

In this equation ϕ is a factor that allows for the orientation of the active slip system, b is the Burgers vector, N is the dislocation density, and v is the mean dislocation velocity. The second form of the equation is found experimentally to represent well the plastic deformation curves for many crystals. The coefficient M measures the rate at which dislocations multiply as deformation proceeds, D is a measure of dislocation mobility in the undeformed crystal, and W allows for work hardening. Because dislocations overcome obstacles on the glide plane with the aid of thermal energy

$$v = v_0 e^{-(D+W\epsilon)/\sigma} = e^{-G^*/kT} \quad (3)$$

showing that the activation free energy G^* is stress dependent for dislocation motion.

Consider now how the strain rate is altered on application of hydrostatic pressure. The term $N_0 + M\epsilon$ in equation 2 is hardly changed even at the pressures of the earth's deep interior. The term remains almost unchanged because the formation energy of a dislocation E_f is large and the volume of formation V_f is small, so that $PV_f \ll E_f$. The influence of pressure on strain rate is through the velocity term in equation 2. The thermodynamic quantity describing this dependence is V_d^* , whose definition is seen from equation 4 [Li, 1967]

$$\begin{aligned} d \ln v &= \left(\frac{\partial \ln v}{\partial \sigma} \right)_{T, P, \epsilon} d\sigma + \left(\frac{\partial \ln v}{\partial T} \right)_{P, \sigma, \epsilon} dT \\ &+ \left(\frac{\partial \ln v}{\partial P} \right)_{\sigma, \epsilon, T} dP + \left(\frac{\partial \ln v}{\partial \epsilon} \right)_{\sigma, P, T} d\epsilon \\ &= \frac{D + W\epsilon}{\sigma^2} d\sigma + \frac{H^*}{kT^2} dT - \frac{V^*}{kT} dP - \frac{W}{\sigma} d\epsilon \end{aligned} \quad (4)$$

to be

$$V^* = kT \left(\frac{\partial \ln v}{\partial P} \right)_{\sigma, \epsilon, T} \quad (5)$$

Equation 5 specifies that V_d^* is to be measured by finding the change in dislocation velocity with pressure in an experiment done at constant stress, strain, and temperature. Observations of this kind can be made in experiments in which the motion of individual dislocations is observed directly, as is done, for example, when dislocation intersections with a free sur-

face are identified by etch pits. The crystal to be studied is subjected to a given deforming stress for a given time with and without hydrostatic confining pressure applied. The average displacements of mobile dislocations are observed by successive etching and from these data the average dislocation velocities are calculated. Since the total dislocation displacement remains small, ϵ is effectively constant during the experiment.

A set of direct dislocation velocity measurements is shown in Figure 1. The scatter in the data is such that $\partial \ln v / \partial P$ is not detected. The scatter arises because dislocations do not advance under a given shear stress at a uniform velocity. They move instead by a series of jumps (in which v is nearly the sound wave velocity) separated by periods of rest. Successive jumps are

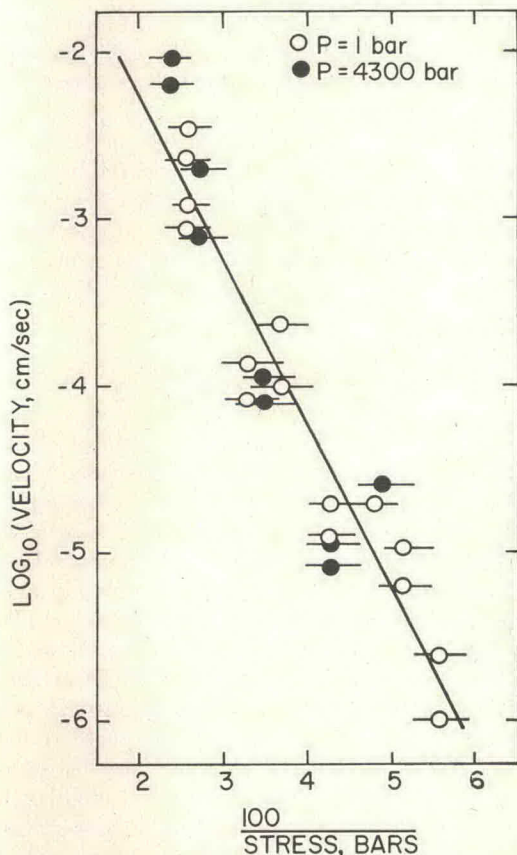


Fig. 1. Results of direct measurements of the velocity of screw dislocations in LiF single crystals with and without applied hydrostatic pressure [Haworth, 1969].

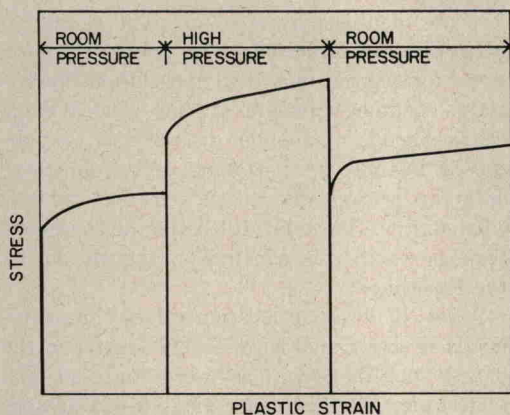


Fig. 2. Method of finding the pressure dependence of the plastic flow stress of a crystal by mechanical testing.

thermally activated and occur at random. The number of displacements that must be averaged to get a reliable mean velocity is so large that the method of direct observation is impractical for determination of $\partial \ln v / \partial P$.

When the strain resulting from plastic glide on a particular slip system is observed, the result of the motion of many individual dislocations is automatically averaged. The activation volume for dislocation movement in a macroscopic stress-strain experiment can then be found from

$$V^* = kT \left(\frac{\partial \ln \dot{\epsilon}'}{\partial \ln \sigma} \right)_{T,P,N} \left(\frac{\partial \ln \sigma}{\partial P} \right)_{T,v,N} \quad (6)$$

The first derivative represents a room pressure determination of the strain rate sensitivity of the plastic flow stress. The second is measured by the increment in stress required to keep a crystal deforming at constant dislocation velocity when hydrostatic pressure is suddenly applied. The appropriate experimental method for determining this value is illustrated in Figure 2. The deforming stress and plastic strain are measured while the crystal is subjected to repeated small deformations. When there is no change in environmental conditions between successive deformations (and there are no large recovery effects), nearly continuous stress-strain curve is generated. When hydrostatic pressure is applied during some of the deformations, a different stress is required to continue the plastic flow (as shown in Figure 2); release of the pres-

sure results in a reversal of the incremental stress.

That a good approximation to the idealized experiment described above can be achieved in practice is shown by the data displayed in Figure 3 for the deformation in compression of a single crystal of CsBr [Davis and Gordon, 1968]. Successive applications of stress at constant pressure generate a smooth σ - ϵ curve and the application of hydrostatic pressure results in a reversible increment in flow stress. The conditions that v and N be constant are met because the stress increment $\delta\sigma$ is measured at a given strain and because the experiment is done in such a way that the rate of plastic deformation is the same at both pressures.

Experiments of the type illustrated in Figure 3 have been performed on a wide variety of crystalline materials (unpublished data obtained by E. Aladag, W. L. Haworth, and L. A. Davis, 1970). The results obtained on metals, alloys, and covalently bonded compounds (selenides) can be summarized as follows. In these materials the fractional increment of plastic flow stress $\delta\sigma/\sigma$ is always less than or equal to the fractional increment of the shear modulus $\delta\mu/\mu$ that results from the application of hydrostatic pressure. (The shear modulus to be used is the elastic coefficient for the type of dislocation under study.) This means that $V_d^* \ll V_a$, that dislocation motion is very much less pressure sensitive than thermally activated atom movements. The active deformation mechanism is such that the strength of the above materials arises from the elastic interactions between dislocations, or between dislocations and point de-

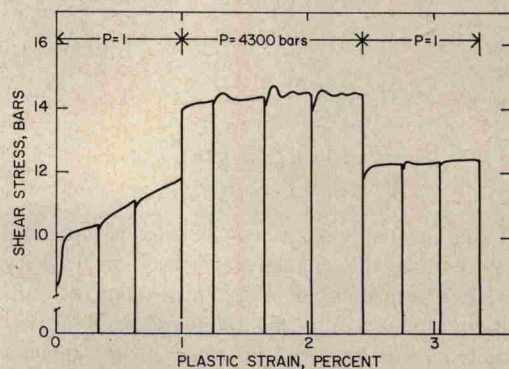


Fig. 3. Compression tests on a single crystal of CsBr with and without applied hydrostatic pressure [Davis and Gordon, 1968].