On the importance of the threshold stress for deformation

Theoretical analyses along with experimental evidence indicate the occurrence of two Newtonian processes, Nabarro–Herring creep [1, 2] and Harper–Dorn creep [3–6], at very low stresses and at temperatures near the melting point. Nabarro–Herring creep is produced by vacancy flow, and its steady-state rate, if boundaries are perfect sources and sinks, are given by

$$\dot{\epsilon}_{\mathrm{NH}} = A_{\mathrm{NH}} \frac{DGb}{kT} \left(\frac{b}{d}\right)^2 \left(\frac{\sigma}{G}\right), \tag{1}$$

where A_{NH} is a constant, D is the lattice diffusivity, G is the shear modulus, $\boldsymbol{b}$ is the Burgers vector, k is Boltzmann's constant, T is the absolute temperature, d is the grain size, and σ is the applied stress. Harper–Dorn creep, on the other hand, is produced by an unknown dislocation mechanism, and its rate, based on analyses [4, 5] of creep data of bulk samples of several materials, is given by

$$\dot{\epsilon}_{\mathrm{HD}} = A_{\mathrm{HD}} \frac{DGb}{kT} \left(\frac{\sigma}{G}\right), \tag{2}$$

where A_{HD} is a dimensionless constant.

One of the basic characteristics of the Newtonian behaviour represented by Equations 1 and 2 is that all the applied stress is responsible for the measured creep rate. However, a number of experimental observations, supported by some theoretical considerations, have suggested the existence of a modified form of the Newtonian behaviour, in which the effective stress, σ_{e}, is given by

$$\sigma_{\mathrm{e}} = \sigma - \sigma_0, \tag{3}$$

where σ_0 is a threshold stress required for the onset of deformation, i.e. no creep occurs at $\sigma = \sigma_0$. In this communication, sources considered responsible for σ_0 are summarized and the importance of knowing σ_0 is examined using low-stress creep data of Al.

Several different sources* of σ_0 have been suggested, including surface tension [7], oxidation effects [8], presence of particles at grain boundaries [9, 10], and inefficiency of grain boundaries as vacancy sources and sinks [11]. The characteristics of σ_0 arising from each source are listed in Table I. As indicated in this table, measurements of σ_0 require either extrapolation, when $\dot{\epsilon} = 0$ for $\sigma \leqslant \sigma_0$, or interpolation, when $\dot{\epsilon}$ changes from positive to negative values. In either case, highly sensitive equipment must be used to monitor very slow creep rates of the order of 10^{-9} or $10^{-10}\,\mathrm{sec}^{-1}$.

A common method of presenting creep data is to plot the logarithm of the applied stress against the logarithm of the creep rate. At high stresses,

TABLE I Sources of σ_0 and their characteristics

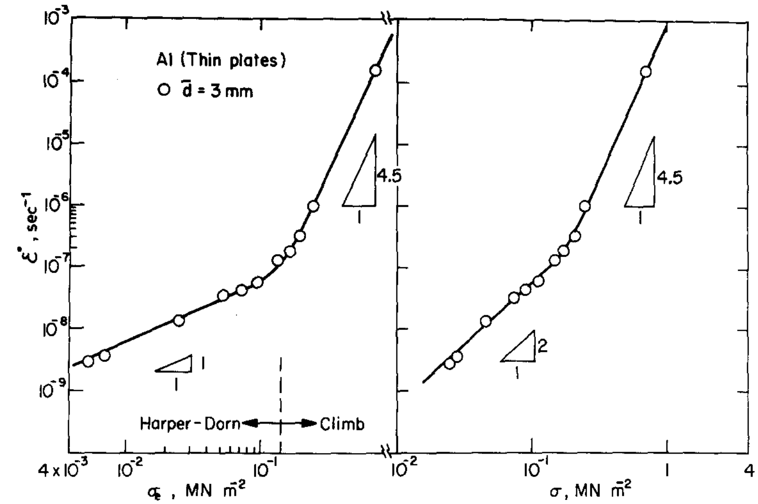
Source	Characteristics			Ref.
(1) Surface tension	$\sigma_0 = \frac{\gamma}{R}\left[1 - \frac{R}{3d}\right]$,	$\sigma < \sigma_0$	$\dot{\epsilon} < 0$	[7]
(2) Oxidation				
(i) vacancy blocking	σ_0 is independent of t,	$\sigma \leqslant \sigma_0$	$\dot{\epsilon} = 0$	
(ii) vacancy injection		$\sigma < \sigma_0$	$\dot{\epsilon} < 0$	[8]
(iii) physical strength of the oxide	$\sigma_0 = 2\lambda\sigma_{\mathrm{f}}/t$,	$\sigma \leqslant \sigma_0$	$\dot{\epsilon} = 0$	
(3) Particles at grain boundaries				
(i) by-passing of dislocation	$\sigma_0 = \frac{2\Gamma}{bL}$			[9]
		$\sigma \leqslant \sigma_0$	$\dot{\epsilon} = 0$	
(ii) punching dislocation loops	$\sigma = \frac{2\Gamma V}{br}$			[10]
(4) Inefficiency of grain boundaries		$\sigma \leqslant \sigma_0$	$\dot{\epsilon} = 0$	[11]

γ = surface energy, R = radius of a tensile wire specimen, λ = the thickness of the oxide layer, t is the thickness of the specimen, σ_{f} = the fracture stress of the oxide, Γ = the line energy of boundary dislocation, L = the particle spacing, V = the volume fraction of the particles, r = the radius of the particle, and d = the grain size.

* Grain growth at very low stresses may lead to a false value of σ_0.

0022–2461/80/041061–03$02.30/0

Figure 1 Creep rate, $\dot{\epsilon}$, versus the applied stress, σ (right-hand side) and the effective stress, σ_e, (left-hand side) for coarse-grained aluminium plates. Data are taken from [3].



such plots normally yield straight lines, with their slopes representing the values of the stress exponent, n. The value of n is then used as a parameter, together with other parameters such as activation energies, to identify the rate-controlling mechanism. However, this procedure could provide misleading information on the actual process at low stresses if the material exhibited a modified Newtonian behaviour due to the operation of one of the sources responsible for σ_0; a plot of log $\dot{\epsilon}$ versus log σ would then not reflect the Newtonian behaviour. Two examples of this situation are provided below, using low-stress creep data of Al.

Harper and Dorn [3] investigated the low-stress creep behaviour of Al at a temperature close to the melting point; their data are plotted as the applied stress, σ, versus creep rate on a logarithmic scale in Fig. 1 (right-hand side). This plot shows the presence of two regions of deformation, high stress and low stress. The high-stress region exhibits a stress exponent of about 4.5, which is typical of behaviour attributable to dislocation climb [12]. In the low-stress region, the datum points fall very close to a straight line with a slope ~ 2, suggesting the presence of a deformation process having $n = 2$. However, Harper and Dorn [3] reported a threshold stress, σ_0, that they attributed to surface tension. Using the value of σ_0, their data are replotted as $\dot{\epsilon}$ versus σ_e in Fig. 1 (left-hand side). The stress exponent, n, of the high-stress region remains unchanged, but the data of the low-stress region are well represented by a modified Newtonian behaviour in which n is equal to 1 instead of 2.

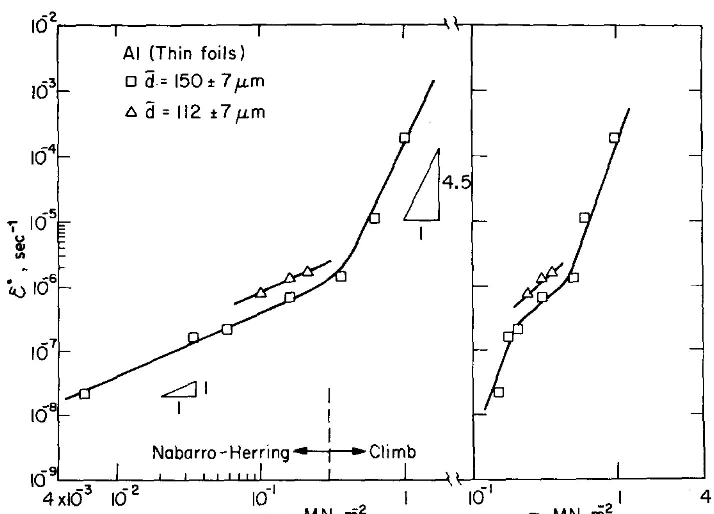


Figure 2 Creep rate, $\dot{\epsilon}$, versus the applied stress, σ (right-hand side) and the effective stress, σ_e (left-hand side) for fine-grained aluminium thin foils. Data are taken from [8].

The low-stress creep behaviour of fine-grained thin foils of Al was investigated [8] recently, and the data indicated the presence of a threshold stress. If the datum points for two grain sizes are plotted as log $\dot{\epsilon}$ versus log σ in Fig. 2 (right-hand side), a sigmoidal behaviour, similar to that reported for superplastic alloys, result. On the other hand, when the data are replotted in terms of the effective stress rather than the applied stress (left-hand side of Fig. 2), the creep behaviour at low stresses exhibits a modified form of diffusional creep, i.e., $\dot{\epsilon} \propto \sigma_e/d^2$.

In conclusion, the present analysis, using low-stress creep data of Al, shows that a plot of log $\dot{\epsilon}$ versus log σ can lead to an incorrect identification of the actual creep process if the creep behaviour exhibits a threshold stress.

Acknowledgement
This work was supported by the National Science Foundation under grant no. DMR 77-27619.

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Received 3 August
and accepted 20 September 1979

FARGHALLI A. MOHAMED
Departments of Materials Science
and Mechanical Engineering,
University of Southern California,
Los Angeles,
California,
USA

Plastic deformation in CaO-stabilized ZrO_2 (CSZ)

The recent availability [1] of high quality calcia-stabilized zirconia (CSZ) single crystals has prompted study of their deformation behaviour. CSZ is isostructural with CaF_2 and UO_2 and we show in this preliminary communication that it has the same preferred slip system as these compounds.

The crystal studied contained 12 mol% CaO and was optically transparent. Compression specimens with an orientation shown in the inset of Fig. 1 were prepared using standard ceramic techniques. This orientation provided approximately equal Schmid factors of 0.44 for (1 1 1) [$\bar{1}$ 1 0] and (0 0 1) [$\bar{1}$ 1 0] slip, and a somewhat lower Schmid factor of 0.29 for (1 1 0) [$\bar{1}$ 1 0] slip.

Deformation experiments were carried out in air at temperatures of 1350 and 1450° C at $\dot{\epsilon} \sim 1 \times 10^{-4}$ sec^{-1}. (At this strain rate, the brittle–ductile transition is between 1200 and 1350° C, as samples failed by brittle fracture when tested at 1200° C.) Samples tested at higher temperatures exhibited considerable ductility (Fig. 1) with a very low work-hardening rate of $\mu/660$, where μ is the shear modulus. As will be shown below, considerable climb is occurring during deformation and the resulting recovery must be responsible for the absence of appreciable work hardening.

Both etch-pit techniques (Fig. 2) and transmitted polarized light examination were used to perform two-surface trace analysis, and (0 0 1) was determined as the preferred slip plane. (One of the side faces of the specimens was cut parallel to (1 1 1); molten KOH at 350° C was found to be an effective dislocation etchant.)

Transmission electron microscopy (TEM) foils were prepared by ion-thinning. A typical dis-

0022–2461/80/041063–03$02.30/0