PHILOSOPHICAL MAGAZINE A, 1983, VOL. 48, NO. 6, L59–L61

Reply to comments on "A model for the effect of grain size on the yield stress of metals"

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[*Received 10 June 1983 and accepted 23 September 1983*]

ABSTRACT

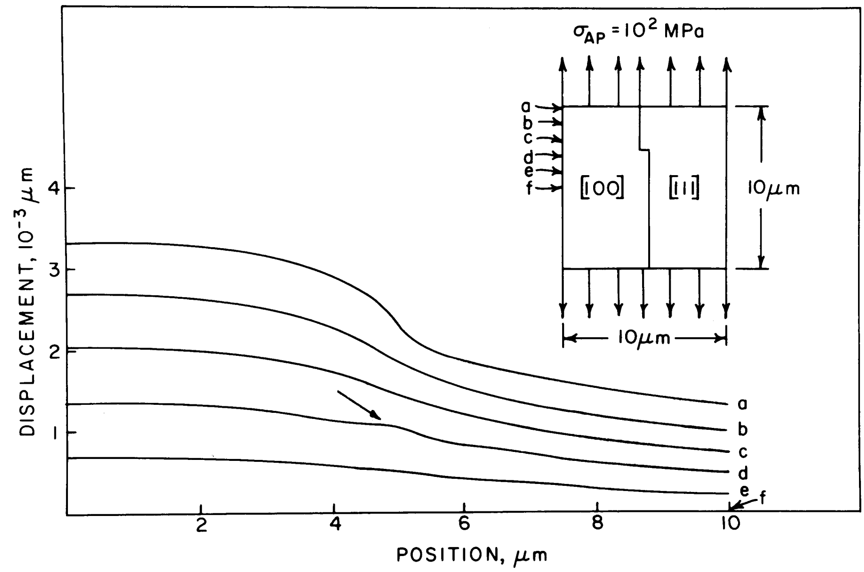
A rebuttal to a discussion (Kurzydlowski and Varin 1983) on a paper by Meyers and Ashporth (1982) is presented.

The interest of Kurzydlowski and Varin (1983) in our work is greatly appreciated. However, there are a number of imprecisions in their discussion that need to be addressed.

First and foremost, they base their discussion on three articles published in two journals only recently introduced into the market (1980 and 1981) which we did not have access to (nor did we have knowledge of their existence) when our work was being conducted (1979–1981). They also base part of their conclusions on a fourth article submitted for publication (Kurzydlowski, Varin and Zielinski 1983) and therefore not available. We consider the similarity between the results obtained on the grain boundary stresses as very positive confirmation of the correctness of our approach.

Their second criticism concerns the fact that we did not consider grain-boundary sliding or ledges in our model. In our initial computations we had incorporated grain-boundary steps and they have, indeed, a marked effect on the stress levels. The figure shows the displacement fields in two grains in the presence of a grain-boundary step; the procedure used is described by Meyers and Ashworth (1982). As one would expect, the step has an enhancing effect on the stress concentration. The displacement field *d* shows a fluctuation in the interface region, marked with an arrow in the figure. The section (*d*) is the one which passes closest to the grain-boundary step shown in the right-hand side of the figure. By decreasing the mesh size in the area surrounding the step one could arrive at a more accurate calculation of the effect of grain-boundary steps. We purposely kept this aspect out of our model and did not make any assumptions regarding the grain-boundary structure or grain-boundary dislocation sources. Hence, our model does not assume nucleation of perfect dislocations at grain boundaries, as implied by Kurzydlowski and Varin (1983). We did not propose nor invoke any specific mechanisms for dislocation generation since we believed that this transcended the scope of the work. Although grain-boundary steps provide the additional stress concentration, there are additional possible mechanisms:

(*a*) gliding of grain-boundary dislocations and reaction, producing matrix dislocations;



Displacement fields for interface with slip; finite-element method using the procedure described by Meyers and Ashworth (1982).

(*b*) decrease of the elastic stiffness constants in the grain-boundary region due to the local destruction of lattice periodicity and relaxation of atomic positions;
(*c*) matrix generation of dislocations in regions adjacent to grain boundaries;
(*d*) emission of matrix dislocations from grain-boundary ledges; specific mechanisms were proposed by Li (1963 a, b) and Price and Hirth (1982).

Thirdly, the unpublished work by Kurzydlowski *et al.* (1983) is invoked; apparently they found that, for stainless steel, dislocations in the grains were activated before any other sources at the grain boundaries operate. This is in conflict with experimental results obtained by Murr and Wang (1982), who found that the onset of plastic flow in AISI 304 stainless steel was characterized by the emission of dislocations from grain-boundary ledges which form emission profiles resembling dislocation pile-ups; in nickel, on the other hand, the higher stacking–fault energy favoured cross-slip, with a resulting dislocation distribution that was more random. Murr (1981) also conducted *in situ* straining experiments similar to the ones described by Kurzydlowski *et al.* (1983), using a high-voltage transmission electron microscope. He concluded from these observations on AISI 304 stainless steel that the grain boundaries were the principal sources of dislocations at plastic strains below 0·02.

The fourth criticism deals with the analogy used in our assumption that the stress (and not the strain) is the same in all grains. We used the rubber and steel sphere analogy only to illustrate the concept, and make no claim that the polycrystal acts exactly as a steel and rubber aggregate. The assumption of constant stress is a very important one; the entire theoretical development of stresses and strains in a real polycrystalline aggregate is obviously more complex but we truly believe that the isostrain hypothesis more closely describes the mechanical response at the onset of plastic deformation than the isostress hypothesis used by Taylor (1938), and Bishop and Hill (1951).

The fifth criticism addresses our proposal of cross-slip, once the dislocations are emitted from the grain boundaries. Again, results from unpublished *in situ* straining TEM experiments are used to back the criticism, without mention of the limitations of the experimental technique (tridimensionality of stress configuration is lost). We obviously recognize the effect of stacking-fault energy on the tendencey for cross-slip. We feel that these complicating factors would clutter the model excessively.

Lastly, we would like to emphasize that none of the criticisms damage the validity or limit the generality of the model. We recognize that we ignored a number of aspects. However, models are only attempts at understanding phenomena. They should incorporate their essential features and be able to predict the most important effects while retaining, if possible, the simplicity essential in making them useful.

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