The mechanism of shear of an amorphous metal

For the shear deformation of a metallic glass, Gilman [1, 2] has suggested a model for the yield stress involving the glide motion of dislocations. These defects are supposed to have variable Burgers vectors along their lines, but to have a mean Burgers vector $\mathbf{b}$ equal to the mean atom spacing. The resistance to glide is assumed to be provided by the elastic resistance to a local dilatation, normal to the glide plane, which accompanies glide motion; somewhat analogous to shear models in soil mechanics. By balancing the work done by the resolved shear stress τ per unit length of line per increment of advance $\mathbf{b}$, $\tau \mathbf{b}^2$, with the work performed in creating the normal dilatation, he obtains an expression for σ. For the latter case, the work is assumed to be one-third of the work of formation of a spherical point defect, as modelled by a ball in a hole [3–5]. Thus, he obtains the expression for tension or compression, where the yield stress $\sigma = 2\tau$,

$$\sigma = 8\pi\epsilon^2 B/(1 + 3B/4\mu), \tag{1}$$

where ϵ is the strain normal to the glide plane, B is the bulk modulus and μ is the shear modulus.

Here, we consider several different cases, which give results for σ close to, but somewhat different from, Equation 1. If the dislocation moves as a line and creates a cylindrical dilatation field with strength $\delta A/A = 2\epsilon$, where A is the area of the core region, then a 2-dimensional analog of the 3-dimensional result of Eshelby [3–5], including both the work done on the surroundings by the core and strain energy stored in the core, gives a work term

$$W_1 = A(2\epsilon^2\mu + 2\epsilon P)(4\mu + 3B)/(\mu + 3B), \tag{2}$$

with a corresponding shear work term τA. Here, the work of interaction $2P\delta A$ of the defect and an external pressure P has been included. Alternatively, akin to the Gilman model, one can suppose that only strain normal to the glide plane is produced. Then the work term, again including strain energy stores in the core, is

$$W_2 \cong A(\epsilon^2\mu - \epsilon\sigma_N)(10\mu + 3B)/(4\mu + 3B), \tag{3}$$

with σ_N the normal stress. Spaepen and Turnbull [6] have suggested a model, not involving the dislocation concept, in which shear is accompanied by a local dilatation at the elastic–plastic boundary. This would lead to an expression such as Equation 2, but with A' and ϵ' now referring to the new boundary, rather than to the core, and with an added term corresponding to the dissipated plastic work per unit volume, W_P

$$W_3 = A'(2\epsilon'^2\mu + 2\epsilon' P)(4\mu + 3B)/(\mu + 3B) + W_P. \tag{4}$$

Another type of dislocation model would be to assume that motion occurs by lateral motion of kinks along a dislocation with work $\tau \mathbf{b}^3$ now balanced by a local point-type strain field. If the local field is solely a strain normal to the glide plane, the work term would be $W = W_2 \mathbf{b}$, which, when balanced with the shear work would give the same result as Equation 3; hence, this model need not be considered explicitly. Alternatively, the kink displacement could be a ball-in-hole-type dilatation which would give

$$W_4 = V(6\epsilon^2\mu + 3\epsilon P)(4\mu + 3B)/3B. \tag{5}$$

Performing the appropriate work balances, the above expressions give corresponding predictions of the yield stress in tension or compression with or without superposed hydrostatic pressure. For this balance we assume core dimensions $A = \mathbf{b}^2$ and $V = \mathbf{b}^3$, giving

$$\sigma_1 = (4\epsilon^2\mu + 4\epsilon P)(4\mu + 3B)/(\mu + 3B) \tag{6}$$

$$\sigma_2 = (2\epsilon^2\mu - 2\epsilon\sigma_N)(10\mu + 3B)/(4\mu + 3B) \tag{7}$$

$$\sigma_3 = (4\epsilon'^2\mu + 4\epsilon' P)(4\mu + 3B)/(\mu + 3B) + W_P \tag{8}$$

$$\sigma_4 = (12\epsilon^2\mu + 6\epsilon P)(4\mu + 3B)/3B. \tag{9}$$

For purposes of correlation with experimentally observed yield strengths the interaction work terms are negligible, but for the strength-differential (SD) effect, discussed next, they are important.

Models of the above type give specific predictions of the SD effect wherein the flow stress varies in general with the total applied stress tensor or, in more limited cases, with superposed hydrostatic stress. The relevant SD expressions for the above forms are

$$(\partial\sigma/\partial P)_1 = [4\epsilon^2(\partial\mu/\partial P) + 4](4\mu + 3B)/(\mu + 3B) \tag{10}$$

$$(\partial\sigma/\partial\sigma_N)_2 = [-2\epsilon^2(\partial\mu/\partial P) - 2\epsilon](10\mu + 3B)/(4\mu + 3B) \qquad (11)$$

$$(\partial\sigma/\partial P)_3 = [4\epsilon'^2(\partial\mu/\partial P) + 4\epsilon'](4\mu + 3B)/(\mu + 3B) \qquad (12)$$

$$(\partial\sigma/\partial P)_4 = [12\epsilon^2(\partial\mu/\partial P) + 6\epsilon](4\mu + 3B)/3B \qquad (13)$$

Experimentally, the SD has been determined for $Pd_{78}Si_{16}Cu_6$ both by determining [7] the variation of flow stress with superposed hydrostatic pressure of 6.9×10^9 Pa and by comparing [7, 8] the flow stress in compression, 1.54×10^9 Pa, and in tension, 1.44×10^9 Pa. The direct variation with P gives $(\partial\sigma/\partial P) = 0.08$. The variation in tension and compression gives $(\partial\sigma/\partial\sigma_N) = -0.07$ [8] and -0.08 [7]. If the variation with σ_N were to reflect a true variation with P, since $\sigma_N = \sigma/2 = -3P/2$, the tension–compression tests would give $(\partial\sigma/\partial P) = 0.10$ [8] and 0.11 [9]. On the other hand, if the variation with actual superposed P reflected a true variation with σ_N, the pressure experiments would give $(\partial\sigma/\partial\sigma_N) = -0.08$, since then $\sigma_N = -P$. Thus, the results to date for the SD effect alone strongly support the σ_N dependent model dependence of Equation 11 as opposed to the P-dependent models. Nevertheless, for comparison of correlations between the SD results and the flow stress results, the value $(\partial\sigma/\partial\sigma_N) = -0.08$ is used for Equation 11, and the mean value $(\partial\sigma/\partial P) = 0.09$ for Equations 10, 12 and 13. The use of $(\partial\sigma/\partial P) = 0.08$ instead in the latter case would give a decrease of about 5% in the values of ϵ computed from Equations 10, 12 or 13 to match the experimental SD results.

Pressure dependences of μ and B have not been measured for metallic glasses, so the correlation among Equations 6 to 9 and Equations 10 to 13 cannot be evaluated directly on the basis of independent experimental data. The correlation can be tested indirectly to assess how closely the necessary values for the pressure dependence of the elastic constants conform to the typical values [7] for metals of $\partial \ln B/\partial P \cong 5 \times 10^{-11}$ Pa^{-1}. Data necessary for the correlation, in addition to those listed above, are [7, 9]: $B = 1.67 \times 10^{11}$ Pa, $\mu = 3.15 \times 10^{10}$ Pa, $(\partial B/\partial P) = -(\partial B/\partial\sigma_N) = 8.35$, $(\partial\mu/\partial P) = -(\partial\mu/\partial\sigma_N) = 1.58$, with the pressure dependences conforming to the typical metal values.

TABLE I Values of normal strain ϵ required in Equations 6 to 9 to fit experimental data for σ and corresponding values of $(\partial \ln B/\partial P)$ computed from Equations 10 to 13.

Equation	ϵ to match σ	$\partial \ln B/\partial P$ (Pa^{-1})
6, 8, 10 and 12	0.100	-3.11×10^{-0}
7 and 11	0.120	-1.96×10^{-0}
9 and 13	0.056	-2.23×10^{-10}

As a first correlation, we suppose that Equations 6 to 9 explain in turn the flow stresses, and we calculate the values of ϵ for such a fit in each case, as listed in Table I. These values of ϵ are then inserted into Equations 10 to 13, assuming only that $\partial \ln B/\partial P = \partial \ln \mu/\partial P$, and the requisite values of $(\partial \ln B/\partial P)$ to fit the SD data are computed in each case, as also listed in Table I. In this correlation, the σ_N-dependent model, Equations 7 and 11 give the best agreement with the expected value of $(\partial \ln B/\partial P) = 5 \times 10^{-11}$ Pa^{-1}, and the model of Equations 9 and 13 gives no agreement at all (a negative value of $\partial \ln B/\partial P$ is physically impossible). However, the agreement is not good even for the best case.

A second correlation is to assume the expected value of $(\partial \ln B/\partial P)$ and to calculate ϵ values required to fit the measured SD dependence for Equations 10 to 13 as listed in Table II. The fraction of the flow stress related to these strain values can then be computed from Equations 6 to 9, as also listed in Table II. The remainder of the flow stress is then supposed to be contributed by a plastic dissipative term with minimal pressure dependence. A model for such a term would be a local bond-breaking dislocation-core-like contribution to the flow resistance. Again the model of Equations 7 and 11 gives the best overall corre-

TABLE II Values of normal strain ϵ required in Equations 10 to 13 to fit experimental data for the SD effect and the corresponding fractional contribution, calculated from Equations 6 to 9, to the experimental flow stress values.

Equation	ϵ to match SD	Fraction of σ
6, 8, 10 and 12	0.019	0.036
7 and 11	0.029	0.058
9 and 13	0.012	0.043

lation. Moreover, it seems more plausible that part of the flow stress has a bond-breaking component, than that the pressure dependence of the elastic constants is abnormally low, so the latter correlation is deemed to be the best.

Attempts to modify Equation 7 by adding a viscous term $\propto \mu$ to the flow stress equation makes the fit worse, so such an alternative can be excluded.

In summary, the SD effect alone and a comparison of SD results and flow stress results favour a model of dilatation normal to the glide plane, rather than one of hydrostatic pressure as a contribution to the flow resistance. This result is consistent with a dislocation-type flow mechanism in metallic glasses as suggested by Gilman [1, 2]. However, an exact fit to experimental results also requires a major contribution of a dissipative-type term to the flow resistance. An improved test of the suggested correlation could be achieved if independent measurements of the pressure dependence of the elastic constants were made available for metallic glasses.

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A new chemical polish and the study of dislocation movement in lithium fluoride crystals

The extent and nature of the dislocation movement produced in rocksalt-type crystals is well documented [1–3]. In particular, in magnesium oxide crystals the dislocation movement produced beneath localized regions of deformation caused by both indenters and sliders has been described recently [4]. The experimental technique used in that previous work involves repeated chemical polishing and etching of the deformed crystal in order to build up a cumulative picture of the dislocation pattern produced in the crystal bulk. In this way, the size and shape of the dislocated zone beneath a given indentation has been determined, and it has been established that the dimensions of this zone are largely defined by the magnitude and duration of the applied load, rather than by the shape of the indenter [4, 5]. A similar study on lithium fluoride crystals was initiated but, whilst a good dislocation etchant for this crystal is available, chemical polishing presented some problems. For the crystals used in this work, the polishing techniques and solutions used by Gilman and Johnston [3] and Borzhkovskaya *et al.* [6] produced surfaces which were too badly pitted to allow satisfactory application and interpretation of the subsequent dislocation etchant. As a result the following chemical polish was developed.

Single crystal specimens, (supplied by Rank Precision Industries, Margate and containing 50 ppm potassium, 20 ppm other elements) approximately 20 mm × 20 mm × 3 mm, were cleaved from large pieces of lithium fluoride; these were subsequently annealed in an argon atmosphere for six hours at 650° C and then slowly cooled to room temperature. Each crystal was washed in fresh water at 50° C and then vigorously agitated for 7 sec in fresh concentrated hydrochloric acid, also at 50° C, and then washed again