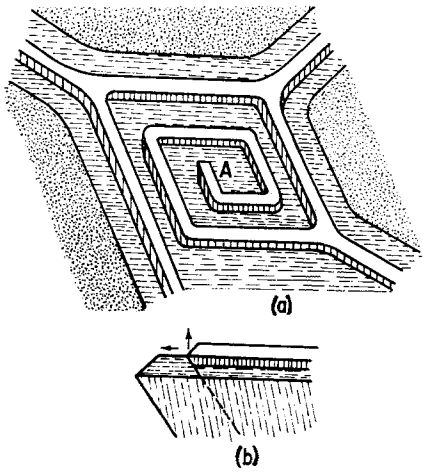

Figure 6 (a) Mg_2Si phase grows ahead of the eutectic interface enabling edgewise growth at A. (b) Leading edge of Mg_2Si grows at angle with vertical.

the eutectic leading to the first turn of the spiral. These features are as observed by metallography and by scanning electron microscopy of extracted crystals.

The Mg_2Si phase in the eutectic grows as (100) and (110) oriented plates. Starting at the Mg_2Si plate marked 1 in Fig. 3b, branching occurs at either edge, as the αAl phase grows over the Mg_2Si plate. The plates then branch again in turn, to give the first spiral loop with attached appendages shown in Fig. 3c.

Fig. 4 shows a longitudinal metallographic section of a spiral eutectic formation. This microstructure is arrived at by the stages of growth shown in Fig. 5 a, b and c. The cross-section of Fig. 5a corresponds to the first loop shown in Fig. 3c.

The mode of edge branching observed in the eutectic, which leads to the first spiral loop, is characteristic of the mode of growth of the primary crystal in hopper form. For the latter type of growth, crystal edges are unstable relative to crystal faces, and this is expressed in the branching behaviour of the eutectic Mg_2Si plate. Once the first loop has formed, growth proceeds as a spiral through the stages shown in Fig. 5, by a combination of lengthwise growth with a lateral component at A. The outer branches of the first loop shown in Fig. 3c have only limited growth and become redundant. Fresh turns add on at the centre by solute attachment to the leading edge A, providing the lateral growth component to wind the spiral. Lengthwise growth of the eutectic, is by the usual short range diffusion between phases.

The Mg_2Si plates could not spiral, if this phase did not grow ahead of the eutectic interface as is noted in eutectics of this irregular character [6]. The situation is shown in Fig. 6 the leading edge of the spiral is required to grow at an angle to the vertical.

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Comments on "Dependence of room temperature fracture strength on strain-rate in sapphire"

In a recent paper, Pollock and Hurley [1] have studied the strain-rate dependence of the fracture of sapphire filaments at room temperature. They show, very convincingly, that a strain-rate dependence of the strength is obtained, even when an environment is excluded from the region adjacent to the fracture initiating flaws.

They interpret the strain-rate dependence as a manifestation of dislocation-assisted slow crack growth. We believe that this interpretation is speculative and propose that the observed behaviour is due to another phenomenon,

probably thermally activated slow crack growth.

First, let us examine several of the basic predictions of dislocation-assisted slow crack growth models for single crystals: (1) the shear stress in the general vicinity of the fracture initiating flaw must exceed the dislocation flow stress to achieve stress enhancement (as correctly stated by Pollock and Hurley); (2) the stress enhancement produced by the dislocations at the tip of the fracture initiating flaw must be large enough to cause crack extension; (3) the strain-rate dependence of the strength should be related to the strain-rate dependence of the dislocation flow stress. Each of these requires detailed consideration before applying a dislocation-assisted slow crack growth model to strength observations. We find, based on these conditions, that the existing information on dislocation mobility in sapphire does not convincingly support the contention that dislocation-assisted slow crack growth occurs at room temperature.

(1) For *c*-axis filaments, gross dislocation plasticity has only been detected above ~1600° C [2] where the flow stress is in the range 200 to 600 MN m^{-2}, and the apparent activation energy for yield is 140 kcal mol^{-1}. Extrapolating these high temperature data to room temperature gives a dislocation flow stress (for the operative slip system) larger than the theoretical shear strength, τ_t, suggesting that no dislocation activity is expected prior to τ_t (i.e. below $\sim 2 \times 10^4$ MN m^{-2} [3]). Hardness indentations also give some indication of the dislocation flow stress, σ_y, for a general deformation (if the indentations form without associated cracking), and these suggest [4] that $\sigma_y \gtrsim 7 \times 10^3$ MN m^{-2}. Both measures of the dislocation flow stress are larger than the fracture strengths measured by Pollock and Hurley, which range from 2 to 3.5×10^3 MN m^{-2}. A dislocation-assisted slow crack growth mechanism can only apply, therefore, if the stress near the fracture initiating flaws is *substantially larger than the applied stress*.

Some stress enhancement occurs in the vicinity of the voids, (which form a core structure in these filaments [1]), and for a spherical void, $\sigma_{\theta\theta}$ for example, is [5];

$$\sigma_{\theta\theta} = \frac{\sigma_a}{2}\left[1 + \frac{d^2}{r^2} - \left(1 + \frac{3d^4}{r^4}\right)\cos 2\theta\right] \quad (1)$$

where σ_a is the applied stress and d, r and θ are shown in Fig. 1. The maximum stress is 3 σ_a, but this rapidly reduces to 1.2 σ_a at $r = 2d$. There is a possibility, therefore, that a flaw (as yet unidentified) within a distance less than d from a void *might* be exposed to dislocation activity. We suspect that this is very unlikely because no dislocation activity has been detected at room temperature at the tip of moving or arrested cracks in alumina [6], and the stress concentration at a crack tip is much larger than in the vicinity of voids. However, it might be argued that this lack of dislocation activity at cracks may be a source problem, (i.e. no active dislocation sources near the crack) and, hence, the possibility of dislocation motion near the voids cannot yet be *unequivocally* eliminated.

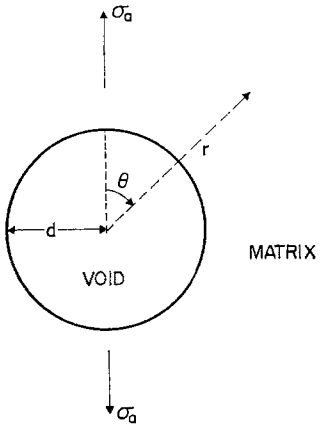

Figure 1 The co-ordinate system used to describe the stresses around a void.

(2) A mechanism for crack extension by dislocations, which permits the "focusing" of dislocations near the crack tip, has been proposed and described by Clarke *et al* [7]. This model requires that dislocations generated near the tip of the fracture initiating flaw should cross-slip to form a cross-slip source which then transmits dislocations back towards the crack tip. This process could operate in the sapphire filaments. It should be noted, however, that the dislocations moving towards the crack are necessarily repelled by the crack tip shear stress field. An applied shear stress significantly larger than the lattice friction stress is thus needed to move the dislocation close to the crack tip and, hence, to produced crack extension. This effect further reduces the distance from a void where dislocation-assisted slow crack growth can operate.

(3) For dislocation-assisted slow crack growth, the strain-rate dependence of the strength is

related to the strain-rate dependence of the dislocation flow stress, m [8].* Values for m have been obtained for c-axis deformation at 1600 to 1800°C by Gooch and Groves [2], and these range from 3.6 to 4.5. The exponent m decreases as the temperature increases, and for an activated process, we expect m to be *inversely proportional* to T; $m \propto 1/T$ (dislocation velocity data tend to confirm this approximate relation [9]). At room temperature, therefore, m is expected to be ~24 for the same slip system. Replotting Pollock and Hurley's strength data as log $\dot{\epsilon}$ versus log σ_f (where σ_f is the fracture stress and $\dot{\epsilon}$ the strain-rate) in Fig. 2, gives a slope of 28, which is not very different from the expected value of m. Some consistency is thus apparent.

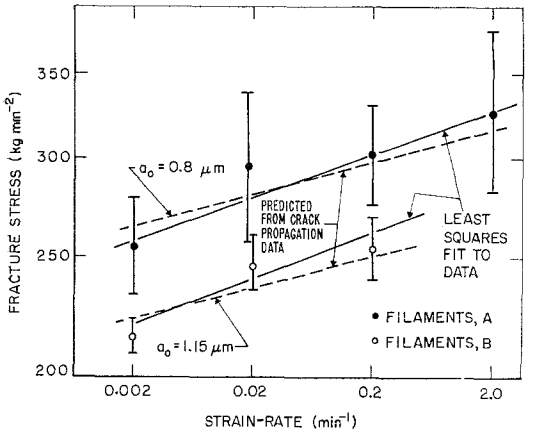

Figure 2 The strain-rate dependence of the strength of sapphire filaments of two lengths (A and B), compared to the strengths predicted from region III fracture mechanics data. The plot is logarithmic in stress and strain-rate; the solid lines are at least-squares fit to the data, and the limits are 95% confidence limits on the mean.

The validity of the model proposed by Pollock and Hurley thus rests on two presumptions; (a) fracture initiating flaws exist close to the voids (within approximately half the void radius), (b) dislocation activity occurs in this same region, close to the voids. The onus for establishing this model thus rests initially on the investigator's ability to confirm these two presumptions. Then some quantification (in terms of flaw size, flaw extension stress, etc.) is needed for final confirmation. This is a formidable task and, we believe, unwarranted;

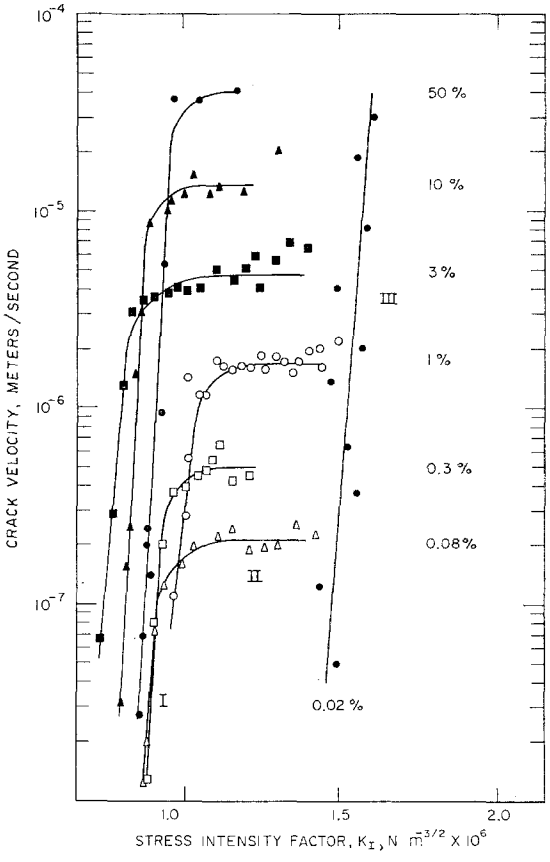

Figure 3 Stress intensity factor, K_I, versus crack velocity data obtained for sapphire (on the $(10\bar{1}2)$ plane) at room temperature for various humidities [11].

because the observations (Fig. 2) can be explained, quantitatively, in terms of the slow crack growth process observed in sapphire, *in vacuo*.

Fracture mechanics data obtained by Wiederhorn [11] and Wiederhorn, Hockey and Roberts [6] show that slow crack growth can occur in sapphire in vacuum or in dry nitrogen (similar behaviour is observed in several glasses [12]). At room temperature, this slow crack growth in inert environments (region III behaviour, Fig. 3) can be represented by the equation [13],

$$V = AK_I^n \qquad (2)$$

where V is the crack velocity, K_I is the stress intensity factor and A and n are system con-

*$m = \dfrac{\Delta \log \dot{\epsilon}}{\Delta \log \sigma}$, as determined by strain-rate change experiments [10].

stants. This behaviour leads to a strain-rate dependence of the fracture strength, σ_f, (as described by Evans), yielding the following relation for σ_f [13, 14];

$$\sigma_f{}^{n+2} = \frac{2}{A}\left(\frac{n+1}{n-2}\right)\frac{E\dot{\epsilon}}{Y^n}\,a_0{}^{1-n/2} \qquad (3)$$

where E is Young's modulus, a_0 is the pre-existing crack size and Y is a constant ($\approx \sqrt{\pi}$). This result predicts that the slope of a log $\dot{\epsilon}$ versus log σ_f plot should be $(n + 1)$. The fracture mechanics data (Fig. 3) give $n = 32$; predicting a slope of 33 for the strength plot. This compares well with the value of 28 obtained from the data in Fig. 2. Additionally, the strengths predicted from the fracture mechanics data (using Equation 3), can be compared directly with the strength data, for the requisite a_0. The comparison is shown in Fig. 2, with $a_0 = 0.8$ μm for the A filaments and 1.15 μm for the B filaments. The predicted strengths fall within the 95% confidence limits (on the *mean*) for the data. These pre-existing flaw sizes appear reasonable or the sapphire filaments, suggesting that the voids themselves might be the fracture initiating flaws – as noted in many other systems [4].) It has clearly been established that the vacuum slow crack in sapphire and glass is *not* due to dislocation (or viscous) activity [6, 12]. Thermally activated bond rupture at the crack tip is the apparent slow crack growth mechanism [12]. We propose that the strength data obtained by Pollock and Hurley is entirely consistent with this mode of slow crack growth.

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Reply to "Comments on 'Dependence of room temperature fracture strength on strain-rate in sapphire'"

In a recent paper [1], the authors presented data showing the strain-rate dependence of the fracture stress in Saphikon* filament. This strain-rate sensitivity did not result from moisture assisted crack growth and was interpreted as being consistent with a particular mechanism of dislocation aided crack growth. Evans *et al.* [2] have indited these arguments as being speculative, and have instead suggested that the data may be explained by thermally activated slow crack growth.

Partly, Evans *et al.* have criticized our hypothesis on the basis of a detailed calculation of the magnitude of the stress enhancement around a spherical void and its fall off with distance from the centre of the void. This stress enhancement becomes small within approximately one void radius of the edge of the void. Thus if flow takes place it must occur from a source close to the stress concentrator. Further, since flow is assumed to occur over limited small distances, the original void must be close to the subcritical crack which is to grow. In Saphikon, these necessary circumstances are believed to be supported by the following: voids are generally observed to lie in closely spaced linear or random arrays (Fig. 1 and see also several figures in [3]); and, dislocations revealed by

*Saphikon filament is sapphire filament grown by the EFG method.

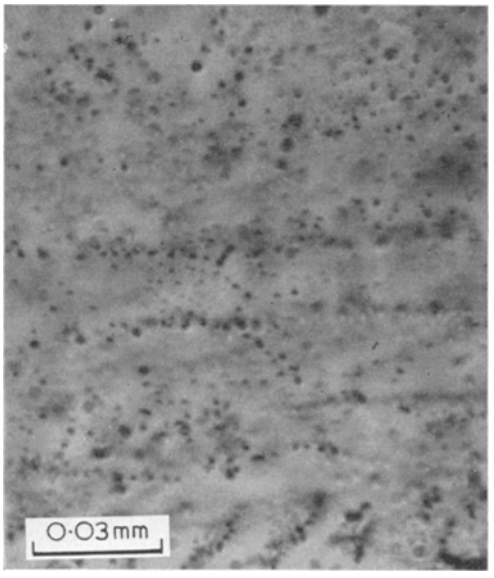

Figure 1 Transmitted light photomicrograph of void distribution near surface of a Saphikon filament.

A second point which must be made in this connection is that unlike the voids formed in Czochralski grown sapphire which are normally spherical, the voids in Saphikon filament are often facetted and, as noted above and in [1], are closely spaced within the volume of filament where fracture is believed to originate. Fig. 2 shows an example of facetted voids intersecting a fracture surface in a sapphire filament [5]. Thus Equation 1 of [2] probably provides only a lower boundary for the maximum stresses which could occur in the filaments under test.

Evans *et al.* [2] also present an alternative explanation of the authors' strain-rate data. They show that the data are consistent with a to-be-published thermally activated slow crack growth model [6]. Nevertheless, it is clear that the major point to be resolved is the occurrence or possibility of dislocation motion in sapphire. Dislocation activity has not to date been noted below 400°C in Czochralski sapphire [7]. However, as indicated above there is substantial evidence for the abundant presence of both stress concentrations and crack precursors in Saphikon. This prevents us from discounting at this time the possibility of dislocation assisted crack growth leading to fracture.

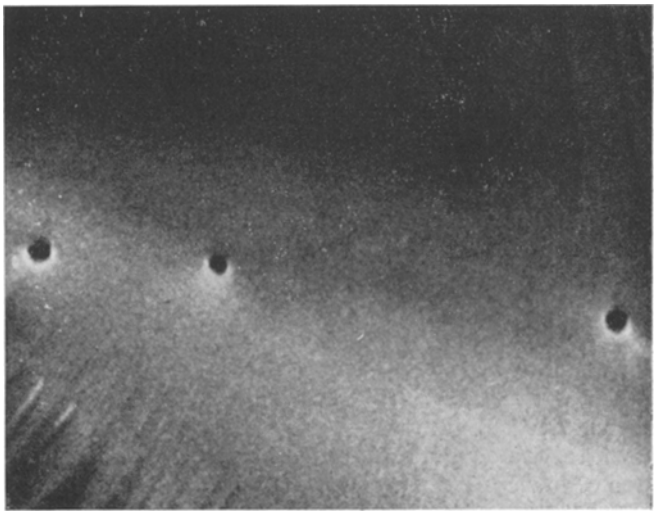

Figure 2 A portion of the fracture surface of a Saphikon filament showing the intersection with three voids (× 4000).

etching the basal plane in larger but identically oriented sapphire shapes grown by EFG are always observed to be most densely packed into the most heavily voided regions. Early observations of etched filament cross-sections have shown the same features [4]. Thus, voids do lie in arrays spaced within a radius, and dislocations are found to be densely arrayed in the same volume. Therefore, it also follows that it is a reasonable if not proven assumption that stress concentrations, dislocations or sources, and crack precursors (voids) lie within the required proximity.

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The effect of a glass substrate on cooling rate in splat-quenching

The technique of splat-quenching to produce thin films of rapidly solidified material is well-established. In the Duwez or "gun" technique [1] a small quantity of the molten alloy is ejected in a fine spray against a high-conductivity substrate. The measurement of cooling rate in such experiments is a much debated topic [2, 3]. Both direct measurements using thermocouples [4] and radiation pyrometers [5] and indirect estimates from the scale of microstructure of the solidified film [6, 7] have been employed. Although at best such methods can be described only as estimates they are useful in comparative studies where an absolute value of the cooling rate is not required. With this in mind the eutectic lamellar spacing method [7] was used by the author in a recent study [8] of the effect of various experimental parameters on the cooling rate attainable in a newly constructed controlled atmosphere gun-type splat-quencher. To obtain high cooling rates it is usual to use a substrate of high thermal conductivity. This note discusses the unexpected results obtained when the high conductivity copper substrate was replaced by one of low conductivity material (soda glass).

Specimens of eutectic composition (17.3 at. % Cu) aluminium copper alloy were splat-quenched from 800°C onto a copper substrate roughened with grade 400 emery paper and onto a sheet of 6 mm thick window glass. Care was taken to ensure that in all cases the other experimental conditions such as specimen mass, atmosphere (600 Torr argon), diaphragm rupture pressure etc, were constant. The resulting foils were sectioned and mounted transversely in araldite prior to polishing to a 1/4 μm surface finish and etching in 10% NaOH. Plastic/carbon replicas were then made for electron microscopic examination. These replicas allowed not only the microstructure to be examined but also the local foil thickness to be measured.

Three types of microstructure were observed; lamellar eutectic, degenerate eutectic and radial lamellar arrangements apparently nucleated at primary particles. These morphologies were identical to those observed in a previous study of the same system by Burden and Jones [7]. These authors showed that the relationship:

$$\lambda = AR^{-n}$$

where λ is the lamellar spacing and R the growth rate of a lamellar eutectic, may be extrapolated to the growth rates expected in splat-quenching. For the Al–$CuAl_2$ eutectic values of $n = 0.5$ and $A = 1.04 \times 10^{-8}$ m$^{3/2}$ sec$^{-1/2}$ were obtained.

Accordingly the finest spacings in each lamellar region were measured, along with the local foil thickness, and hence the appropriate growth rate R (m sec^{-1}) was evaluated. As a consequence of sectioning each lamellar region had a wide range of spacing; in every case the finest observed spacing was taken as the true spacing. (Even if the lamellae with the smallest apparent spacing were inclined at 20° to the normal to the foil section the error in estimating the true spacing by this assumption would be only 6%.)

From a knowledge of the growth rate and foil thickness the cooling rate was calculated as follows. The splat-substrate heat transfer coefficient, h (J m^{-1} K^{-1} sec^{-1}), and growth rate, R (m sec^{-1}), are related by the expression [4]:

$$h = \frac{\rho LR}{\theta_F - \theta_A} = 2.5 \times 10^6 R \qquad (1)$$

where ρ is the density of the splat material, taken as 3.84×10^6 g m^{-3} (assuming the values for Al and $CuAl_2$ [9] and the proportions in the eutectic as in the phase diagram). L is the latent heat of solidification, taken as 340 J g^{-1} on the same basis as ρ, θ_F and θ_A are respectively the freezing and ambient (substrate) temperatures, taken as