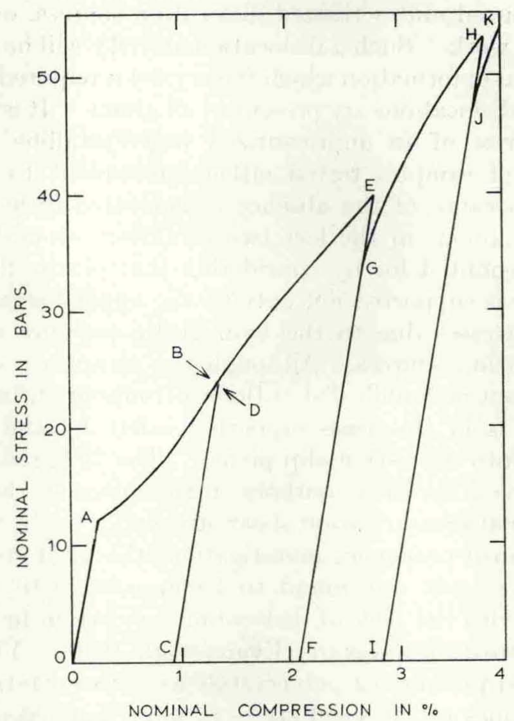
'seasoning' treatment at 10 kbar, given to all polycrystalline samples, may have introduced and activated dislocation sources, on the evidence from our present work. Such a dislocation activity will have an influence on any subsequent deformation which the crystal is required to undergo, in particular if free dislocations are present in all grains. It is to be expected that the flow stress of an unpressurized polycrystalline sample would be higher than of samples tested either subsequent to pressurization or at pressure, because of the absence of activated dislocation sources. Difference in behaviour in the last two conditions (reported by Aladag *et al.*) can be accounted for by considering that plastic flow in samples tested at 10 kbar is supported not only by the applied stress but also the inherent shear stresses due to the hydrostatic pressure near the grain boundary dislocation sources. Although the samples tested at atmospheric pressure contain mobile dislocations introduced during 'seasoning', the plastic flow is in this case supported solely by the applied shear stress resolved onto the active slip planes. The 20% reduction in yield stress, σ_Y, observed by these authors, may represent the contribution from the additional pressurization shear stresses.

In our mechanical properties investigations the yield stress in precipitate-free single crystals was found to be insensitive to pressurization, which is in line with the lack of dislocation generation in single crystals subjected to hydrostatic pressure (Evans *et al.* 1970). Yield stresses of unpressurized and pressurized polycrystals were also determined, but the scatter in the values of the yield stress of unpressurized samples (11–16 bar) precludes for the time being a quantitative evaluation of the effects of pressurization on as-annealed specimens. The confining pressure, however, was reported by Aladag *et al.* (1970) to influence the entire stress–strain relationship of a polycrystal; we suggest, *inter alia*, through the generation of dislocations. A quantitative test of this hypothesis, therefore, is the pressurization and compression of pre-compressed polycrystals. If fresh dislocations are generated by the high pressure treatment, the stress to re-initiate macroscopic plastic flow should be smaller than the stress reached in the test on an unpressurized sample.

These experiments were carried out and, for 10 kbar pressurizations, reductions in the flow stress of ~ 4 bar (20–40% σ_Y) were observed (e.g. fig. 2). By carrying out unload–reload experiments it was established that a real effect of pressurization was being studied. Pressurization of strained pure monocrystalline specimens did not result in a discernible effect on the flow stress.

An estimate of the number of new dislocations contributing to plastic flow in a pressurized polycrystal can be made using a modification of the model Johnston and Gilman (1959) proposed for the deformation of LiF monocrystals. Hahn (1962) and Cottrell (1963) extended the theory to the deformation of polycrystals of body-centred cubic transition metals and Mellor and Wronski (1970 a) recently used the same semiquantitative approach to interpret pressurization effects in polycrystalline chromium.

Fig. 2



The initial part (OAB) of the nominal stress–compression curve of a NaCl polycrystal tested at room temperature at a rate of $\sim 1{\cdot}8\times10^{-4}\,\mathrm{sec}^{-1}$. The specimen was then unloaded (BC), reloaded (CDE), unloaded (EF), pressurized at 10 kbar, reloaded (FGH), unloaded (HI), repressurized at 10 kbar and reloaded (IJK). All the compressions were carried out at atmospheric pressure.

All these workers incorporate the empirical formula for the mean dislocation velocity,

$$v=\left(\frac{\sigma}{\sigma_0}\right)^n \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad (1)$$

into the relation for plastic strain rate,

$$\dot\epsilon=\Phi b\rho_Y v_Y=\Phi b\rho_Y\left(\frac{\sigma_Y}{\sigma_0}\right)^n, \quad . \quad . \quad . \quad . \quad . \quad . \quad (2)$$

where σ is the tensile stress, Φ an orientation factor (0·5), b the Burgers vector, σ_0 and n constants and ρ_Y, v_Y and σ_Y the values of, respectively, the mean dislocation density, ρ, v and σ when macroscopic plastic flow commences. For the conditions: $b\approx4\times10^{-8}$ cm, $\sigma_Y\approx13$ bar (σ(A) of fig. 2), $\dot\epsilon\approx1{\cdot}8\times10^{-4}\,\mathrm{sec}^{-1}$ and (Gutmanas, Nadgornyi and Stepanov 1963) $\sigma_0\approx20$ bar, $n\approx8$, ρ_Y and v_Y evaluate to $\sim3\times10^5\,\mathrm{cm}^{-2}$ and $\sim3\times10^{-2}\,\mathrm{cm\,sec}^{-1}$, respectively.