wide temperature range, e.g. [5, 6, 11]. In contrast to the fcc single crystals, the work hardening rates in stage A and in stage B (for the case of hcp crystals, stages in the work hardening curve are usually called A, B and C) of the work hardening curve are very sensitive to temperature. Above about 0.3 T_{m}, where T_{m} is the absolute melting point, a strong decrease in the work hardening rate ϑ in both stages with increasing temperature is observed. The decrease in ϑ could be caused by recovery processes. Dislocation climb and/or local cross slip of basal dislocations can be taken into account. In this connection it is interesting to note that presence of the double cross slip of basal screw dislocations through the prismatic planes was estimated in Mg single crystals deformed at 300 K [6, 12]. During deformation in stage B of the work hardening curve of single crystals, the motion of dislocations in the pyramidal slip systems is assumed. The number of the moving pyramidal dislocations depends on the density of all pyramidal dislocations, and it changes with the deformation temperature. An increase of non-basal dislocations in stage B was observed [12, 13]. The sessile dislocations increasing hardening can arise due to the following reaction between basal and non-basal dislocations

$$\frac{1}{3}[2\bar{1}\bar{1}3] + \frac{1}{3}[\bar{2}110] \rightarrow [0001]. \tag{1}$$

An interaction among the pyramidal $\boldsymbol{c}+\boldsymbol{a}$ dislocations can take place according to the following dislocation reaction

$$\frac{1}{3}[11\bar{2}\bar{3}] + \frac{1}{3}[\bar{2}113] \rightarrow \frac{1}{3}[\bar{1}2\bar{1}0]. \tag{2}$$

This dislocation reaction can result in softening. The observed strain hardening rate is a sum of hardening and softening mechanisms. Activities of both mechanisms may be influenced by temperature.

Temperature has a great influence on the deformation behaviour of magnesium polycrystals. The flow stress increases with strain, and the flow stress at a certain strain decreases with increasing temperature. The work hardening rate decreases with increasing strain (stress). The stress dependence of the work hardening rate is influenced by temperature and by the strain rate. The stress increase with strain is caused by an increase of the density of dislocation obstacles during deformation. The dislocation reaction described by Eq. (1) and the following one

$$\frac{1}{3}[11\bar{2}0] + \frac{1}{3}[\bar{2}113] \rightarrow \frac{1}{3}[\bar{1}2\bar{1}3] \tag{3}$$

may produce additional dislocation obstacles. The observed increase of non-basal dislocations can occur according to the reaction (3). The macroscopic work hardening rate of Mg polycrystals is a result of a sum of two effects: hardening and softening. The activity of non-basal slip systems depends on the critical resolved shear stress (CRSS) necessary for the dislocation motion in the systems. It has been reported [14, 15] that the CRSS for the second order pyramidal slip in Mg single crystals shows an anomalous temperature dependence; in a certain temperature range the CRSS increases with increasing temperature. The value of the CRSS for pyramidal slip at 293 K is about one hundred times higher than that for basal slip. At certain temperatures the CRSS for non-basal slip systems can decrease with addition of solute atoms, e.g. of lithium [16, 17] or zinc [8, 18].

During deformation of Mg polycrystals the motion of not only $\boldsymbol{a}$ (basal) dislocations but also $\boldsymbol{c}+\boldsymbol{a}$ (pyramidal) dislocations is assumed. Edge dislocations of $\boldsymbol{c}+\boldsymbol{a}$-type can decompose into single $\boldsymbol{a}$ and $\boldsymbol{c}$ dislocations. Because $\boldsymbol{c}$ dislocations cannot move in hcp crystals, $\boldsymbol{c}+\boldsymbol{a}$ dislocations become immobile; a storage of dislocations occurs and, therefore, the flow stress increases. Screw dislocations of $\boldsymbol{c}+\boldsymbol{a}$ type can move to the next slip planes by double cross slip and then annihilation of dislocations can follow. This causes a decrease in the work hardening rate. During deformation, the interactions among dislocations according to the dislocation reactions described above can cause hardening and softening.

In what follows, the work hardening rate of polycrystalline magnesium alloys will be analysed. The yield stress σ_{y} and the flow stress σ depend on the density of dislocations, the concentration of foreign atoms, the volume fraction and distribution of precipitates, the grain diameter, temperature, and the strain rate. We assume that the concentration of solute atoms and the grain size influence only the yield stress, i.e.

$$\sigma_{\mathrm{y}} = \sigma_0(c, d, T) + \alpha G b \rho^{1/2}, \tag{4}$$

where σ_0 is a function of the concentration of solute atoms c, grain size d and temperature T. The flow stress σ changes with strain ε and one can write

$$\sigma = \sigma_{\mathrm{y}} + \sigma_{\mathrm{D}}(\varepsilon, \rho), \tag{5}$$

where the dislocation component σ_{D} depends on strain and on the dislocation density.

As mentioned above, the dislocation density changes with strain, temperature and the strain rate. In the following, the density of the total dislocations will be considered, for the sake of simplicity, as the characteristic parameter of the microstructure. Similarly as Malygin [2] and Lukáč and Balík [3], we will take into account processes of multiplication (storage) of dislocations at both impenetrable non-dislocation obstacles and forest dislocations. The evolution of the dislocation density with strain can be described by the following equation

$$\frac{\partial \rho}{\partial \varepsilon} = K_1 + K_2 \rho^{1/2} - K_3 \rho^2, \tag{6}$$

where $K_1 = 1/bs$, s is the spacing between impenetrable obstacles, K_2 is a geometrical factor based on the assumption that the mean free path of dislocations is proportional to the average spacing between forest dislocations, and K_3 is the coefficient of dislocation annihilation depending on recovery process(es) [2, 3, 19, 20].

The stress dependence of the work hardening rate for polycrystals can be then written in the following form

$$\frac{\partial \sigma}{\partial \varepsilon} = \Theta = A/(\sigma - \sigma_{\mathrm{y}}) + B - C(\sigma - \sigma_{\mathrm{y}})^3, \tag{7}$$

where A depends on the spacing between impenetrable obstacles, B depends on the mean dislocation spacing and C is determined by the dislocation annihilation intensity. The detailed discussion of the presented model of the work hardening rate will be given elsewhere [21].

3. Comparison with experiment

In order to compare predictions of the proposed model – Eq. (7) – with experimental results, an AM20 (Mg-2wt.%Al-0.1wt.%Mn) magnesium alloy was used. Figure 1 shows the stress-strain curves of AM20 polycrystals deformed at a constant strain rate of 6.2×10^{-5} s^{-1} for various temperatures. The stress dependence of the work hardening rate was computed by numerical derivation of the experi-

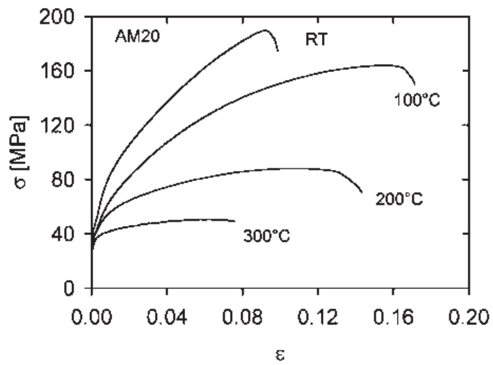

Fig. 1. True stress-true strain curves of AM20 polycrystals deformed at 6.2×10^{-5} s^{-1} at various temperatures.

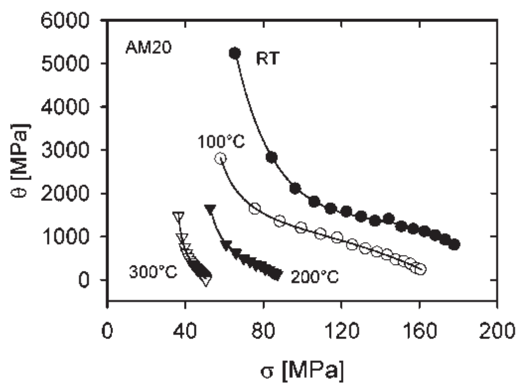

Fig. 2. Stress dependence of the work hardening rate; experimental data (points) fitted by the predicted equation (solid lines).

Table 1. Fitting parameters of the stress dependence of the work hardening rate for AM20 polycrystals deformed at various temperatures

	A [MPa^2]	B [MPa]	$C \times 10^4$ [MPa^{-2}]	σ_y [MPa]	σ_{02} [MPa]	R^2
RT	58849	749	1.03	52	48	0.991
100 ℃	31931	699	4.42	42	41	0.998
200 ℃	13479	22	25.34	44	40	0.992
300 ℃	4503	0.01	0.049	33	36	0.997

mental stress-strain curves. The experimental data (points) have been fitted using Eq. (7) (solid lines). Results of fitting of the proposed model are presented in Fig. 2. It is obvious that close agreement between predictions of the model and the measured stress dependence of the work hardening rates is found. The optimum model parameters obtained from the fit are given in Table 1. While the values of σ_y were obtained from the fit, the values of σ_{02} were determined experimentally as the flow stress at 0.2 offset strain. Good agreement between σ_y and σ_{02} is clearly apparent.

The parameter A, which is connected with the non-dislocation obstacles, should not depend on the temperature. The AM20 alloy contains reduced amounts of the intermetallic phase $Mg_{17}Al_{12}$. Above about 200 ℃, the microstructure of this alloy may change due to dissolution of precipitates [22]. Therefore, the spacing between non-dislocation obstacles should increase and the parameter A should decrease. This is consistent with experimental observations. It should be mentioned that changes in the microstructure of AM alloys at about 200 ℃ were detected by non-destructive methods [23].

The parameter B is related to the work hardening due to the interaction with forest dislocations. In fcc alloys, it should not depend on temperature, which has been verified experimentally [3, 4]. In alloys with the hexagonal structure, the density of forest dislocations can significantly change with temperature, especially at higher temperatures. In magnesium alloys, the activity of non-basal slip systems and, therefore, the changes in the forest dislocation density can occur at about 200 ℃. A decrease in the density of forest dislocations increases the average slip length of dislocations and, therefore, causes a decrease in the parameter B, which has been observed at 200 ℃.

The C parameter relates to recovery due to annihilation of dislocations. The C parameter should increase with increasing temperature due to thermally activated character of recovery process. This fact has been also confirmed by experimental results.

The extremely low values of the A, B, and C parameters in the fit for the stress dependence of the work hardening rate of specimens deformed at 300 ℃ can be attributed to an onset of another recovery process that changes the shape of the stress-strain curves at 300 ℃, which is observed. Recrystallization is observed and it may become crucial and suppress the other recovery processes. The recrystallization behaviour of magnesium has been described in detail by Ion et al. [24] and Sitdikov and Kaibyshev [25].

4. Conclusions

In the present paper, a new model for the work hardening behaviour of alloys with hexagonal structure has been presented. In the model, the change of dislocation density with strain is considered to be a result of the storage of dislocations (due to forest dislocations and non-dislocation obstacles) and annihilation of dislocations. The work hardening behaviour of an AM20 magnesium alloy was investigated at several temperatures between room temperature and 300 ℃ at strain rate of 6.2×10^{-5} s^{-1}. The test temperature influences significantly the shape of the stress-strain curve. The work hardening rates were calculated from the stress--strain curves. The stress dependence of the work hardening rate was fitted with the dependence predicted by the model. The results of fitting are in good agreement with the predictions. The course of the stress-strain curve at 300 ℃ is a consequence of the onset of recrystallization.

Acknowledgements

The paper is dedicated to Professor Dr. Z. Trojanová on the occasion of her 60th birthday. The authors are grateful for the support offered by the Czech and German authorities under the Exchange Programme CZE 01/029.

REFERENCES

[1] MÁTHIS, K.—TROJANOVÁ, Z.—LUKÁČ, P.: Mater. Sci. Eng., *A324,* 2002, p. 141.
[2] MALYGIN, G. A.: Phys. Stat. Sol. (a), *119,* 1990, p. 423.
[3] LUKÁČ, P.—BALÍK, J.: Key Eng. Mater., *97-98,* 1994, p. 307.
[4] KRÁL, R.—LUKÁČ, P.: Acta Univ. Carolinae, Math. Phys., *39,* 1998, p. 49.
[5] LUKÁČ, P.: J. Sci. Ind. Res., *32,* 1973, p. 569.
[6] LAVRENTEV, F. F.—SALITA, O. P.—LUKÁČ, P: Acta Univ. Carolinae, Math. Phys., *23,* 1982, p. 33.
[7] VON MISES, R.: Z. Angew. Math. Mech., *8,* 1928, p. 161.
[8] LUKÁČ, P.: In: Proceedings of the Third International Magnesium Conference. Ed.: Lorimer, G. W. London, The Institute of Materials 1997, p. 381.
[9] LUKÁČ, P.—TROJANOVÁ, Z.—DROZD, Z.: Key Eng. Mater., *188,* 2000, p. 121.
[10] TROJANOVÁ, Z.—LUKÁČ, P.—GÁBOR, P.—DROZD, Z.—MÁTHIS, K.: Kovove Mater., *39,* 2001, p. 368.
[11] LUKÁČ, P.: Czech. J. Phys. B, *35,* 1985, p. 275.
[12] LAVRENTEV, F. F.—POKHIL, J.—SOKOLSKIJ, S. V: In: Strength of Metals and Alloys. Eds.: Haasen, P., Gerold, V., Kostorz, G. Toronto, Pergamon Press 1979, p. 163.

[13] LAVRENTEV, F. F.—POKHIL, J.: Phys. Stat. Sol. (a), *32,* 1975, p. 227.
[14] STOHR, J. F.—POIRIER, J. P.: Phil. Mag., *25,* 1972, p. 1313.
[15] ANDO, S.—NAKAMURA, K.—TAKASHIMA, K.—TONDA, H.: J. Japan Inst. Light Metals, *42,* 1992, p. 765.
[16] URAKAMI, A.—MESHI, M.—FINE, M. E.: In: Proceedings ICSMA 2. Vol. 1. Asilomar, ASM 1970, p. 272.
[17] ANDO, S.—TONDA, H.: Mater. Sci. Forum, *350-351,* 2000, p. 43.
[18] AKHTAR, A.—TEGHTSOONIAN, E.: Acta Metall., *17,* 1969, p. 1351.
[19] GOTTSTEIN, G.—ARGON, A. S.: Acta Metall., *35,* 1987, p. 1261.
[20] KUHLMANN-WILSDORF, D.: In: Work Hardening. Eds.: Hirth, J. P., Weertman, J. New York, Gordon & Breach 1968, p. 97.
[21] LUKÁČ, P.: to be published.
[22] ASM Speciality Handbook Magnesium and Magnesium Alloys. Eds.: Avedesian, M. M., Baker, H. Materials Park, OH, ASM International 1999.
[23] TROJANOVÁ, Z.—LUKÁČ, P.—RIEHEMANN, W.—VON BUCH, F.: In: Magnesium Alloys and their Applications. Ed.: Kainer, K. U. Weinheim, Wiley-VCH 2000, p. 678.
[24] ION, S. E.—HUMPHREYS, F. J.—WHITE, S. H.: Acta Metall., *30,* 1982, p. 1909.
[25] SITDIKOV, O.—KAIBYSHEV, R.: Mater. Trans. JIM, *42,* 2001, p. 1928.

Received: 24.6.2002