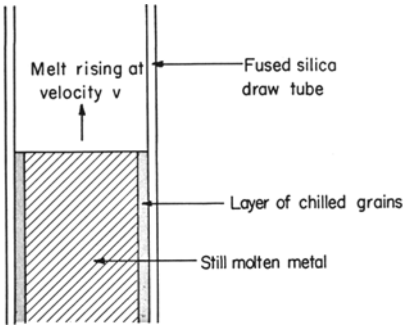

Fig. 3—A schematic diagram of the melt rising in the fused silica tube.

which raised the temperature of the silica above T_A, the ambient temperature, according to

$$q_{\text{silica}} = 2\pi r d' c' \rho' \left(\frac{T_m + T_s}{2} - T_A\right) \qquad [6]$$

and that heat is the total heat transferred into the silica from the superheated melt which has passed by it at velocity v during the filling of the tube to a height H, *i.e.*,

$$q_{\text{silica}} = (\dot{q}_{\text{melt}} - \dot{q}_{\text{radiation}}) \frac{H - h}{v} \qquad [7]$$

in which both $\dot{q}_{\text{melt}}$ and $\dot{q}_{\text{radiation}}$ are taken to be constant during filling of the tube. In order to evaluate $\dot{q}_{\text{melt}}$ after the tube has been filled, we use a tabulated solution, f (and its derivative f') of the diffusion equation for a cylindrical specimen:[2]

$$\frac{T_0 - \overline{T}}{T_0 - T_m} = f\left(\frac{\alpha t}{r^2}\right) \qquad [8]$$

where $\alpha = k/\rho c$ is the thermal diffusivity of the melt. Because we are concerned with the melt at height h, f is to be evaluated for the time since the melt at that height has been cooling, *i.e.*, since it left the reservoir, and that time is $t = h/v$.

Combining the above equations in a straight forward manner leads to the following equation in which each term is dimensionless:

$$\left(\frac{T_0 - T_m}{T_m - T_A}\right) f'\left(\frac{\alpha h}{v r^2}\right) = \frac{2 r h_r}{k} + \frac{4\left(\frac{r}{d'}\right)\left(\frac{k'}{k}\right)}{1 + \frac{2\alpha'}{d'^2}\left(\frac{H-h}{v}\right)}. \qquad [9]$$

The factor $\left(\frac{T_0 - T_m}{T_m - T_A}\right)$ may be regarded as a fractional superheat. In solving this equation T_m, T_A, r, d', h_r, ϵ, k, k', α, and α' are known,[3-5] values for T_0, H, and v are assumed, and h is calculated.

The results of these calculations (which must be done in an iterative fashion since f, and hence its derivative f', is a tabulated rather than an analytical function) are given in Fig. 4. All of the curves are calculated values of the chill layer remelt height for a velocity of 20 cm/s, which is a reasonable value based on observation. Superimposed on the calculated curves are the observed heights at which large crystals were observed from several runs, with the superheats given in each case. It is seen that despite some simplifying assumptions the agreement between calculated curves and the observed points is good, especially considering that there are no arbitrary constants involved.

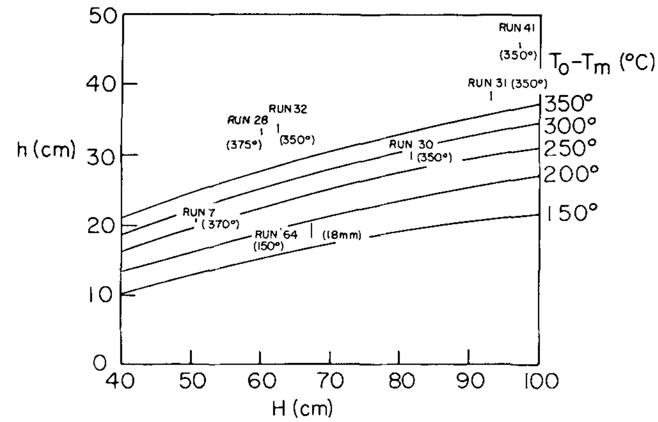

Fig. 4—Calculated curves of h *vs* H for different superheats $(T_0 - T_m)$, as marked, with experimental data (superheats given in each case). Calculated curves are for v = 20 cm/s.

The rate of solidification was not observed directly. There is some limited information by which the solidification rate may be deduced from the observed grain size, but it is not applicable to alloys of the composition we were considering.[6]

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Evidence for Subgrain Formation in an Al-Mg Alloy at Low Stresses

FARGHALLI A. MOHAMED

Creep characteristics of solid-solution alloys[1-5] indicate that substructures developed in alloys of the metal class (stress exponent ≃5) and those of the alloy class (stress exponent = 3) are vastly different. In alloys of the metal class, a regular array of subgrains is formed, and the substructure observed is, therefore, similar to that of pure metals.[2,5] On the other hand, substructures observed in alloys of the alloy class exhibit two features: a) subgrains are not generally developed, or, if formed, are found in the vicinity of grain boundaries, and b) dislocations are wavy and homogeneously distributed.

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Recently, it has been shown[6-9] that the stress exponent in some alloys changes from a value of 3 (typical of the alloy class) at high stresses to a value close to 5 (typical of the metal class) at low stresses. This change, which is in accord with earlier predictions,[2,3] suggests that creep features reported for both metal and alloy classes can be observed in a single solid-solution alloy, provided that a wide range of stresses is scanned. Consequently, it is anticipated that if an alloy of the alloy class is crept at sufficiently low stresses, the substructure developed will be typical of the metal class, *i.e.*, a tendency to form well developed subgrains. This possibility has motivated the work reported in this paper on the creep behavior of an Al-Mg alloy.

An Al-2 pct Mg alloy was chosen for two reasons. First, high-stress substructural data[10-12] of the Al-Mg system clearly show that this system behaves as an alloy type. Second, mechanical testing of Al-Mg alloys presents no difficulties since the tests are conducted in air.

Double shear specimens of Al-2 pct Mg were machined and tested on a suitably designed creep machine.[6,7] Although the general test procedure employed in the present study has been described elsewhere,[6,7,13] it is important to mention these points:

a) Specimens were annealed for 10 h at 888 K, giving a grain size of about 4 mm.

b) The range of stresses from 0.2 MN/m^2 to 3 MN/m^2 was investigated at a single temperature of 878 K.

c) After deformation, specimens were cooled under load to preserve the substructure developed.

d) X-ray back reflection and etch-pit techniques were used to examine the possibility of subgrain formation at low stresses.

Figure 1 shows the data of creep tests on Al-2 pct Mg at T = 878 K, where the shear strain rate, $\dot{\gamma}$, is plotted logarithmically as a function of shear stress, τ. At the higher stresses ($\tau > 0.8$ MN/m^2, the stress exponent is 3 and the creep curves exhibit brief normal transient. At the lower stresses ($\tau < 0.8$ MN/m^2), the stress exponent is 4.5 and an extensive primary stage was observed.

Weertman[14] suggested that the creep behavior of a solid-solution alloy may be governed by the two sequential processes of viscous glide (which obeys a third-power law) and dislocation climb (which obeys a fifth-power law). On the basis of this suggestion, it was predicted[2,3] that the creep behavior of a solid-solution alloy would change from that of the alloy class to that of the metal class as the stress is reduced below a certain critical value. A recent investigation[6,7] on Al-3 pct Mg verified this prediction. More recently, Mohamed *et al*[4,9,15] showed that the transition from metal class to alloy class is consistent with a criterion of the form

$$\left(\frac{kT}{ec^{1/2}G\mathbf{b}^3}\right)^2 = B\left(\frac{\gamma}{G\mathbf{b}}\right)^3 \frac{D_c}{D_g}\left(\frac{\sigma}{G}\right)^2, \qquad [1]$$

where k is Boltzmann's constant, T is the absolute temperature, e is the solute-solvent size difference, c is the concentration of solute atoms, $\mathbf{b}$ is the Burgers vector, G is the shear modulus, B is a dimensionless constant, γ is the stacking-fault energy, D_c is the diffusion coefficient for the climb process, D_g is the diffusion coefficient for the glide process, and σ/G is the normalized stress ($\sigma/G = 2\ \tau/G$).

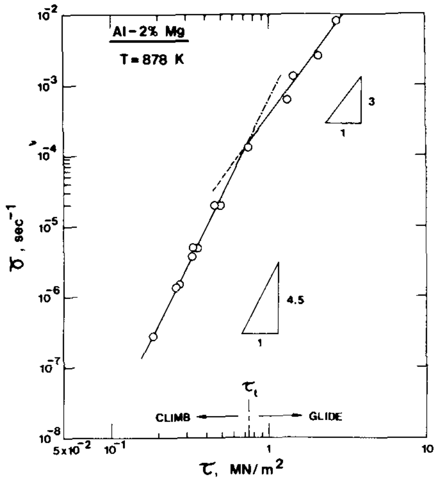


Fig. 1—Creep data of an Al-2 pct Mg alloy at T = 878 K.

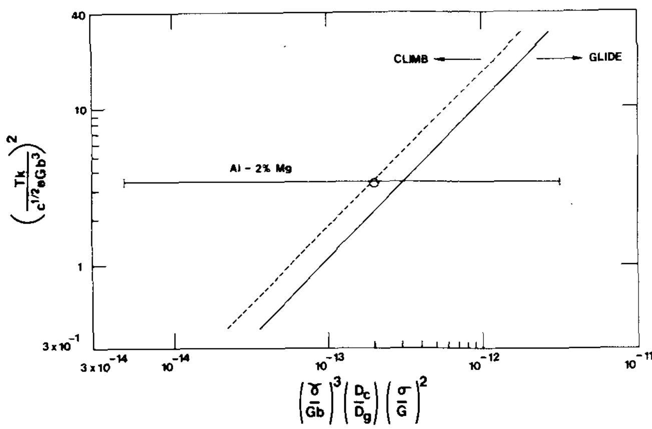
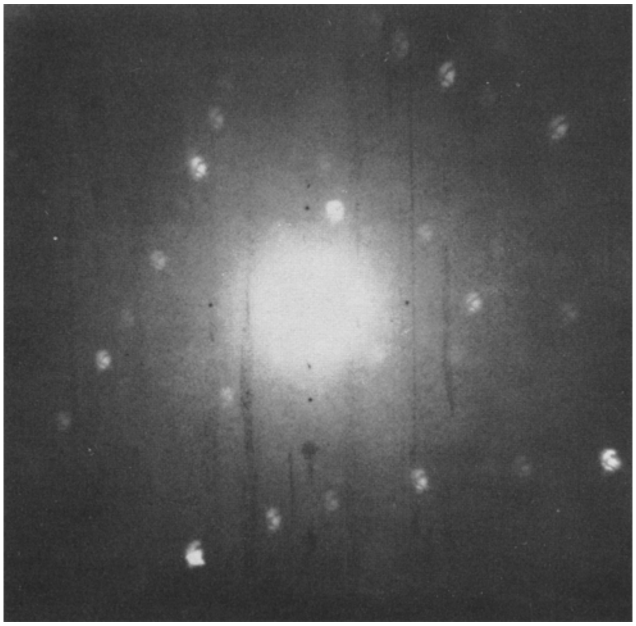


Fig. 2—Correlation between the normalized glide-climb criterion and data of an Al-2 pct Mg alloy at T = 878 K.

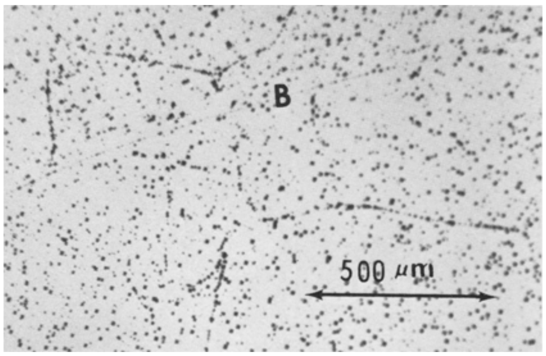
Figure 2 shows a plot of $\left(\frac{kT}{ec^{1/2}\mathbf{b}^3G}\right)^2$ *vs* $\left(\frac{\gamma}{G\mathbf{b}}\right)^3 \frac{D_c}{D_g}\left(\frac{\sigma}{G}\right)^2$ on a logarithmic scale. The boundary between viscous glide and climb control is represented, according to Eq. [1], by a straight line at 45 deg. The solid and broken lines were determined from data on Al-3 pct Mg deformed at different temperatures[6,7] and a single temperature,[6,9] respectively. The difference in position between the two lines could be due to experimental scatter which occurs in the transition region.

The stress range employed and the transition point observed in the present experiments on Al-2 pct Mg are shown in Fig. 2. As can be seen, the experimental transition point, which is marked as a circle, agrees very well with that determined from the broken-line boundary.

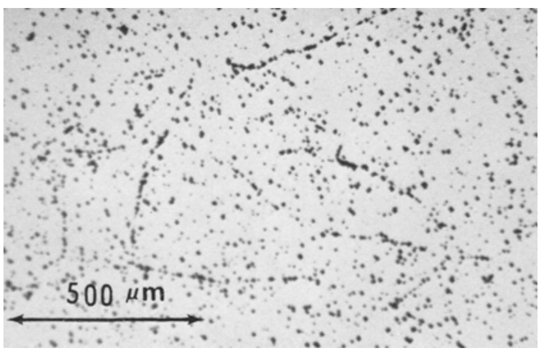
The first indication of subgrain formation at low stresses (stress exponent = 4.5) was obtained by the back-reflection Laue photograph shown in Fig. 3(*a*). The Laue spots are split, indicating subgrain formation within the grain. Confirmation of this observation is provided by photographs of etch pits obtained for a specimen tested at τ = 0.35 MN/m^2. Figure 3(*b*) shows polygonized regions in the vicinity of a grain boundary (marked as *B*), and Fig. 3(*c*) demonstrates that the



(a)



(b)



(c)

Fig. 3—(a) A back reflection Laue X-ray photograph of an Al-2 pct Mg specimen tested at $\tau = 0.35$ MN/m^2, (b) distribution of etch pits after creep testing at $\tau = 0.35$ MN/m^2. A grain boundary is shown at B, magnification 44 times, and (c) distribution of etch pits within the grain at $\tau = 0.35$ MN/m^2, magnification 44 times.

dislocation structure is also polygonized within the grain. However, it is clear that the subgrain size is inhomogeneous, *i.e.*, very fine and large subgrains could be found.

Dorn and coworkers[2,16] have concluded that whenever the climb mechanism is rate-controlling, subgrains are formed during creep. On the other hand, several systematic investigations[10-12] on Al-Mg alloys have indicated that under viscous-glide conditions (stress exponent = 3) there is no significant tendency to form subgrains. These experimental findings along with the present experimental evidence support the prediction that, as the stress is sufficiently reduced, creep substructures developed in a solid-solution alloy may change from those observed in the alloy class to those observed in the metal class.

In summary, i) creep data of Al-2 pct Mg show a transition in behavior from viscous glide control to climb control when the shear stress is reduced, and ii) there is a tendency to form sub-boundaries in the low-stress regime.

This work is the outgrowth of research inaugurated by the late Professor John E. Dorn to whom I am deeply indebted.

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