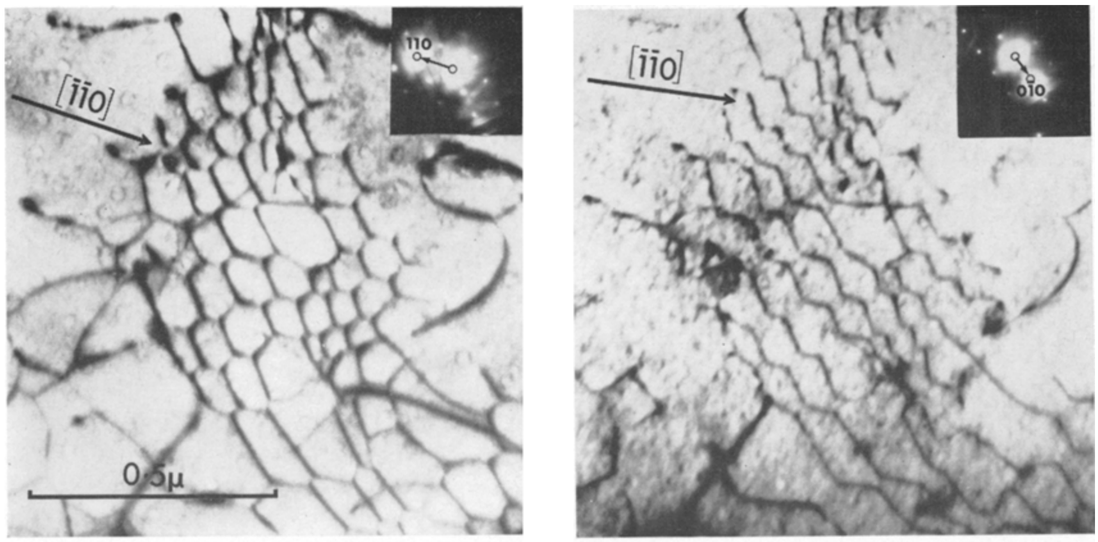

(a) (b)

Figure 2 Hexagonal dislocation network in a specimen deformed at 1400° C to 0.1 plastic strain. (a) with **g** = (110) (b) with **g** = $(0\bar{1}0)$. (Diffraction patterns corrected for image rotation.)

that a similar method is also applicable to other fully-dense polycrystalline ceramics.

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A Note on the Fracture of Polyester Resin

Armed with the ideas of fracture mechanics, designers are able to approach with more equanimity than hitherto the problems of designing with brittle and potentially brittle materials. Conventional methods of studying ductile/brittle behaviour – Charpy and Izod tests for example – have always been regarded with some suspicion, and these are being relinquished without regret in favour of tests based on the more formal stress-intensification and crack propagation notions. In comparison with the old impact tests, which are mechanistically more complex, the newer methods are certainly more easily related to physical processes. Charpy tests and the like are notorious for their unreliability and for the fact that the properties measured by them

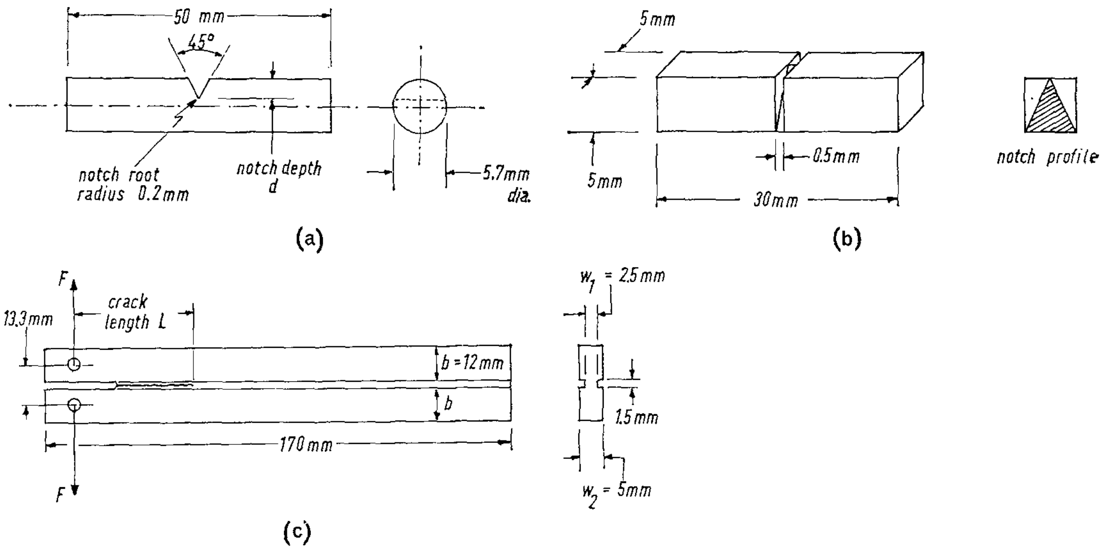

Figure 1 Specimen dimensions for the three types of test. (a) Charpy V-notch rod specimen. (b) Notched three-point bend specimen. (c) Double cantilever cleavage specimen.

are not fundamental material parameters. Scaling up to service sizes and conditions from Charpy data is regarded with apprehension by designers. But how much of this unreliability results from an imperfect understanding of the deformation process and how much is inherent in the test is difficult to ascertain.

Srawley and Brown [1] observe that the work of fracture, γ_F, of pre-cracked bars measured in a Charpy test may be related to G, the strain energy release rate during crack extension, only if the fracture surface is square and only if the crack extension resistance is constant during propagation of a crack through the specimen. They point out that these assumptions have not been thoroughly investigated and advise that Charpy tests should be used only as a method of preliminary screening of materials. We suggest that for certain types of material this may be an unduly pessimistic view.

We have measured the work of fracture of Bakelite SR 17449 polyester resin (resin: MEK peroxide: cobalt naphthanate = 100:2:2) in the following ways:

(i) By impact of V-notched rod specimens in a Hounsfield miniature Charpy impact machine.

(ii) By slow three-point bending of notched specimens of the form described by Tattersall and Tappin [2].

(iii) By cleavage of double cantilever beam samples in tests of the kind described in some detail by Berry [3].

For the first of these tests, notches of the shape shown in fig. 1a were cut with a sharp lathe tool. The energy absorbed from the pendulum during fracture, which is read directly from the machine, was determined as a function of notch depth. A second series of tests was carried out on similar specimens in which the notches had been sharpened with a scalpel. The fraction of pendulum energy absorbed was divided by twice the area of cross-section at the notch to give the work of fracture, γ_F.

For the second test, saw cuts were used to give a triangular test section as shown in fig. 1b. In three-point bending a crack starts at the apex of this triangle and propagates across the sample in the plane defined by the notch. The area under the load/deflection curve was converted to work of fracture, as before, by dividing by twice the area of the triangular section.

In the third test, crack growth is again controlled because the load falls with each increment of crack extension and, as shown in fig. 1c, the crack is forced to propagate in a predetermined plane by the shape of the sample. During the experiment the crack-opening force, F, the crack-opening displacement, δ, and the crack length, L, were measured and used in the expressions for fracture energy given by Gillis and Gilman [4]:

$$\gamma_F = \frac{6F^2L^2}{Ew_2b^3w_1} \text{ (from crack-opening force),}$$

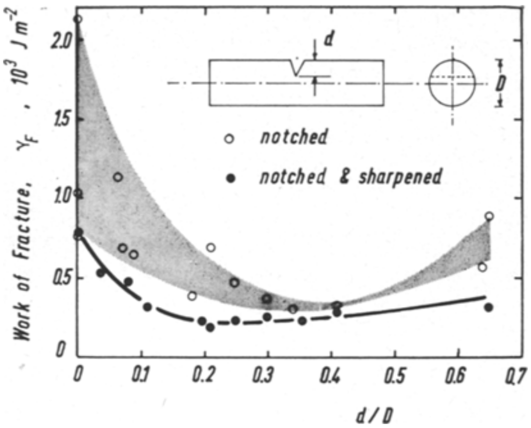

Figure 2 Work of fracture of polyester resin, as determined in Charpy tests, as a function of maximum notch depth.

$$\gamma_F = \frac{3w_2b^3E^2\delta^2}{8w_1L^4} \text{ (from crack-opening displacement),}$$

where E is Young's modulus and the other symbols are defined in fig. 1c.

Results from the Charpy tests are given in fig. 2. It can be seen that machine notching gives rise to considerable scatter, particularly at notch depths much less than a third of the rod diameter, and γ_F passes through a minimum at $d/D \sim 1/3$. Sharpening the notch roots eliminates scatter and reduces the measured values of fracture energy at all notch depths. Again there is a broad minimum between d/D values of about 0.2 and 0.35 where the mean value is about 220 J m^{-2}. The minimum in these curves corresponds to the prediction [5] from Neuber's analysis that the maximum stress concentration occurs for $(D - d)/D = 0.707$ and it is only at this point, therefore, that we can begin to think of γ_F in relation to G_{IC}, the critical energy release rate. The predicted variation of stress concentration factor, K_σ, with notch depth has been shown graphically by Wilshaw *et al* [6].

The minimum value of γ_F from the Charpy tests can be compared with values from the other two tests in table I: it can be seen that the agreement is very good.

For glass, γ_F is about 5 J m^{-2} [7] while for polystyrene it is about 1000 J m^{-2} [2]. For a resin such as polyester in which there are many cross-links, tending to cause glassy behaviour, and also residual polystyrene chains which are free to contribute some viscoelastic behaviour, a value of about 200 J m^{-2} seems quite reasonable. For comparison, Corten [8] has obtained a value of 1.6 MN m$^{-3/2}$ for the critical stress intensity, K_{IC}, for epoxy resin, and this gives a value of γ_F of about 500 J m^{-2} from the relation

$$K_{IC} = (G_{IC}E)^{\frac{1}{2}} \equiv (2\,\gamma_F E)^{\frac{1}{2}}.$$

On the other hand, measurements of γ_F for polyester resin by Irwin and Kies [9, 10] and by Broutman and McGarry [11] are much lower than 200 J m^{-2}, being 88 and 12 J m^{-2} respectively. However, Broutman and McGarry also give values of 43 J m^{-2} for epoxy resin and about 400 J m^{-2} for polystyrene, and these are also much lower than the usual values for these materials.

There are two interesting features of the results in table I. First, the rates of strain in the three tests are by no means similar. The rate of straining in the bend tests was about 10^{-3} sec^{-1}, whereas the rate of deformation at the root of a sharp notch in an impact test might be up to 6 orders of magnitude greater than this. Second, the values from the cantilever test refer to propagation of naturally-sharp cracks (i.e. subsequent to the initial extension of the machined notch) whereas in the Charpy test an artificial flaw is

TABLE I Work of fracture of polyester resin.

Type of test	Measured value of γ_F, 10^3 J m^{-2}	Mean value of γ_F, 10^3 J m^{-2}
Charpy test – mean, minimum value for samples with sharpened notches, $d/D = 0.3$	—	0.22
Three-point bend test with triangular notch	0.25	0.2
	0.15	
Cleavage (double cantilever)		
(a) from crack length	0.17	0.2
	0.23	
(b) from crack-opening displacement	0.30	0.22
	0.22	
	0.16	

propagated. The first point implies that the rate of propagation of the cracks in all of these tests was independent of and presumably much faster than the speed of the stress pulse. The second point is more significant and suggests that the minimum flaw radius that is needed in a Charpy test to obtain consistent results is by no means as small as that of the "natural crack" or the fatigue-induced crack usually regarded as indispensable. Wilshaw *et al* have measured experimentally the variation of critical stress intensity, K_{IC}, for mild steel at 77° K with notch root radius, ρ. They show that below a value of $\rho \sim 0.127$ mm, K_{IC} changes very little with further decrease in ρ. Their theoretical model predicts a linear variation of K_{IC} with $(\rho)^{\frac{1}{2}}$, but their experimental data fit a cube root better than a square root relationship. On the basis of the theoretical model, if $K_{IC} \propto (\rho)^{\frac{1}{2}}$, γ_F should vary linearly with ρ. The results for polyester resin do not fit the model either, for, as fig. 3 shows, the variation is more like $\gamma_F \propto \rho^{0.2}$. It is interesting to note that fracture energies for poly(methyl methacrylate) determined by Vincent [12] do indeed vary linearly with ρ down to a root radius of 0.025 mm, whereas for polycarbonate of bisphenol A the relationship is more like $\gamma_F \propto (\rho)^{0.24}$ – very close to that in fig. 3 for polyester – down to a value of $\rho = 0.25$ mm. Fig. 3 extrapolates back to the value of γ_F for a natural crack (from the cleavage tests) at a root radius of about 0.017 mm. This is presumably

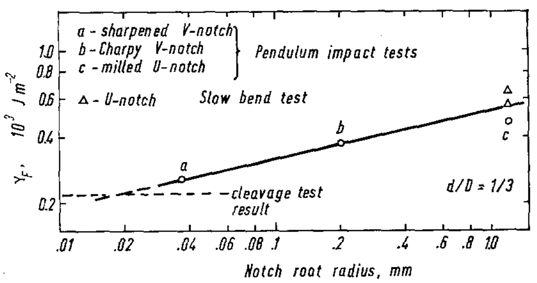

Figure 3 Variation of the fracture energy of a polyester resin with notch root radius. In all cases the notch depth was 1/3 specimen thickness.

the effective limiting sharpness, ρ', at which, according to Neuber [13], in the transition from a rounded to a sharp notch the stress concentration becomes a material function determined by microstructural features.

It appears, then, that Charpy tests can be made to give values identical with those from more elaborate tests, at least for polyester resin, without the large disagreement usually observed (see Tattersall and Tappin, for example) and without the universally-lamented inconsistency.

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