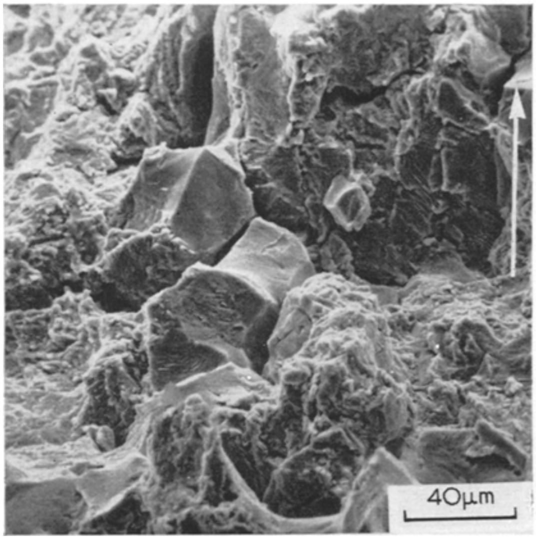

Figure 1 Structure sensitive stage IIa fatigue fracture in OFHC copper.

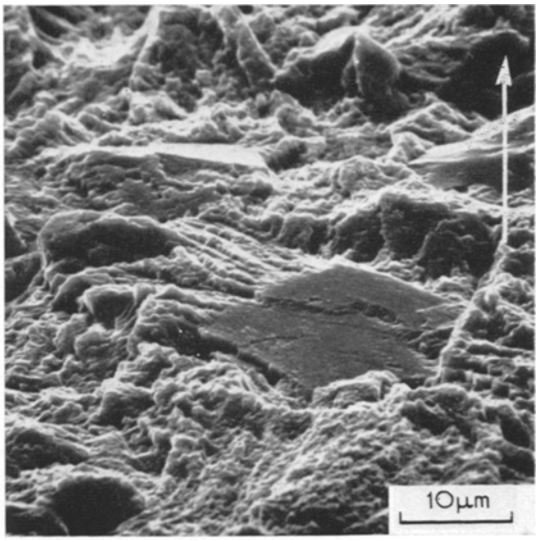

Figure 2 Structure sensitive stage IIa fatigue fracture in 18/8 stainless steel.

Structure sensitive fatigue fracture in FCC materials

The occurrence of structure sensitive fatigue fracture in low carbon steel has been described in a recent paper [1] where it was also stated that the effect had been observed on the stage II fracture surfaces of some fcc metals.

Tests were carried out on edge notched specimens of O.F.H.C. copper and commercial 18/8 stainless steel in push-pull using a 2 ton Amsler Vibrophore. Figs. 1 and 2 show the fracture surfaces of copper and stainless steel respectively, close to the notch root ($da/dN <$ 100Å); whilst Figs. 3 and 4 show areas of rapid fracture close to failure ($da/dN >$ 1000Å). (The arrow indicates the direction of crack propagation.)

It is evident that at slow crack growth rates, structure sensitivity occurs in both materials. The facets on the copper fracture (Fig. 1) are clearly grain boundaries whilst those on the stainless

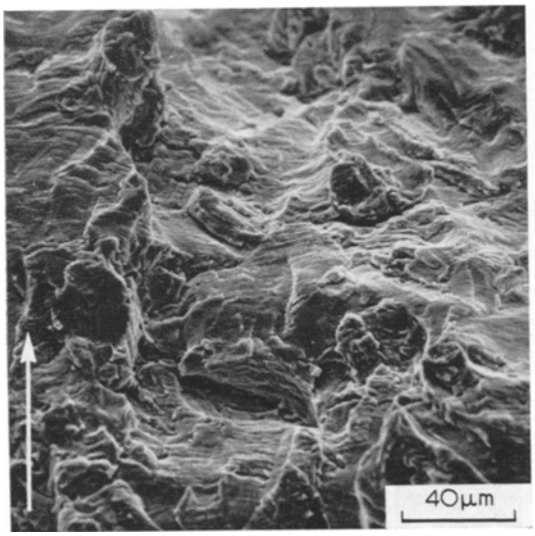

Figure 3 Stage IIb fatigue fracture in OFHC copper.

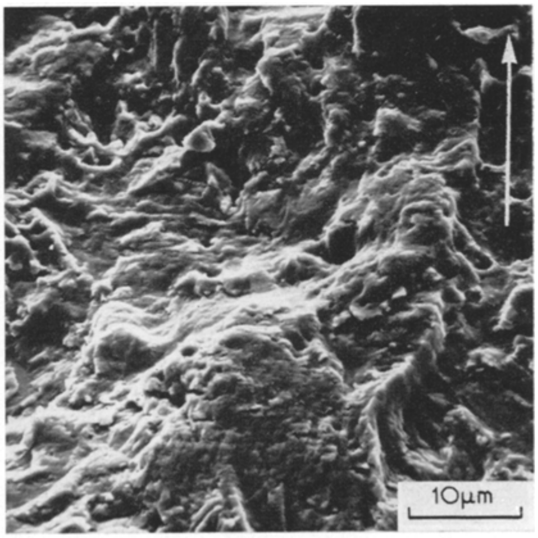

Figure 4 Stage IIb fatigue fracture in stainless steel.

steel (Fig. 2) may be associated with twin boundaries. The "hill and valley" features observed on the low carbon steel are less pronounced in these materials. At higher crack growth rates the more usual stage IIb mode, exhibiting striations and branch cracks, is operative in both cases (Figs. 3 and 4). Similar effects have recently been reported in titanium alloys [2] and may well occur in many other metals and alloys.

It is thought that the size of the plastic zone at the fatigue crack tip relative to the grain size of the materials limits the extent of stage IIa fracture [1] although the precise mode of IIa failure may depend on environmental effects (e.g. corrosion) in the particular metal or alloy.

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Received 21 March and accepted 26 March 1973

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Slow crack growth in proton- and deuteron-irradiated quartz

It is well known that cracks in silicates can grow subcritically in the presence of a chemically reactive external environment [1, 2]. In principle, there is no reason why analogous crack growth behaviour should not arise from migration of an active chemical species *within* the material, although bulk diffusion might be expected to slow down the process [3]. Thus, for instance, large cracks in soda-lime glass grow noticeably at constant load in vacuum, whereas in pure fused silica no such effect is observed [4]: the network-modifier component in the glass clearly plays an active role in the crack extension.

In this note, we present observations of slow crack growth in irradiated quartz as evidence for the above phenomenon. Single crystal slabs of (0001) α-quartz were pre-abraded to introduce a uniform layer of surface microcracks some 10 μm in depth. A selected area of surface was then exposed to a uniform beam of protons or deuterons (accelerated through 0.9 and 1.0 MV respectively) such that the mean ion penetration depth [5] extended just below the abrasion microcracks. The total irradiation dose was $\simeq$ 1-2 $\times 10^{22}$ ions m^{-2} in all cases. To avoid surface fragmentation arising from charging effects, beam currents were maintained well below 5×10^{18} ions m^{-2} sec^{-1}. Hertzian fracture tests were then carried out on both irradiated and unirradiated portions of the specimen surface in an evacuated chamber [6, 2].

In the Hertzian test, a spherical indenter is loaded onto the crystal surface at a constant rate until a "cone crack" propagates from a particularly favourable flaw just outside the contact

TABLE I Critical loads to cone fracture (N) at three temperatures and at "fast" (75 N sec^{-1}) and "slow" (2.5 N sec^{-1}) load rates

Temperature (°C)	Load rate	Unirradiated	Proton-irradiated	Deuteron-irradiated
25	fast	650 ± 38	712 ± 55	864 ± 82
225	fast	565 ± 32	663 ± 40	690 ± 59
	slow	558 ± 18	665 ± 55	714 ± 70
500	fast	287 ± 19	260 ± 29	271 ± 31
	slow	282 ± 23	186 ± 17	179 ± 21