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Breakdown of the power-law creep in a Class I Al–10 at % Zn alloy

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The creep behaviour of Al–10 at % Zn at 573 K is divisible into three deformation regions; low stress region, intermediate stress region and high stress region. The creep characteristics of the low stress region and intermediate stress region are consistent with dislocation climb and viscous glide, respectively. In the high stress region, the stress exponent, n increases with stress, the activation energy is higher than those observed in the other two regions, the activation area is slightly decreasing with stress and the internal stress is almost negligible. Present analysis shows that these characteristics are consistent with the thermally-activated glide motion of dislocations as a rate controlling mechanism at high stresses.

1. Introduction

Following recent reviews [1–3], the creep behaviour of solid solution alloys at intermediate stresses (power-law region) can be divided into two classes according to the value of the stress exponent, n [$= (\partial \ln \dot{\gamma}/\partial \ln \tau)$, where $\dot{\gamma}$ is the shear strain rate and τ is the applied shear stress]. For Class I alloys, n has a value of ~ 3 while for Class II alloys n is similar to that observed for pure metals and is close to 5. However, at high normalized stresses $\sigma/G > 5 \times 10^{-4}$, where σ ($= 2\tau$) is the applied normal stress and G is the shear modulus, the creep rates increase rapidly with stress to higher values than those predicted by the extrapolation from the power-law region.

For pure metals and metal-type (Class II) alloys in which the creep rate in the power-law region is controlled by the climb of dislocations, three suggestions have been offered to explain the power-law breakdown. Firstly, Weertman [4] suggested that power-law breakdown may be attributed to the transition in climb velocity of dislocations from a linear (power-law domain) to an exponential function of stress at high stresses. Secondly, Sherby and Young [5] explained the breakdown on the basis of transition from climb behaviour controlled by lattice diffusion at high temperatures to that controlled by pipe diffusion at low temperatures. Thirdly, it is suggested by many investigators [6–8] that power-law breakdown results from the transition from diffusion controlled climb at intermediate stresses to thermally-activated dislocation glide at high stresses.

Breakdown of power-law creep in alloys in which viscous glide processes are rate controlling (Class I) is expected when a critical value of stress is reached for dislocations to breakaway from their solute atmospheres [4, 9]. For these alloys, it is suggested [10–12] that some form of dislocation climb is rate controlling at high stresses based on the observations that n in this region is not far from 5 and the occurrence of extensive primary creep. Although there is general agreement between those observations and the creep characteristics of Al–10 at % Zn [12, 13] at high stresses, the following two points are not in harmony with the above explanation.

1. n values tend to increase with increasing stress (see Fig. 1) [12, 13].
2. The activation energy for creep in this region is higher than those observed in the low stress region (climb) [12] and intermediate stress region (viscous glide) [12, 13].

In this paper, the thermally-activated dislocation glide mechanism will be examined as a possible rate controlling process at high stresses and will be shown that this mechanism is consistent with the creep characteristics of the alloy in the high stress region.

2. Analysis and discussion

Fig. 1 shows a logarithmic plot of shear strain rate, $\dot{\gamma}$, against shear stress, τ, for Al–10 at % Zn at 573 K [12, 13]. The creep data in Fig. 1 can be divided into three regions: climb region ($n = 4.4$), viscous glide region ($n = 3$) and high-stress region with n increasing with stress; it increases from ~ 5 at $\tau = 8$ MPa to ~ 7.8 at 20 MPa. The activation energies measured in these regions are independent of stress and have values of 128 kJ mol^{-1} [12], 140 kJ mol^{-1} [12, 13], and 171 kJ mol^{-1} [12, 13] for the climb, viscous glide and high-stress regions, respectively. The creep characteristics of the climb region and viscous glide region along with the transition between these two regions were discussed in detail elsewhere [12] and for high-stress region it is suggested that climb is rate controlling [12]. However, the values of the exponent, n, (Fig. 1) and the activation energy in this region are not in agreement with those for the climb region and hence refuting the possibility that dislocation climb is rate controlling in high-stress region.

The activation area, A^*, as measured by stress change experiments [13] in viscous glide and high-

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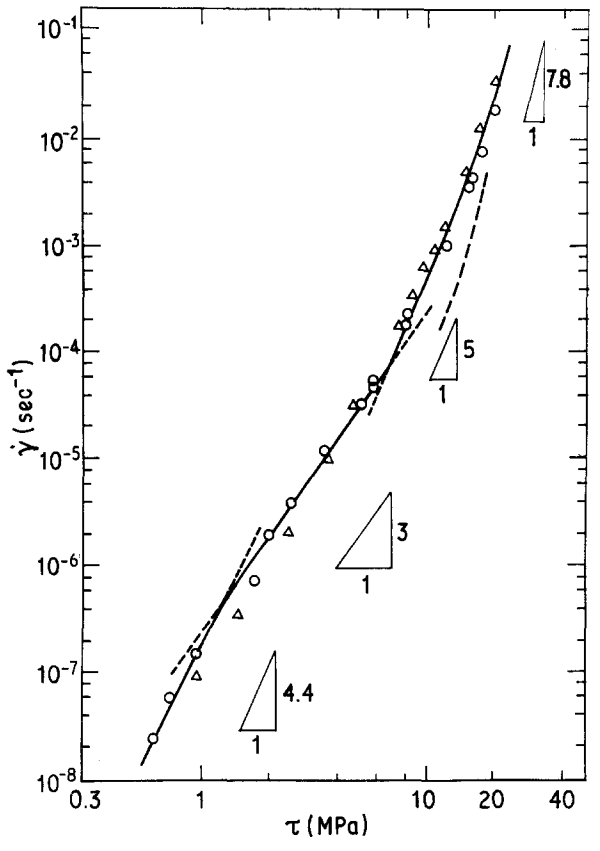

Figure 1 Log $\dot{\gamma}$ against log τ for Al–10 at % Zn at 573 K. The relations: $\tau = \sigma/2$ and $\dot{\gamma} = 3/2\dot{\varepsilon}$ were used to convert the tensile creep data of Kuchařová and Čadek [13]. The dotted line represents $\dot{\gamma}$ values calculated from the rate equation of the thermally-activated dislocations glide. (△) Kuchařová and Čadek [13]. (○) Soliman and Mohamed [12].

stress regions, is plotted against the applied stress, τ on a logarithmic scale in Fig. 2. As shown in the figure, A^* is inversely proportional to τ in the viscous glide region ($A^* \propto \tau^{-1}$) and A^* is related to stress in the high stress region by the relation $A^* \propto \tau^{-0.25}$. The change in the stress dependence of the activation area from viscous glide region to high-stress region suggests a transition in the rate controlling mechanism. In addition, it has been shown by a recent analysis [12] that high stress region occurs at stresses higher than that required for dislocation breakaway from a solute atmosphere lending support to the conclusion that viscous glide is not rate controlling at high stresses.

The glide motion of thermally-activated dislocations will be examined as a possible rate controlling process for the high-stress region. For this mechanism when dislocations cutting or by-passing a regular array of square obstacles, it is established that the steady state creep rate is given by [7, 8, 14]

$$\dot{\gamma} = \dot{\gamma}_0 \exp\left[-\frac{\Delta F}{kT}(1 - \tau_e/\hat{\tau})\right] \quad (1)$$

where $\dot{\gamma}_0$ ($= \varrho_m \boldsymbol{b}^2 \beta \nu$) is independent of temperature, ϱ_m is the mobile dislocation density, $\boldsymbol{b}$ is the Burger's vector, β is a dimensionless constant, ν is Debye frequency, ΔF is the total free energy required to overcome the obstacle in the absence of any external stress, τ_e is the effective stress, $\hat{\tau}$ is the stress required to overcome obstacle without any thermal activation, and kT has the usual meaning (Boltzman's constant and temperature). At steady state, ϱ_m, is a function of stress and the well accepted one consistent with theory and experiment is [15]

$$\varrho_m = \alpha(\tau/G\boldsymbol{b})^2 \quad (2)$$

where α is a constant of order of unity. The value of ΔF is defined as

$$\Delta F = A^* \boldsymbol{b} \hat{\tau} \quad (3)$$

where A^* is the activation area swept by a dislocation in an activated event.

As can be seen from Equation 1, $\dot{\gamma}$ depends on ΔF, A^*, τ_e and τ. Data on A^* and τ_e are available [13];

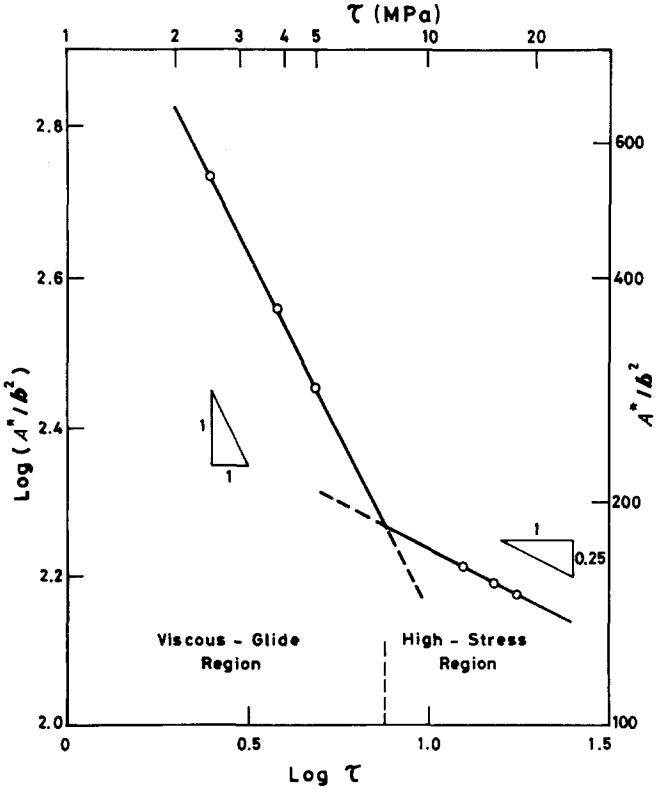

Figure 2 A logarithmic plot of activation area, A^* as a function of the applied stress, τ at $T = 573$ K.

values of A^* are shown in Fig. 2 and τ_e can be approximated by a linear function of the applied stress τ ($\tau_e = 0.91\tau$). The internal stress τ_i ($=\tau - \tau_e$) is a small fraction of the applied stress which is in agreement with the assumption that the internal stresses in fcc metals are negligible at high stresses [8].

To calculate ΔF, we consider the two types of obstacles which exist in solid solutions, namely; the forest of dislocations and the isolated solute atoms. For the forest of dislocations, ΔF_d, usually approximated by the relation $\Delta F_d \approx 0.5Gb^3$ [14], and for isolated solute atoms ΔF_a is given by [16, 17].

$$\Delta F_a \simeq 2\left(W_a G b^3 \frac{r_s^2}{bd}\right)^{1/2} \quad (4)$$

where r_s is the linear dimension of the obstacle ($r_s \simeq 1.5b$), d is the distance between solute atoms in the dislocation core, and W_a is the energy of elastic interaction of dislocation with a solute atom. According to Friedel [18] W_a is given by

$$W_a = -\frac{1}{3\pi}\left(\frac{1+\nu}{1-\nu}\right) G|\Delta V_a| = -0.3Gb^3 e \quad (5)$$

where ν is Poisson's ratio ($\nu \simeq 1/3$), and ΔV_a and e are the difference in volume and the misfit ratio between the solute and solvent atoms, respectively; e for the Al–Zn solid solution is $\simeq -0.02$ [19].

The distance between solute atoms in the dislocation core, d, is given by [18]

$$d = (b/c)\exp\left(\frac{-|W_a|}{kT}\right) \quad (6)$$

where c is the solute concentration $= 0.1$. Taking the shear modulus, G [20, 21]

$$G = G_0 - (\Delta G)T \quad (7)$$

where $G_0 = 2.926 \times 10^4$ MPa and $\Delta G = 17\,\text{MPa K}^{-1}$, W_a was calculated and substituting for W_a and c in Equation 6, d was found to be equal to $\sim 7b$. Also, ΔF_a was determined to have a value of $\sim 0.09Gb^3$. This value is in good agreement with what is expected [14].

The hardening due to dislocation forest and solute atoms represent two independent processes and their effect can be assumed linear, then the total height of the barrier to be surmounted by a moving dislocation in the absence of external stress is

$$\Delta F = \Delta F_d + \Delta F_a = 0.59Gb^3$$

and Equation 1 can be rewritten in the form

$$\dot{\gamma} = \dot{\gamma}_0 \exp\left[-(0.59Gb^3 - A^* b\tau_e)/kT\right] \quad (8)$$

Taking the natural logarithm of both sides of Equation 8 and differentiating with respect to $1/T$, the activation energy for thermally activated glide process, Q_c, is found to be

$$Q_c = -k\left[\frac{\partial \ln \dot{\gamma}}{\partial\left(\frac{1}{T}\right)}\right]_\tau = 0.59b^3\left(G - T\frac{\partial G}{\partial T}\right) - A^* b\tau_e + Tb\left(\tau_e \frac{\partial A^*}{\partial T} + A^* \frac{\partial \tau_e}{\partial T}\right) \quad (9)$$

The effect of temperature on the values of A^* and τ_e

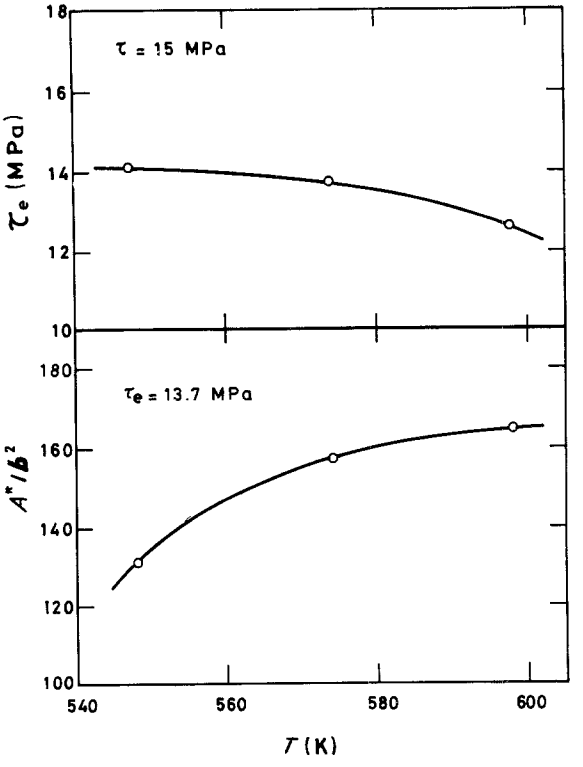

Figure 3 Values of τ_e and A^* as a function of temperature.

[13] is shown in Fig. 3. As can be seen from the figure A^* increases with temperature ($\partial A^*/\partial T$ is positive) while τ_e decreases with increasing temperature ($\partial \tau_e/\partial T$ is negative). Then if the values of the two terms in the last parenthesis of Equation 9 are comparable, Q_c may become independent of stress (the effect of the second term of Equation 9 is negligible) in agreement with the experimental data [12, 13]. To examine this possibility the two terms are calculated at 573 K and $\tau_e = 13.7$ MPa (data on A^* as a function of temperature is only available at this stress). The difference between those two values was found to be 0.1 eV (10 kJ mol^{-1}) which is within the experimental accuracy. Assuming that the value of the last parenthesis of Equation 9 is approximately zero, the value of Q_c, being independent of stress, is calculated to be 50.7 kcal mol^{-1} (212 kJ mol^{-1}) which is higher than the experimental value (171 kJ mol^{-1}).

But since the high-stress region occurs at stresses higher than that required for dislocations to breakaway from their solute atmospheres [12], it is expected that the effect of solute atmospheres on the motion of dislocations at high stresses is negligible [22]. Therefore, the barrier for moving dislocations is represented by the intersection process, i.e., $\Delta F \simeq \Delta F_d \simeq 0.5Gb^3$. Then Q_c is calculated to have a value of 41.8 kcal mol^{-1} (175 kJ mol^{-1}) which is in good agreement with the experimental value (171 kJ mol^{-1}).

Now, we calculate the stress exponent n using Equation 8:

$$n = \left(\frac{\partial \ln \dot{\gamma}}{\partial \ln \tau}\right)_T = \frac{\partial \ln \dot{\gamma}_0}{\partial \ln \tau} + \frac{A^* b\tau_e}{kT}\left(1 + \frac{\partial \ln A^*}{\partial \ln \tau}\right) \quad (10)$$

Since $\dot{\gamma}_0$ is proportional to ϱ_m, then from Equation 2,

TABLE I Values of n (Equation 11) and n_c (calculated from the slope of the creep data, Fig. 1) at $T = 573$ K

	τ (MPa)		
	12.5	15	17.5
n	6.1	6.7	7.6
n_c	5.4	5.7	6.8

$(\partial \ln \dot{\gamma}_0/\partial \ln \tau) = 2$, and from Fig. 2, $(\partial \ln A^*/\partial \ln \tau) = -\frac{1}{4}$; then

$$n = 2 + \frac{3}{4}\left(\frac{A^* b \tau_e}{kT}\right) \quad (11)$$

Table I gives the values of n calculated by means of Equation 11 using the experimental data of A^* and τ_e. Table I also gives n_c values calculated from the slope of $\log \dot{\gamma} - \log \tau$ data as shown in Fig. 1. As can be seen from the table the agreement is satisfactory. As it is established that thermally-activated glide motion of dislocations can predict satisfactorily the Q_c and n values of the creep data, we have attempted to calculate the creep rate using Equation 8 for $\Delta F = 0.5Gb^3$, $\beta \simeq 1$ and $\nu = 8.71 \times 10^{12}$ sec^{-1}. The estimated values of $\dot{\gamma}$ are plotted in Fig. 1 as a function of stress (dotted line); these values are a factor of 3 below the experimental ones. This consistency between the experimental creep data of the alloy at high stresses and the prediction of the thermally-activated dislocation glide mechanism indicates that the agreement obtained is not fortuitous.

Finally, if a limited number of data points is considered in the high stress region, then they can be connected by a straight line having a slope of ~ 5 and hence the high stress region may be interpreted as controlled by dislocation climb. Therefore determination of, Q_c, A^*, τ_e and other creep parameters will help in unambiguously identifying the creep controlling mechanism.

3. Conclusions

The present analysis shows that high-stress creep data of the Al–10 at % Zn alloy, whose creep behaviour is controlled by viscous glide motion of dislocations at intermediate stresses, seems to correlate well with the prediction of thermally-activated dislocation glide. This correlation is manifested by the good agreement between the experimental parameters of the alloy and their values predicted by using the rate equation of the controlling mechanism.

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