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Elevated Temperature Deformation Behavior in an AZ31 Magnesium Alloy

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An AZ31 magnesium alloy was tested at constant temperatures ranging from 423 to 473 K (0.46 to 0.51 T_m) under constant stresses. All of the creep curves exhibited two types depending on stress levels. At low stress ($\sigma/G<4\times10^{-3}$), the creep curve was typical of class A (Alloy type) behavior. However, at high stresses ($\sigma/G>4\times10^{-3}$), the creep curve was typical of class M (Metal type) behavior. At low stress level, the stress exponent for the steady-state creep rate was of 3.5 and the true activation energy for creep was 101 kJ/mole which is close to that for solute diffusion. It indicates that the dominant deformation mechanism was glide-controlled dislocation creep. At low stress level where $n=3.5$, the present results are in good agreement with the prediction of Fridel model.

Key Words : Magnesium Alloy, Creep Deformation, Stress Exponent, Activation Energy, Dislocation Glide

1. Introduction

Magnesium alloys are emerging as potentially good candidates for numerous applications, especially in the automotive industry due to their superior specific elastic modulus and specific strength, and so on (Friedrich and Schumann, 2001). Sufficient creep resistance is major requirement for use of magnesium alloys in automotive power-train components that are currently made in aluminum alloy or cast iron. Automatic transmission cases and engine blocks are used at elevated temperatures up to 473 K (Luo, 2004). However, these alloys have poor creep resistance at temperatures above 400 K, which has made them inadequate for power-train applications. The poor creep strength of magnesium in these components can cause clamping load reduction in bolted joints, resulting in leakage in the power-train components.

Researches on creep deformation in pure magnesium and its alloys were mainly done at high temperatures ($T>0.5\ T_m$) (Vagarali and Langdon, 1981 ; 1982 ; Somekawa et al., 2005 ; Suzuki et al., 1998 ; Watanabe et al., 2001). For example, Vagarali and Langdon (1982) performed creep tests on pure magnesium over the temperature 473 to 820 K. According to these authors, creep is controlled by dislocation climb. The activation energy for creep is independent of stress and is 135 kJ/mole, which is agreement with the value for lattice self-diffusion. The stress exponent is close to about 5 but increases to 6.5 at 473 K. Somekawa et al. (2005) investigated creep behavior in Mg-Al-Zn alloys with different aluminum contents over the temperature 473 to 623 K. At

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high temperatures, the stress exponent was 5 and the activation energy was close to that for lattice diffusion of magnesium, whereas at low temperatures, the stress exponent was 7 and the activation energy was close to that for pipe diffusion for all alloys. They suggested that the climb-controlled dislocation creep was governed by pipe diffusion at low temperatures and by lattice diffusion at high temperatures in these alloys. Suzuki et al. (1998) investigated creep behavior in Mg-Y alloys at 550 K under 50～200 MPa. Two stress regions were recognized based on the stress dependence of the minimum creep rate. The stress exponent is about 5, but it takes a higher value in a higher stress range. The cause of the high stress exponent in the high stress region is considered to be due to the power-law breakdown.

However, deformation behavior at low temperatures ($T<0.5\ T_m$), where the automotive power-train components operate, has rarely been investigated in pure magnesium and its alloys (Shi, 1994 ; Spigarelli et al., 2000 ; Evangerlista et al., 2005). Spigarelli et al.(2002) investigated on AZ91 magnesium alloy in the temperature range from 393 to 453 K. They described the minimum creep-rate dependence on applied stress by means of the sinh equation rather than the conventional power law due to that fact that the applied stresses used in the creep test were higher than the transition stress for power break-down. They suggested that the creep was controlled by dislocation climb at the test temperatures in this alloy. Therefore, the objective of this study is to evaluate creep resistance of conventionally extruded AZ31 magnesium alloy and to investigate creep mechanisms operating in this alloy at the temperature range from 423 to 473 K.

2. Experimental Procedures

All the creep tests were conducted using double-shear specimens from rods of conventionally extruded AZ31 magnesium alloy with a diameter of 25 mm. The nominal composition of the alloy is Mg-2.4Al-0.89Zn-0.39Mn in wt.%. The configuration and dimensions for the creep specimen is shown in Fig. 1. Prior to testing, all the specimens were annealed for 2 hours at 693 K. The grain size was found to be 48 μm after the heat treatment. The creep tests were conducted in air in a three-zone furnace. The test temperature was monitored with chromel-alumel thermocouples held in contact with the specimen and was maintained within ±1 K of the reported temperature. The strain during creep was measured with a linear variable differential transformer (LVDT), accurate to 1.7×10^{-3} mm. The LVDT signal was amplified, and monitored directly on a strip chart recorder. Details of the creep test procedure are described in the work by Kim et al.(1998). The shear stress, τ, and shear strain, γ, were converted to normal stress, σ, and normal strain, ε, using the expression $\sigma=2\ \tau$ and $\varepsilon=2/3\ \gamma$ (Isshiki et al., 1997). The samples were tested at constant temperatures ranging from 423 to 473 K (0.46 to 0.51 T_m, T_m=923 K based on T_m of pure magnesium) under constant stresses.

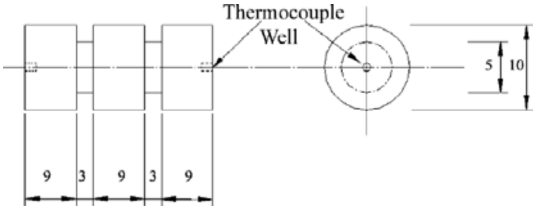
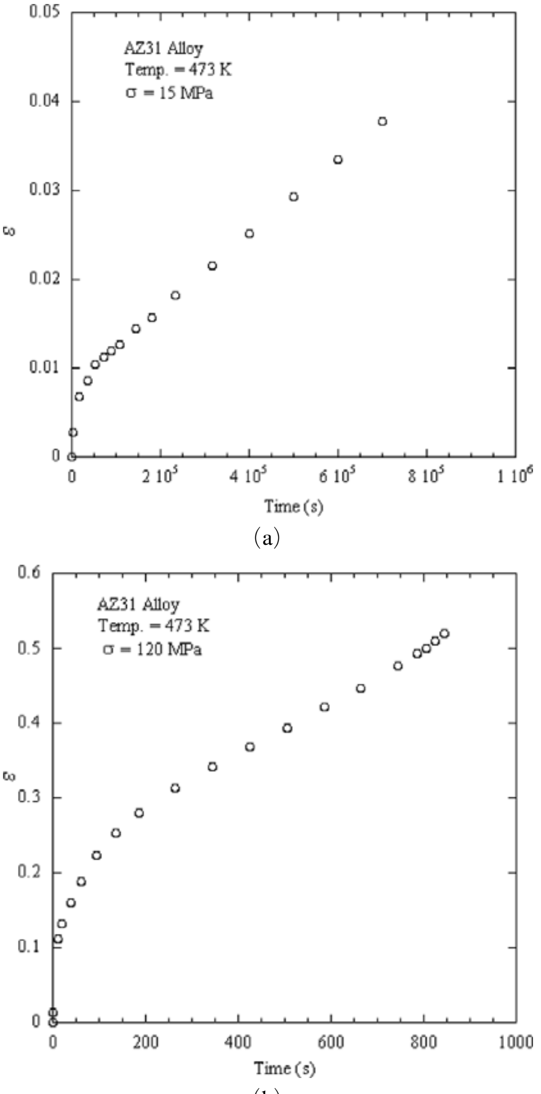


Fig. 1 Schematic of double-shear creep specimen geometry (in mm)

3. Results and Discussion

3.1 Creep curves

A large number of creep tests were conducted at stresses ranging from 10 to 140 MPa. All of the creep curves obtained in the present study exhibited two types dependent on stress levels. At low stress, as shown in Fig. 2(a), the shape of the creep curve was typical of class A (Alloy type) behavior (Cannon and Sherby, 1970): there was a very small instantaneous strain on application of the load, and then a very brief primary stage which preceded steady-state flow. However, at high stresses, as shown in Fig. 2(b), the shape of the creep curve was typical of class M (Metal type) behavior (Cannon and Sherby, 1970) : there is a



(a)

(b)

Fig. 2 Creep strain against time curves under (a) 15 MPa and (b) 120 MPa at 473 K

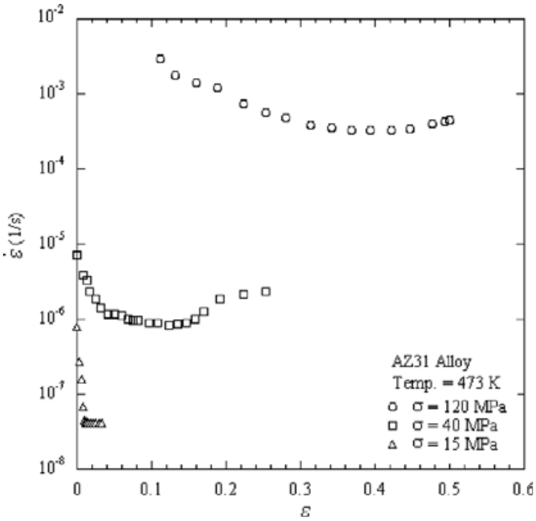


Fig. 3 Strain against strain rate for specimens tested at 473 K at different stress levels

normal primary creep stage, during which the creep rate decreases continuously with increasing time, that is then followed by a well-defined steady state period for which the creep rate remains essentially constant. The two types of creep curve are clearly illustrated by the plots shown in Fig. 3 of the logarithmic strain rate against the total strain $\dot{\varepsilon}$ for tests conducted at different stress levels at the same temperature. At the low stress (15 MPa), the primary stage is very brief. As the stress increases (40 and 120 MPa), the normal primary stage extends to higher strain.

3.2 Stress dependence of the steady state creep rate

The results from a number of tests conducted at three different temperatures are shown in Fig. 4 in the form of steady state creep rate, $\dot{\varepsilon}$, against the applied tensile stress, σ, on a logarithmic scale. The data show that the value of stress exponent, n $(=\partial \ln/\partial \ln \sigma)$ was independent of temperature and that there is a significant variation in the stress exponent with stress level. The value of stress exponent was found to be 3.5 ± 0.2 at low stress level and increased up to 6 ± 0.5 at high stress level. The transition stress delineating these two regions decreases with increasing temperature. Similar trend was reported in Mg-0.8% wt.% Al alloy where the value of n was 3 at low stress level and 6 at high stress level at 523 to 573 K (Vagarali and Langdon, 1982). At present, it is not clear whether creep power-law break-down occurs at the transition stress. It will be discussed later in detail. The value of the stress exponent 3.5 is also in good agreement with value of ~ 3 reported for AM50 magnesium alloy at 398 K (Evangerlista et al., 2005). However, the value of $n=8$ exhibited in AZ91 alloy at 393 to 453 K (Spigarelli et al., 2000). The stress exponent of

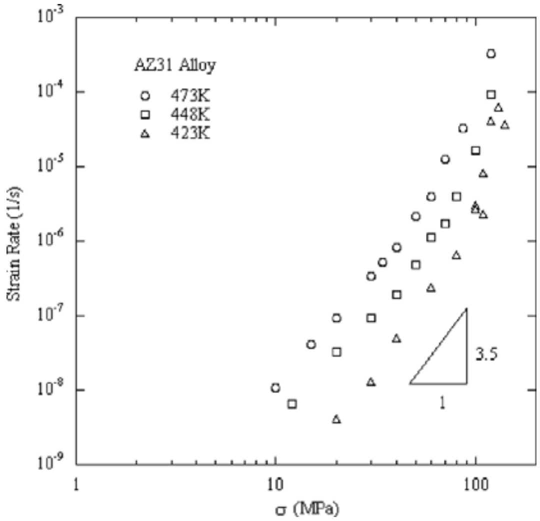


Fig. 4 Steady-state strain rate against stress for specimens tested from 423 to 473 K

3.5 suggests that the creep of AZ31 magnesium alloy is controlled by a dislocation glide process at low stress level.

3.3 Activation energy for creep

An activation energy for creep can be determined using the power-law creep equation

$$\dot{\varepsilon}=A\sigma^{n}\exp(-Q/RT) \tag{1}$$

where $\dot{\varepsilon}$ is the steady state creep rate, n the stress exponent and A material constant, Q the activation energy, R the gas constant and T the absolute temperature. For calculating the activation energy for the creep, the data of Fig. 4 were used to plot logarithmic $\dot{\varepsilon}G^{n-1}\ T$ against $1000/T$ at low stresses (σ=20, 30, 40 and 60 MPa) and at high stress (σ=80 MPa). The activation energy was then determined from the slop of the resultant straight line which, according to well-documented analysis (Robinson and Sherby, 1969) is equal to $-2.3R\dfrac{d\log\dot{\varepsilon}}{d\left(\frac{1}{T}\right)}$. In estimating Q from the method, information concerning the shear modulus, G, was taken from the data available on the pure magnesium (Frost and Ashby, 1982). The modulus G for magnesium was expressed as $1.92\times10^{4}-8.6T$ (MPa). The average value of Q obtained from the analysis of the data is 101 kJ/mole at low stresses from 423 to 473 K. The value of Q= 101 kJ/mole, as shown in Fig. 5(a), is close to that for pipe diffusion of magnesium (=92 kJ/mole) (Frost and Ashby, 1982). The value of Q is in good agreement with value of 106 kJ/mole reported for pure magnesium alloy at 423 to 523 K (Shi et al., 1994). In general, the activation energy Q is equal to that for lattice diffusion (Q_L) above 0.6 T_m. And, the activation energy is equal to that for dislocation pipe diffusion (Q_p) and $Q=2/3Q_L$ from about 0.4 to 0.6 T_m (Robinson and Sherby, 1969). The test temperatures employed in the present study for AZ31 magnesium alloy were in the range of 0.46 to 0.51 T_m and the measured Q value (Q/Q_L=0.75) is consistent with that has been reported in the literature (Robinson and Sherby, 1969). It is thus suggested that the glide-controlled dislocation creep is controlled by pipe diffusion at low stress level. However, as shown in Fig. 5(b), the value of Q obtained at high stress (σ=80 MPa) is 119 kJ/mole. The value of Q=119 kJ/mole is slightly lower than that for lattice self-diffusion in magnesium (135 kJ/mole) (Shewmon and Rhines, 1954). Thus, it can be expected that there is a transition to a new mechanism at high stress level.

In fact, there are some conflicting results on the creep mechanism of magnesium alloys at low temperature (Evangerlista et al., 2005; Somekawa, 2005). For example, Evangerlista et al. (2005), investigated on AZ91 magnesium alloy in the temperature range from 393 to 453 K. They described the minimum creep-rate dependence on applied stress by means of the sinh equation rather than the conventional power law due to that fact that the applied stresses used in the creep test were higher than the transition stress for power breakdown. They suggested that the creep was controlled by high temperature climb at the test temperatures in this alloy. On the other hand, Somekawa (2005) reported the stress exponent of 7 when tested at 473 in AZ31 alloy and proposed the dislocation climb controlled creep is governed by pipe diffusion. These results indicate that creep deformation mechanisms of magnesium alloys, especially at low temperature ($T<0.5\ T_m$), are complicate depending on several factors such as stress, temperature, microstructure and alloy system.

3.4 Creep mechanism in AZ31 magnesium alloy

The experimental results therefore divided into two distinct types of behavior as follows ;

(i) Stress regime with n=3.5 behavior

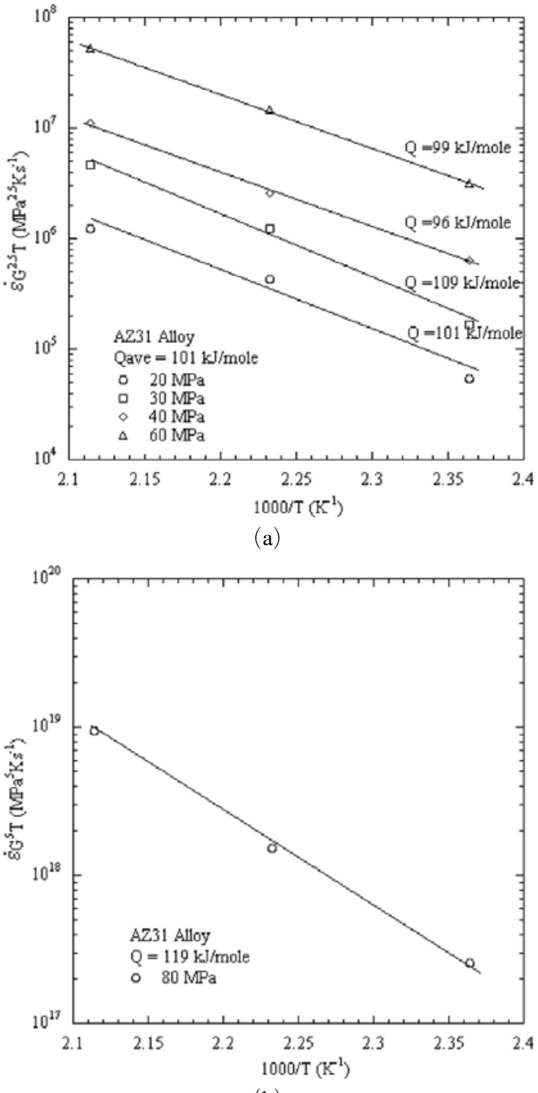
The experimental results in the region where $n=3.5$ provide strong evidence that creep of the AZ31 alloy occurs by a viscous glide process as in class A behavior. The various points of agreement with this mechanism include not only the stress exponent of 3.5 but also the lack of a significant instantaneous strain upon loading and the very brief normal primary stage of creep.

In class I or Alloy type creep, dislocations are understood to glide in a viscous manner due to their interaction with solute atoms. Several different drag processes have been proposed (Weertman, 1957 ; Takeuchi and Argon, 1976 ; Fridel, 1964). Among these mechanisms, calculations show that the major force retarding the glide of dislocations often arises from the presence of impurity atmospheres. In the theory of Weertman (1957), it is assumed that the motion of dislocations occurs as sequential glide and climb processes, and the slower of these two processes is the rate-controlling mechanism. In the solid solution alloys when glide is slower than climb, the steady-state creep rate is given by

$$\dot{\varepsilon}=\frac{0.35}{e^2 c}\left(\frac{kT}{Gb^3}\right)^2\left(\frac{\bar{D}Gb}{kT}\right)\left(\frac{\sigma}{G}\right)^3 \tag{2}$$

where e is the solute-solvent size factor, c the solute concentration, k Boltzmann's constant $(=1.38\times10^{-23}\ \mathrm{J/K})$, b Burgers, $\bar{D}$ the chemical interdiffusivity of the solute atoms. On the other hand, the theory of Takeuchi and Argon (1976) is based on Cottrell-Jaswan interaction. By considering the rates of dislocation multiplication and annihilation, the steady-state creep rate due to dislocation glide is given by

$$\dot{\varepsilon}=\frac{0.125}{e^2 c}\left(\frac{kT}{Gb^3}\right)^2\left(\frac{\bar{D}Gb}{kT}\right)\left(\frac{\sigma}{G}\right)^3 \tag{3}$$



(a)

(b)

Fig. 5 Determination of the true activation energy for creep in AZ31 (a) at low stresses and (b) at high stress by plotting $\dot{\varepsilon}G^{n-1}T$ against $1/T$

In the theory of Fridel (1964), it is assumed that the diffusion of solute atoms is assisted by the line tension of the dislocations. The steady-state creep rate is given by

$$\dot{\varepsilon}=0.18\left(\frac{\bar{D}Gb}{kT}\right)\left(\frac{\sigma}{G}\right)^3 \tag{4}$$

To inspect the prediction of these three dislocation glide models, the datum points shown in Fig. 4 were replotted in logarithmic form of $\bar{D}kT/Gb$ against σ/G, putting $e=-0.1373$ (King, 1966), $b=3.21\times10^{-10}$ m $G=1.92\times10^4-8.6T$ (MPa) (Frost and Ashby, 1982) and taking $\bar{D}$ as the chemical

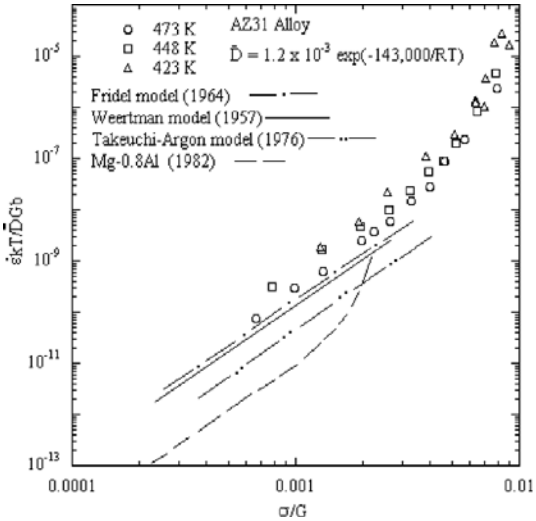


Fig. 6 Normalized strain rate against normalized stress for temperature from 423 to 473 K

interdiffusivity of aluminum in magnesium ; $\bar{D}=1.2\times10^{-3}\exp(-143{,}000/RT)$ m^2s^{-1} (Vagarali and Langdon, 1982). The effect of solid solution of zinc and manganese, which are minor alloys addition to Mg-Al-Zn system alloys, are neglected for the sake of simplicity plots, the solute (=aluminum) concentration $c=0.022$. The results is shown in Fig. 6, and all of experimental points now lie on a single line of slope of about 3 at low stresses but with an increase in slope at $\sigma/G>4\times10^{-3}$. Fig. 6 shows also the predicted creep rates for the theories of Weertman (1957), Takeuchi and Argon (1976), and Fridel (1964), as given by equations (2), (3) and (4), respectively. It included the experimental data reported by Vagarali and Langdon (1982) for Mg-0.8%Al. Within the region where $n=3$, the present results are in better agreement with the prediction of Fridel model (1964). However, the experimental creep data reported by Vagarali and Langdon (1982) for Mg-0.8%Al are over an order of magnitude lower than the present results.

(ii) Stress regime with n ~6 behavior

There is a sharp deviation from the behavior with n ~3 at high stress ; the stress exponent in this region is ~6 and activation energy is again 119 kJ/mole. A similar trend was reported earlier for Mg-0.8wt%Al alloy (Vagarali and Langdon, 1982), where n ~3 at low stress level and there was a deviation to higher value of n (~6) at $\sigma/G>10^{-3}$. Vagarali and Langdon (1982) argued that the deviation from linearity at high stress was not due to power-law breakdown.

It is necessary to inspect whether creep power-law break-down occurs at the transition stress. In practice, materials exhibiting class M behavior show a break-down in power-law creep and a consequent increase in the value of n at values of $\dot{\varepsilon}/D$ above ~$10^{13}m^{-2}$, where D is the appropriate diffusion coefficient (Robinson and Sherby, 1969). However, the situation is more complex in alloys exhibiting class A behavior because there are the possibilities both of normal power-law break-down, as in class M, and of an increase in n due to the break-away of dislocations from their solute atmospheres (Fridel, 1964).

There have been several experimental investigations of solid solution alloys to examine the deviation at high stress from viscous glide behavior with n ~3. For example, Yavari and Langdon (1982) showed that there was marked change in the creep behavior of Al-Mg alloys with increasing stress such that there was an increase in the stress exponent from 3 to 4.5 and this change was due to the break-away of dislocations from their solute atmospheres. Also, it was demonstrated that for various solid solution alloys, the values of the experimental stresses marking the transitions from viscous glide to a break-away condition were in very good agreement with the stresses predicted by break-away relationship developed by Fridel (1964). Following Fridel (1964), the break-away stress may be expressed as

$$\tau_b=A_b\left(\frac{W_m^2\,c}{kTb^3}\right) \tag{5}$$

where τ_b is the shear stress necessary to break a dislocation from its solute atmosphere, A_b is a dimensionless constant, W_m is the maximum interaction energy between a solute atom and an edge dislocation and c is the solute concentration. The value of W_m may be calculated from the theoretical expression (Fridel, 1964)

$$W_m=-\frac{1}{2\pi}\left(\frac{1+\nu}{1-\nu}\right)G\,|\Delta V_a| \tag{6}$$

where ΔV_a is the difference in volume between the solute and solvent atom, ν Poisson's ratio. For aluminum in magnesium, $\Delta V_a = 8.2 \times 10^{-30}$ m^3 (King, 1966). The values of W_m at $T = 423$, 448 and 473 K were calculated, putting $b = 3.21 \times 10^{-10}$ m, $\nu = 0.34$, $G = 1.92 \times 10^4 - 8.6T$ (MPa) (Frost and Ashby, 1982), $c = 0.022$ and $\sigma = 2\tau$. However, equation (5) predicts values for the break-away stress are too high when the value of dimensionless constant Ab is taken as unity by Endo et al. (1984). On the other hand, Kuchařová et al. (1974) deduced that A_b was close to 0.2 for Al-Mg alloys and Murty (1973) included the size effect and obtained a value of A_b equal to 0.25. If A_b is taken as 0.3 instead of unit, eq. (5) can correctly predict the magnitude of the break-away stresses of the present result. The predicted values of 63, 69.6 and 76.2 MPa for $T = 473$, 448 and 423 K, respectively, are very close to the experimental transition stress values of ~60, 70, and 80 MPa for $T = 473$, 448 and 423 K, respectively. In addition, the value of $\dot{\varepsilon}/D$ for the present AZ31 alloy was at least one order of magnitude above the calculated value. It should be noted that the stress exponent $n \sim 6$ at the high stress in the AZ31 alloy is similar to the values of $n \sim 5$ obtained in pure magnesium (Vagarali and Langdon, 1981) and $n \sim 6$ obtained in Mg-0.8%Al (Vagarali and Langdon, 1981) in the dislocation climb. Finally, it might be concluded that the creep behavior of AZ31 magnesium alloy at 423 to 473 K at the high stress when $n \sim 6$ is controlled by dislocations break away from their solute atmospheres. It is also anticipated that there will be breakdown in power-law behavior at even higher stress levels in AZ31 alloy.

4. Conclusions

An AZ31 magnesium alloy was tested at constant temperatures ranging from 423 to 473 K (0.46 to 0.51 T_m) under constant stresses. The conclusions can be summarized as follows :

(1) All of the creep curves exhibited two types dependent on stress levels. At low stress ($\sigma/G < 4 \times 10^{-3}$), the creep curve was typical of class A (Alloy type) behavior. However, at high stresses ($\sigma/G > 4 \times 10^{-3}$), the creep curve was typical of class M (Metal type) behavior.

(2) At low stress level, the stress exponent for the steady-state creep rate was of 3.5 and the true activation energy for creep was 101 kJ/mole which is close to that for solute diffusion, indicating that the dominant deformation mechanism was glide-controlled dislocation creep.

(3) At high stress level, the stress exponent for the steady-state creep rate was of 6.0 and the true activation energy for creep was 119 kJ/mole which is slightly lower than that for lattice self-diffusion in magnesium, indicating that there is a transition at high stress level when the dislocations break away from their solute atmospheres.

(4) At low stress level where $n = 3.5$, the present results are in better agreement with the prediction of Fridel model rather than that of Takeuchi and Argon or Weertman models.

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