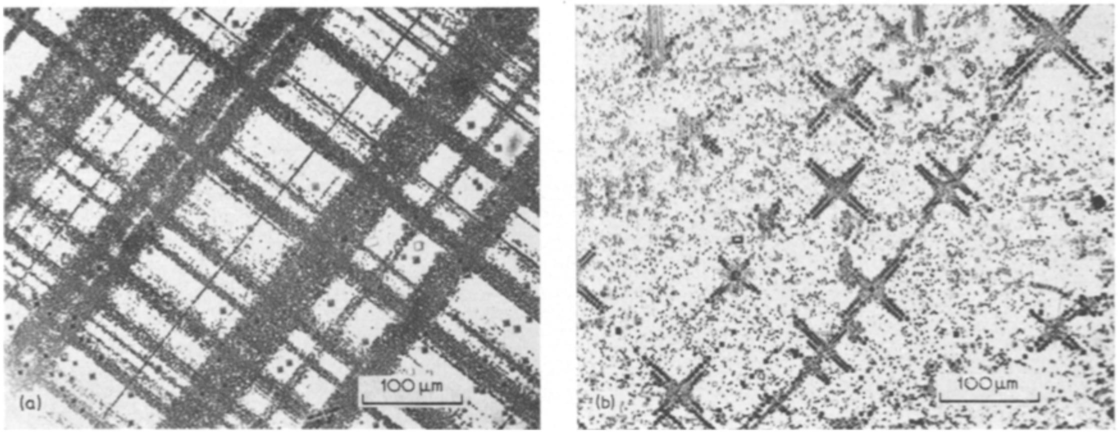

Figure 2 Comparison of the dislocation activity following 0.6% compressive strain in single crystal LiF irradiated and annealed to produce voids: (a) etched before and after straining and (b) etched before and after pressurization at 3.5 kbar and after straining.

⟨100⟩ direction, the arrays appear symmetrical about the cavity. There is a size dependant threshold pressure, below which no plastic activity is detectable, and above this threshold the extent of resultant arrays varies with pressure.

Preliminary compression testing of irradiated and annealed material, both unpressurized and pressurized, indicates that the pressure-induced dislocations suppress stage I [8] of the stress-strain curve and raise the flow stress and work-hardening rate. Etching of strained crystals has shown that pressurization inhibits the formation of well-defined slip bands characteristic of stage I (Fig. 2).

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The failure of polycrystalline chromium between 657 and 706 K

Controversy continues concerning the relative importance of crack nucleation and crack propagation in the fracture of body-centred cubic transition metals. In chromium in particular, some investigators have postulated that the critical event is always crack nucleation [1, 2] i.e. that the first crack to be nucleated propagates. Thus, it has been suggested [2] that the fracture of polycrystals deformed plastically in

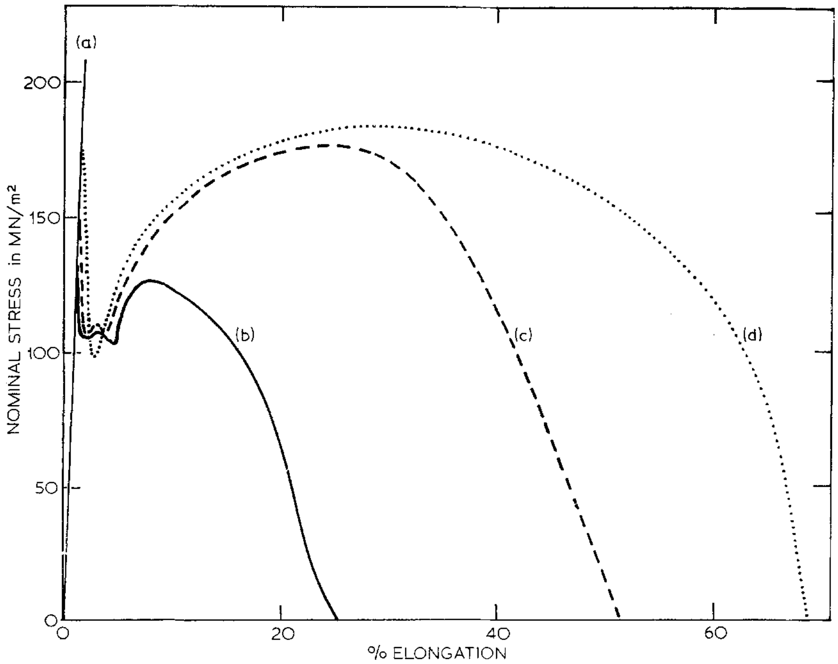

Figure 1 Load-elongation curves for polycrystalline chromium specimens pulled at (a) 657K, (b) 664K, (c) 688K and (d) 706K at a rate of $\sim 5 \times 10^{-4}$ sec^{-1}.

excess of 30% is nucleation-controlled. This hypothesis implies that crack propagation is easy in chromium when dislocations can move and would be expected to multiply near a crack tip to blunt it [3, 4]. Supporters of this model [1, 2] have never found non-propagating microcracks in fractured specimens; and this observation is consistent with their hypothesis.

In some recent studies of the ductile-brittle transition of polycrystalline cast chromium however [5, 6] microcracks were seen in tensile specimens which had undergone macroscopic plastic deformation before fracture. This observation indicates that crack propagation is more difficult than crack nucleation above the ductility transition temperature, T_T. In order to examine this problem thoroughly a detailed examination of the failure process of fine-grained (34 μm) material just above T_T was carried out and some unusual tensile failures were observed. The purpose of this communication is to report these observations, which unambiguously show that crack propagation can sometimes be exceptionally difficult in chromium.

Double-shouldered specimens with 10 mm gauge lengths and 2.1 mm gauge diameters were machined from arc-cast and extruded electrolytic flake chromium supplied by Fulmer Research Institute. The principal interstitial impurities were $\sim$ 500 ppm oxygen, $\sim$ 15 ppm carbon and $\sim$10 ppm nitrogen. The samples were lightly electropolished, annealed for 8 h at 1470K in a vacuum of $\sim 10^{-6}$ torr, further electropolished to remove 0.05 to 0.1 mm from their diameters and tensile tested. Test pieces were pulled on a modified Hounsfield Tensometer at a strain rate of $\sim 5 \times 10^{-4}$ sec^{-1} in an argon atmosphere at temperatures between 657 and 706K. The gauge length and fracture surfaces of failed specimens were examined by both optical and scanning electron microscopy.

The ductility transition temperature of the material, T_T, defined as the highest temperature at which <0.1% plastic strain was observed, was 660 ± 3K. Thus, at 657K, macroscopic yielding was not detected (Fig. 1a) and brittle fracture was initiated at an inclusion (Fig. 2a). This fracture propagated predominantly by cleavage and occasionally by grain boundary parting. At 664K, only 7K above the temperature of this typical brittle failure, an entirely different behaviour was observed. The specimen exhibited a yield point drop of over 50% of the value of the lower yield stress, σ_Y, work

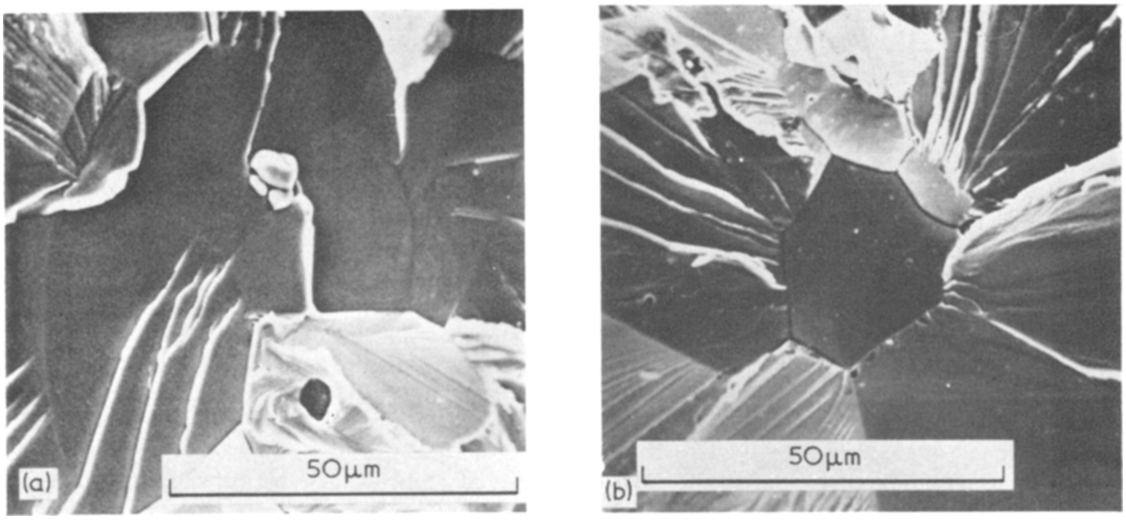

Figure 2 Cleavage fracture origins: (a) in a brittle specimen that failed at 657K and whose failure face consisted wholly of cleavage and intergranular facets, (b) in a specimen whose final failure, at 664K, was by micro-void coalescence. (This fracture initiating site formed part of a large area of cleavage fracture in the centre of the failure face.)

hardened some 3% and then the load-elongation curve dipped steeply down and failure took place after an elongation of ~25% (Fig. 1b). The slope of the final part of the load-elongation curve was always finite and negative, which indicates that a continuous failure mechanism was operating and that it was only complete at zero applied load.

The failure face of this specimen was made up of two distinct regions: a central area of transgranular cracking and grain boundary parting and an outer rim of dimple fibrous

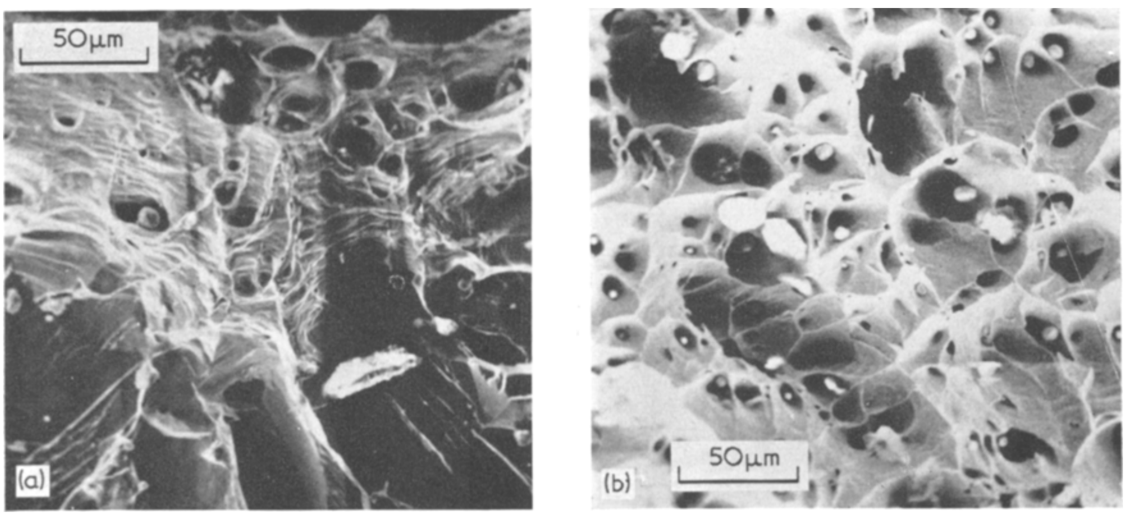

Figure 3 Dimple fibrous failure: (a) in a specimen tested at 664K that only had a rim of fibrous failure (part of the cleaved core of the specimen where failure originated, is seen at the bottom of the micrograph), (b) in a specimen broken at 706K that failed entirely by micro-void coalescence The particles in the dimples are believed to be oxides.

shear. Failure of the central region appears to have been initiated at two separate grain boundary facets (e.g. Fig. 2b); the general appearance of which closely resembled that of the failure faces of brittle specimens. Similarly the rupture in the outer region seems to have been a classical case of the nucleation and coalescence of voids formed on oxide particles (Fig. 3a). Furthermore, this region closely resembled the complete failure faces of those specimens which were pulled at and above 706K (Fig. 3b), work-hardened extensively before necking (e.g. Fig. 1d) and ruptured after large ($\sim$ 70%) reductions in area.

The rims of the failure faces of the test-piece broken at 657K were heavily deformed for most of their circumferences, but over one small arc of each there were only a few slip lines. This undeformed arc was where the cracking in the centre had propagated to the specimen surface. The comparative absence of slip on this arc strongly suggests that cleavage fracture preceded the earliest stages of fibrous failure, otherwise the whole rims of the failure faces would have been deformed. Specimens tested at 675 and 684K were very similar to the one tested at 664K; including at 675K, emergence of cleavage at the specimen surface over a few degrees of arc. A test-piece pulled at 688K showed a greater degree of work-hardening before necking and continuous failure (Fig. 1c). There was only a small cleavage patch on its failure face (Fig. 4a) but fracture in this central region could not be distinguished from the cleavage of completely brittle specimens. At temperatures above 706K test-pieces work-hardened extensively before necking (e.g. Fig. 1d) and failed after $\sim$70% elongation and large reductions in area (e.g. Fig. 4b); their failure faces consisted entirely of dimples associated with inclusions (e.g. Fig. 3b).

Non-propagating grain boundary micro-cracks were found on the gauge length of some of the test-pieces pulled at 664K or above. Examples are shown elsewhere [6].

To ascertain whether cracking preceded fibrous rupture in the failure process in this critical temperature range, a specimen was strained at 675K just past the point of plastic instability and unloaded. It was seen that, after only $\sim$3% elongation the test-piece was nearly broken through at the neck; in fact a "crack" through the neck went about 270° round the specimen. Most of the edge of this "crack" was very deformed, but over one small arc, about three grain diameters long, only a few slip lines could be seen on either side of it. The specimen was then broken at room temperature by cracking through the remaining area of the nearly separated neck. The revealed 675K failure face, sketched in Fig. 5, was, as expected, composed of a central region of transgranular and intergranular cracking and a boundary area of dimple, fibrous rupture. (The small

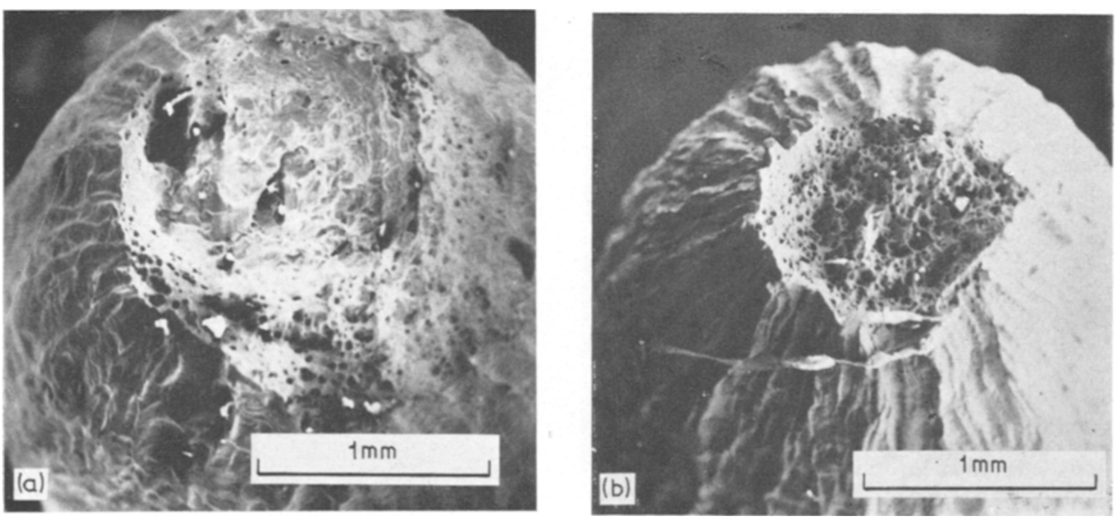

Figure 4 Entire failure faces: (a) of a test piece broken at 688K: in the centre by cleavage and at the edge by fibrous parting, (b) of a specimen that was pulled at 706K and broke entirely by fibrous parting.

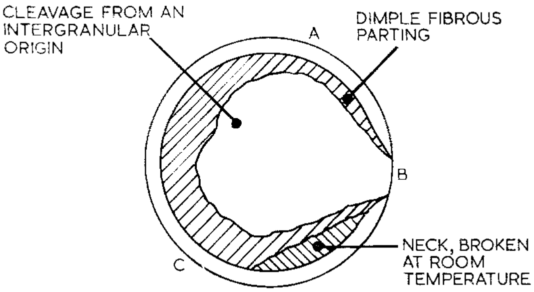

Figure 5 A sketch of the failure face of a specimen that was partly broken at 675K, unloaded, and the remaining neck, C, broken at room temperature. Examination before the fracture was completed showed that most of the gauge length near the failure zone (e.g. at A and C), was heavily deformed, but over one small arc, at B, there were only a very few slip lines on the gauge length close to the failure face.

undeformed part of the edge of the cracked neck proved to be where the central cleaved region reached the surface.) Since the rim fails after the centre, the experiment establishes that failure by cleavage was prevented by the inability of a large crack front to propagate, rather than through a difficulty of crack nucleation. Non-propagating surface grain boundary cracks were in fact seen in specimens deformed at and above 664K.

The amount of work-hardening preceding failure was negligible in some tests; it appears therefore that crack nucleation takes place first concurrently with or soon after yielding. The remaining question is why cracks can grow to lengths of as much as $\sim$1.45 mm, but not propagate to failure, as is typical of the other Group VIa bcc transition metals. (The observation of fibrous failure in chromium is in itself noteworthy, as even after extensive deformation and necking, frequently, fracture by sharp crack propagation has been observed [2].) A tentative interpretation of the large increase in the surface energy of fracture, which causes the crack front to stop is the pronounced effect of freshly produced dislocations. This hypothesis is consistent with the work on pressurized chromium, for which Mellor and Wronski [7] estimated that an increase in the mobile dislocation density of $< 10^7$ cm^{-2} more than doubled the surface energy of fracture and reduced the ductility transition temperature by 280K.

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Anomalous internal friction in linear polyethylene at low temperatures

In linear polyethylene (LPE) the dominant relaxation below room temperature is the γ_I relaxation at $-127°$C (0.67 Hz) [1-3]. This relaxation is attributed to the relaxation of the non-crystalline fraction since its magnitude increases with decreasing crystal volume fraction [3]. We have examined in torsion pendulum experiments the effect of quenching from room temperature to liquid nitrogen temperature. The