fith–Orowan–Irwin equation,

$$\sigma = Y\sqrt{(E\gamma c)}/\sqrt{a}, \qquad (2)$$

where Y is a geometrical constant (assumed to be 1) [15], E is the elastic modulus ($E = 9.83 \times 10^6$ psi $= 6.77 \times 10^4$ MN m^{-2}) [2], γc is the fracture energy ($\gamma c = 6.3$ J m^2 $= 63 \times 10^{-6}$ MN m^{-2}) which is the mean value for Tennessee (quartz) sandstone determined by two teams [16, 17], and $a = 0.1$ mm is the flaw size. The value obtained for σ is 65.3 MN m^{-2}, and values of K (given in Table I) fall between 4.3 and 7.95 MN m$^{-3/2}$ for the various specimens studied.

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Comment on "Deformation in spinel"

As dislocation dissociation in metals is expected to influence slip in ionic and covalent crystals [1] at least at low temperatures, experimental attempts to derive the stacking fault energy (SFE) for these compounds are of fundamental interest. In the spinel $MgO(Al_2O_3)_n$, most of the work on SFE determination has been performed by Mitchell *et al.* [2]. The aim of this letter is to point out that their results are subject to some controversy, because all parameters influencing the final dislocation configurations have not been taken into account, resulting in possible erroneous conclusions.

In $MgO(Al_2O_3)_n$, perfect dislocations generally have $a/2\langle 1\,1\,0\rangle$ Burgers vectors. We have reported previously [3] that as a consequence of the dislocation dissocation into two partials with collinear Burgers vectors $a/4\langle 1\,1\,0\rangle$, dislocation network formation should involve climb unless the network geometry obeys precise criteria which, according to the Burgers vector $\boldsymbol{b}$ of the junction segment, are:

(1) $\boldsymbol{b} = a/2\langle 1\,1\,2\rangle$, the network is parallel to $\{1\,1\,1\}$ but the angles between its edges are 120°;

(2) $\boldsymbol{b} = a/2\langle 1\,1\,0\rangle$, the network is again parallel to $\{1\,1\,1\}$ but the angles may be 60° or 120°;

(3) $\boldsymbol{b} = a\langle 1\,0\,0\rangle$, the network is parallel to $\{0\,0\,1\}$, the angles are 90° or 45°.

In their analysis on plastic deformation in spinels, Mitchell *et al.* [2] stated that because glide has been observed to be activated on $\{1\,1\,1\}$

and especially on {1 1 0} for $n = 1$ and $n > 1$, respectively, dislocations must be dissociated in these planes in the related crystals, even if climb had to be taken into account in order to explain why dislocations were rarely observed in well defined slip planes ($n = 1$). Mitchell *et al.*'s data were calculated on the basis of TEM observations [4–6] carried out on specimens deformed above $0.5T_M$, a temperature range in which climb is generally expected.

In non-stoichiometric spinels [4, 5], dislocations which have dissociated into two partials with collinear Burgers vectors are observed in networks verifying none of the above conditions. Furthermore, no special information concerning dissociation width, actual dissociation plane or conditions of diffraction is given with the related micrographs. The accuracy of Mitchell *et al.*'s calculation of SFE is, therefore, questionable, as a plot of SFE versus n could be completely different for $n > 1$.

In stoichiometric spinel, the SFE was taken directly from a paper by workers of the same group [6] in which dislocation climb was not taken into account, for which evidence can be obtained from a simple analysis of the network. For example, it is unlikely that the "vertical $[0\,\bar{1}\,1]$ dislocation" does not exhibit dissociation in the plane of the foil because of its "approximate screw orientation"; in fact, the angle between the direction of the dislocation line and this screw orientation actually varies within 10° to 15°, which is sufficient to prevent the related dislocation from being dissociated in any plane other than the (1 1 1) foil plane, except if climb were involved or if the dislocation curvature were accommodated by convenient jogs. These jogs would then also be present on the other segments and, therefore, affect their apparent width. This aspect is also substantiated in Figs. 4 and 5 [6], where those dislocations which cannot be dissociated in the (1 1 1) foil plane do not exhibit a large change in their apparent dissociation width within an important range of curvature. Furthermore, in Fig. 4 [6] the lower dislocation with $a/2$ [0 1 1] Burgers vector has a $[1\,0\,\bar{1}]$ direction in the area of its maximum apparent width (~100 Å) and should, therefore, on the basis of glide dissociation only, be dissociated in the $(1\,\bar{1}\,1)$ plane and not in the foil plane as stated in [2], leading to an actual dissociation width of approximately 300 Å.

Finally, the exact meaning of the SFE versus n plots [2] is not clearly understood, because different fault configurations are mixed, i.e. $a/4\langle 1\,1\,0\rangle\{1\,1\,1\}$ at $n = 1$ and $a/4\langle 1\,1\,0\rangle\{1\,1\,0\}$ for $n > 1$.

We conclude, therefore, the SFE data [2], certainly of the right magnitude, are possibly correct, but will remain controversial unless they can be supported by new observations in which the actual dissociation plane is determined for spinels deformed at temperatures lower than $0.5T_M$ (for example from indentation tests) in which dislocations may show a different behaviour.

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Deformation diagrams of chip forming mechanisms

A novel way of presenting metal cutting data in the form of a deformation diagram is suggested which illustrates the way in which changes in both the properties of the workpiece and the cutting conditions can influence the mechanism of chip formation.

During machining, material is removed from the surface of the workpiece by the passage of a hard, sharp, wedge-shaped tool. Generally speaking,