We wish to thank the Directors of the Plessey Company Limited for permission to publish this paper.

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Comments on "On the influence of deformation rate on intergranular crack propagation in Type 304 stainless steel"

I should like to dissent with some of the discussion of the paper by Nahm *et al* [1]. These authors present data relating to the growth characteristics of wedge cracks and infer that the growth rate depends on wedge height (rather than crack length) and that the angular orientation of the fastest growing cracks is also independent of crack length. This information is most useful but, unfortunately, in discussing it, the authors have tended to use an analysis, formulated by Evans [2], for the growth of grain-edge cavities. Such an approach is incorrect and leads to erroneous conclusions regarding the applicability of the cavity growth model.

My objection may best be summarized by reference to Figs 1 and 2. Fig. 1 represents schematically the geometry of a wedge crack growing along grain boundary A and fed by sliding on boundaries B and C to form a wedge of height (na). This is the type of crack found by Nahm *et al* for which an analysis by Williams [3] is applicable. He predicts that

$$\frac{\delta l}{\delta(na)} = \frac{\mu}{\pi(1-\nu)\sigma}\left\{\left[1 - \frac{\sigma(na)}{2\gamma_p}\right]^{-\frac{1}{2}} - 1\right\} \quad (1)$$

where μ is Poisson's ratio, σ is the applied normal stress and γ_p is the effective fracture energy. Clearly, from this equation, the growth process depends directly on the wedge height rather than on the crack length and this is entirely in accord with the observations of Nahm *et al.*

On the other hand, Evans' analysis relates only to the situation depicted in Fig. 2. Here a cavity lies on a single grain boundary D experiencing a sliding displacement, a, produced by the shear stress τ. The cavity growth relationship is [2]

$$\frac{\delta l}{\delta a} = \frac{\tau}{2\gamma NW}\left\{1 - \frac{\pi(1-\nu)\sigma_n^2 l}{8\mu\gamma_p} - \frac{\sigma_n h}{2\gamma_p}\right\}^{-1} \quad (2)$$

where N is the number of cavities per unit area of boundary, W is the average cavity width, h

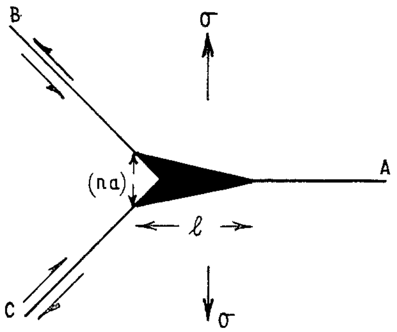

Figure 1 Wedge crack geometry.

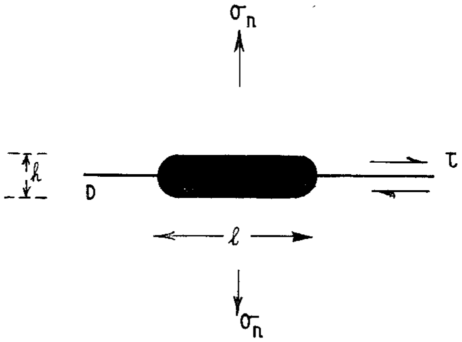

Figure 2 Cavity geometry.

is the cavity height which is taken as constant for small cavities and equals the height of the discontinuity at which the cavity nucleated and σ_n is the applied stress component normal to the plane of the boundary. Equation 2 shows that for small, thin cavities, their growth rate is not significantly dependent on their length (stage I behaviour) whereas the growth rate of large cavities increases with increasing cavity length (stage II behaviour). Because of the presence of both a shear stress and normal stress term in Equation 2, it transpires that small, stage I cavities will behave as shear cracks and will grow fastest when situated on boundaries experiencing the maximum sliding rate – probably those situated at 45° to the stress axis; large, stage II cavities, however, will behave as tensile cracks and grow fastest when situated on boundaries making large angles to the stress axis. This transition in behaviour has experimental verification [4] but is predicted for cavities only and not for the wedge type cracks studied by Nahm *et al.*

This difference highlights the importance of applying the theoretical models only to the appropriate physical situation. The detailed conclusions obtained from cavitation models do not necessarily apply to wedge cracking models even though the basic mechanism of nucleation and propagation are the same.

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Reply to "Comments on 'On the influence of deformation rate on intergranular crack propagation in Type 304 stainless steel'"

Evans [1] has dissented with a portion of the discussion in a recent paper by the present authors [2] related to the growth behaviour of intergranular wedge cracks. The position taken by Evans with respect to the clear distinction between the growth of wedge cracks [3] and grain edge cavities [5] is appreciated. However, it is unfortunate that Evans has determined that the discussion by Nahm *et al* [2] was based on the analysis for grain edge cavities rather than on the analysis for wedge cracks.

Contrary to Evans' contention that the incorrect analysis was used, the discussion by Nahm *et al* [2] clearly included the analysis of wedge crack growth by Williams [3]. However, the results of Nahm *et al* are in general agreement with this analysis only in so far as the rate of crack growth with respect to crack width (or wedge height), $\mathrm{d}c/\mathrm{d}(na)$, is directly dependent on crack width, na, rather than on crack length, c. The equation given by Williams [3], based on Cottrell's model [4], to describe this relationship is

$$\frac{\mathrm{d}c}{\mathrm{d}(na)} = \frac{\mu}{\pi(1-\nu)\sigma}\left[\left(1-\frac{\sigma(na)}{2\gamma}\right)^{-1/2}-1\right] \quad (1)$$

where σ is the applied tensile stress, γ is the effective fracture surface energy, ν is Poisson's ratio, and μ is the shear modulus. This relationship was not discussed by Nahm *et al* because it was considered that the limited agreement with the results was self-evident.

The discussion by Nahm *et al* was primarily concerned with the crack growth behaviour as a function of deformation rate. The simplified equation given by Williams [3] to describe the stable growth of wedge cracks in terms of the change in crack length, c, with respect to time, t, is

$$\frac{\mathrm{d}c}{\mathrm{d}t} \simeq \frac{\mu\sigma D^2 \dot{\epsilon}_{\mathrm{mcr}}^2\, t}{4\pi(1-\nu)\gamma} \quad (2)$$

where D is the grain size and $\dot{\epsilon}_{\mathrm{mcr}}$ is the minimum creep rate. This relationship indicates that the crack growth in terms of length will depend directly on the applied stress and $\dot{\epsilon}_{\mathrm{mcr}}$, and that the crack length increases with increasing crack width.

Equation 2 was derived from Equation 1 on the assumption that the deformation is predominantly restricted to grain-boundary sliding

[3]. However, in the paper under discussion [2], the contribution of grain-boundary sliding to the total deformation should vary significantly over the stress levels, as reflected in the deformation rates, considered. The information on the ratio of the grain-boundary sliding component of deformation to the total deformation is presently being generated at this laboratory. The applied tensile stress used in Equations 1 and 2 is the stress normal to the cracked grain boundary, σ_n, rather than the tensile stress applied to a polycrystalline specimen, σ_a. In the case of steady-state creep for which Equations 1 and 2 are primarily concerned, one may reasonably assume that $\sigma_n \simeq \sigma_a$. The data of Nahm *et al* however, are for deformation rates which cover nearly six orders of magnitude, and include both tensile and creep results, so that this assumption is not necessarily justified for all deformation rates. This circumstance, then, requires that the angular orientation of the cracked grain boundaries with respect to the direction of the stress applied to the entire polycrystalline specimen either be known or assumed to confidently apply Equation 2 to all of the results. For crack angular orientations of less than about 80 to 90° to the applied stress direction, as observed in the tensile specimens, a significant shear stress component must be considered in addition to the normal stress component to explain the crack growth behaviour. It is also recognized that the cracks which develop during the deformation of polycrystalline materials at elevated temperatures are actually three-dimensional, rather than simply two-dimensional. It was for these reasons, in addition to the results obtained by Nahm *et al*, that the suggestion was made that the change in crack *volume*, i.e. in the amount of damage produced during the deformation process, and not just the change in crack *length*, as a function of time, may best represent the crack growth behaviour of the 304 stainless steel investigated.

In summary, we agree with Evans' [1] contention that the correct theoretical model must be applied only to the appropriate physical circumstances. We believe, however, that the analysis by Williams [3] is not appropriate to adequately describe all of the results given in the paper in question. Clearly, the analysis by Evans [5] is not appropriate for these results either. The data do indicate that a new approach, possibly based on features from either both or other types of models, be considered to explain all of the results by Nahm *et al* [2] consistent with the physical circumstances. This possibility is being explored by the authors.

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Transverse compressive properties of an unbonded model cermet

In a previous paper [1], the transverse tensile properties of an unbonded model composite were found, in the belief that the behaviour of the model would indicate the type of variations in behaviour that might be expected in real materials. However, such materials might be used in compression: for instance, nuclear fuel cermets could be stacked vertically in a thermonuclear pile, and a possible use for soft metal/carbon fibre composites is bearing material. Therefore, a comparable piece of work has been done to investigate the properties of a similar model in compression.

Fig. 1 shows the dimensions and patterns of the specimens made from aluminium $2\frac{1}{4}$ wt % magnesium alloy plate 12.7 mm thick. Volume fraction was varied by increasing the hole size, and silver steel plugs were force-fitted into the holes. All specimens were annealed at 300°C.