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 $[10\bar{1}]$ direction in the area of its maximum apparent width (~100 Å) and should, therefore, on the basis of glide dissocation 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 approxmately 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. $a/4$ ⟨1 1 0⟩ {1 1 1} at $n = 1$ and $a/4$ ⟨1 1 0⟩ {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, in the analysis of this process it is assumed that the workpiece material shears plastically in some narrow and relatively well-defined zone extending from the cutting edge of the tool to the free surface. This enables the concepts of mathematical plasticity to be applied and, in many situations, there is reasonable correlation between theory and observation. One of the most important variables in such analyses is the severity of the friction force between the material being removed, in the form of a chip, and the rake face of the tool.

However, plastic shearing is not the only possible mechanism which enables the workpiece to accommodate the movement of the tool; brittle materials, e.g. ceramics, some plastics and even metals under some conditions, lose material by a process which involves fracture and the propagation of a crack extending from the edge of the cutting tool into bulk of the material. In metals the initiation of such a mechanism leads to the production of discontinuous chips and it is known that these can be encouraged by both metallurgical changes (e.g. the addition of free lead which affects material fracture toughness [1]), or changes in the cutting parameters, e.g. increases in the depth of cut or chip-tool friction.

In addition, another material response is also possible involving the thermal as well as the mechanical properties of the workpiece. When a metal is slowly deformed plastically the process is essentially isothermal. Initially, plastic strain is restricted to a few inherently weaker zones. Strain-hardening increases the resistance of these to further deformation, so that the material adjacent to them, as yet undeformed plastically and so weaker, now accommodates the applied strain until eventually the whole of the specimen is effectively work-hardened. As the rate of deformation increases, so the idealization of isothermal conditions becomes less tenable. The greater part of the energy of plastic deformation is dissipated as heat within the deforming material, so raising its temperature and reducing its resistance to further flow. Strain-hardening and thermal softening, therefore, oppose one another within the deformation zones. If the rate of decrease in strength, resulting from the local increases in temperature, equals or exceeds the rate of increase in strength, due to the effects of strain-hardening, the material will continue to deform locally. Deformation will concentrate in a narrow band and this unstable process is known as "adiabatic shear" [2]. During the cutting process the strains and strain-rates can be very large (greater than

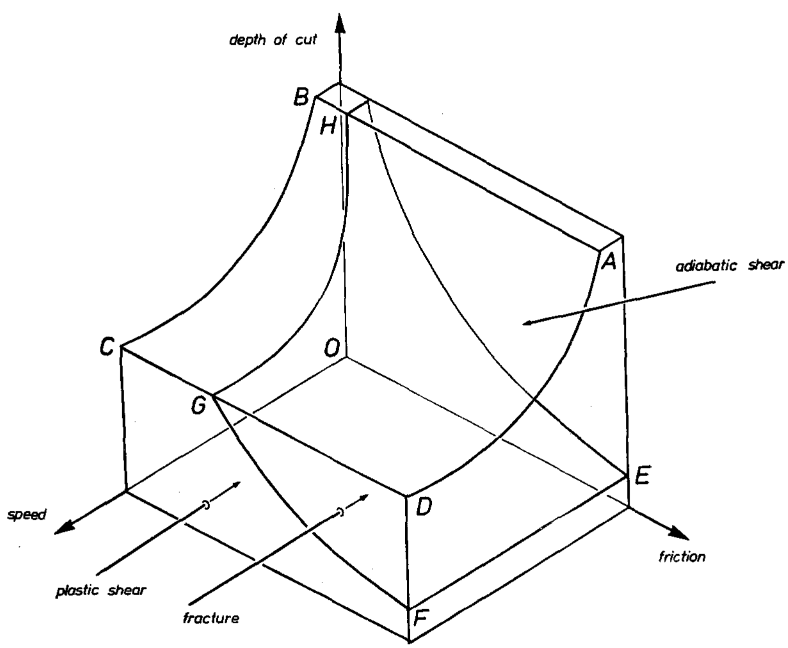

Figure 1 Diagrammatic representation of three possible modes of material removal during machining; namely, plastic shearing, fracture and adiabatic shearing.

unity and 10^5 sec^{-1}, respectively) so that the specific energy consumption is also high ($\sim 10^2$ MJ m^{-3}) and it is possible, under favourable circumstances, for material removal to involve this phenomenon of thermal instabilities.

There is now considerable experimental evidence that even within the deformation zone designated above as plastic shearing, there is a regular pattern of micro-instabilities leading to a fine lamellar chip form [3, 4]. A distinction is here drawn between this form of structural instability, which occurs even at very low cutting speeds, and that which may occur at much higher deformation rates when temperature rises can be considerable: it is this latter form of instability that has been denoted by adiabatic shear.

We can thus imagine three possible mechanisms leading to loss of material during machining – plastic shearing, fracture and adiabatic shearing – further, each will be described by an appropriate constitutive equation.

For a given material of known properties in a particular cutting geometry we can conveniently display these relations graphically on three orthogonal axes representing depth of cut, chip-tool friction and cutting speed. The boundaries of the fields appropriate to each deformation mechanism are obtained by equating the constitutive equations, setting values of two of the variables, say speed and friction, and solving for the third. Such a diagram is illustrated in Fig. 1, and can be thought of as being analogous to the "deformation mechanism maps" due to Ashby [5]. The surface ABCD represents the onset of the conditions necessary to promote adiabatic shear. Any particular experimental combination of depth of cut, cutting speed and chip-tool friction can be represented by a point in this three-dimensional space. If this point lies on the concave side of ABCD then adiabatic shear will be promoted preferentially. Material removal by a fracture mechanism is also shown on the diagram. Surface EFGH represents the boundary between plastic shearing and cutting with fracture, so that if the experimental point falls in the volume contained between EFGH and ABCD loss of material will be primarily by a fracture mode.

Fig. 2 illustrates how a diagram of this form can be used to predict the changes in removal mechanisms as the cutting conditions are varied. Suppose, in the first case, that the initial operating point, P, lies well within the region of plastic shearing. Increasing the depth of cut moves P in the direction of X so that the shear-fracture boundary is traversed. Drilling a piece of PMMA (Perspex) illustrates this transition dramatically; at low feed rates, i.e. depths of cut, a curly continuous chip is produced but above some critical value

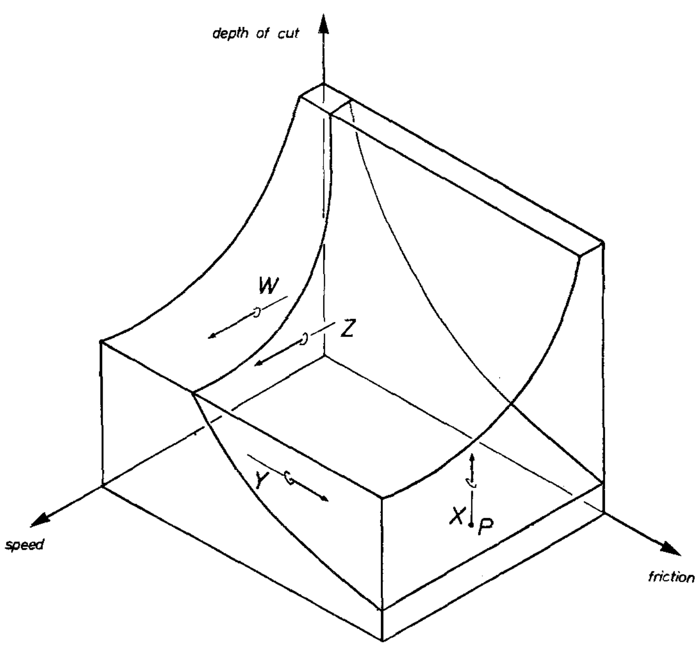

Figure 2 Effect of changing cutting parameters on mode of deformation. Point P represents one particular set of machining conditions; arrow X involves an increase in depth of cut, arrow Y an increase in chip-tool friction and arrows Z and W increases in the rate of deformation.

this behaviour ceases and lumps are simply knocked out of the surface [6, 7]. The same kind of transition can occur if there is an increase in chip-tool friction and this is suggested by the movement of a second operating point along the arrow Y: this accords with the evidence that the initiation of cracking and hence the production of discontinuous chips can be delayed by improved lubrication [8], i.e. by the maintenance of low chip-tool friction. Arrow Z suggests a transition from fracture mode to adiabatic shear and this has been observed [9] in tests in which high strength steel specimens were cut at varying speeds. At low cutting rates the chip was discontinuous and ragged, the machined surface rough and uneven and gave every indication of having been produced by some mechanism involving fracture. As the cutting speed was increased so cutting became smoother; the operating point moving in the direction of arrow Z. In excess of a speed of about $200\,\text{mm sec}^{-1}$, examination of the chip showed it to have the strong segmental form characteristic of the cyclical process of adiabatic shearing, and this corresponds to crossing the surface ABCD. In other materials increasing the cutting speed can bring about a transition from plastic to adiabatic shearing shown by arrow W. In low carbon ferrous materials this transition appears not to occur until superfast machining speeds are approached [10], i.e. greater than $3000\,\text{m min}^{-1}$; however, in material exhibiting limited strain-hardening and poor thermal conductivity, e.g. titanium, the onset of adiabatic shear will occur at much lower speeds; Recht [11] suggests as low as $300\,\text{mm min}^{-1}$.

It must be borne in mind that Fig. 2 is essentially diagrammatic having been drawn to illustrate four possible changes in chip formation, and it is for this reason that the axes have not been calibrated. Nevertheless it does suggest the form that such a diagram might take, although in practice other variables such as tool rake angle might be included. From the data on the machining of both metallic and non-metallic materials available in the literature, it should prove possible to assemble diagrams for specific materials, and these could be used both to gauge the effect of specific changes in cutting conditions and to identify those areas of material behaviour on which future work might usefully be focused.

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Comment on "On the background damping in the vicinity of the grain-boundary peak in zinc"

Burdett and Wendler [1] in a recent publication have given some results on the amplitude-dependent damping of zinc. The data were interpreted in terms of a movement of solute atoms from grain boundaries to dislocations and a value of the binding energy between solute and dislocations was obtained by applying the Granato–Lücke [2] theory.

It will be shown that some of the assumptions made by Burdett and Wendler are incorrect and consequently the value given for the binding energy is doubtful. In fact, the data from Figs. 7 to 9 have been plotted as $\log \delta_H$ versus $\log \gamma$ so that each curve can be compared directly with the theoretical expression for torsion [3, 4]. The experimental points and the part of the theoreti-