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A critical examination of the impact test for glassy polymers

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A model is described for the variation of impact strength with specimen dimensions. By comparison of its predictions with experiment, it is possible to decide whether the impact energy goes into crack initiation or propagation. In this way, it is shown that crack initiation is the major dissipation mechanism in a number of polymers including razor-notched polycarbonate.

1. Introduction

It has often been stated that impact strength is one of the least understood of the mechanical properties of polymers in spite of its great technological importance. This is partly because impact strength is not as well defined a mechanical property as, for example, modulus in that its definition includes a description of how it is measured. This means that the use of a standard specimen shape, although necessary to compare different materials, causes a severe limitation on the amount of useful information obtained on these materials. In this paper we demonstrate that it is possible to obtain information on the mechanisms of energy dissipation in the impact test by examining the notched impact strength as a function of specimen dimensions.

Two models for the mechanism of energy dissipation in the notched Charpy impact test are described in Section 2 of this paper. The first model, which assumes that all the impact energy goes into crack propagation, has been used for a long time. The second model, which is based on the concepts of fracture mechanics, assumes that the energy lost by the pendulum is that energy necessary to start the crack moving. It has not previously been described. The predictions for the variations of impact strength with specimen dimensions given by these two models are compared in Section 3 with experimental results measured on a number of systems. From these comparisons it is concluded that the notched impact strength of homopolymers is a measure of difficulty of initiating the moving crack.

2. Models

2.1. Model A

In this model it is assumed that the energy lost by the pendulum in a notched impact test is solely the energy required to form the two new surfaces as the material breaks. In addition, it is assumed that the surface energy is independent of crack speed and so the energy lost by the pendulum varies as the area of new surface formed. If G_c is the critical energy release rate for the material (equal to twice the surface energy) and A is the cross-sectional area of the specimen under the notch then the energy lost by the pendulum, U, is given by

$$U = G_c A \,. \tag{1}$$

This model has been used many times before [1] but not with great success.

2.2. Model B

When a cracked beam of a notch brittle material is fractured in a slow three- or four-point flexual test the energy supplied by the testing machine is normally just that required to deform the specimen elastically to its failure point. The material then fails catastrophically as this energy is sufficient to propagate the crack through the specimen. Charpy impact tests are high speed flexual tests so it is not unreasonable to assume that in these tests the energy lost by the pendulum is the energy required to deform the specimen sufficiently to cause failure. This is model B.

The loading geometry of a Charpy testing machine is shown in Fig. 1. It is possible to calculate the strain energy required to start

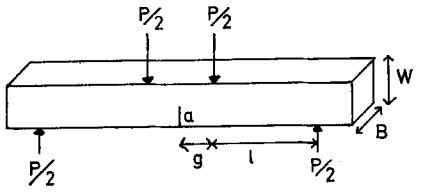

Figure 1 Specimen dimensions.

the crack moving if G_c is known for the material.

Suppose that the load required to fracture the specimen is P_0 and the specimen compliance (that is the movement of the loading points for unit load) is C, then the energy required to break the specimen, U, is given by

$$U = \tfrac{1}{2} P_0^2 C . \tag{2}$$

As the geometry of specimen is known, the load, P_0, can be related to the critical stress intensity factor K_{IC} and the compliance to the modulus of the material, E.

Brown and Srawley [2] have calculated the stress intensity factor for unit load, K_I/P, for this shape of specimen in the form

$$\frac{K_I}{P} = \frac{3Y(a/w)\, la^{\frac{1}{2}}}{BW^2} \tag{3}$$

where $Y(a/w)$ is a known function. We can use this expression immediately to find P_0 in terms of the critical stress intensity factor K_{IC}.

$$P_0 = \frac{K_{IC} BW^2}{3a^{\frac{1}{2}}\, Y\, l} . \tag{4}$$

The most convenient way to calculate the compliance of the specimen, C, is to again use the stress intensity calibration, Equation 3. This can be done because Equation 3 is a purely geometric relation obtained from a solution of the elasticity equations for the specimen shape and loading geometry. Irwin [3] has shown that the stress intensity factor can be related to the strain energy release rate by the equation

$$G = \frac{K_I^2}{E'} \tag{5}$$

where E' is reduced modulus, equal to E in plane stress conditions and to $E/(1 - \mu^2)$ in plane strain (μ is Poisson's ratio). In addition, the strain energy release rate can be related to the rate of change of compliance with crack length [4],

$$G = \frac{P^2 \mathrm{d}C}{2B\mathrm{d}a} . \tag{6}$$

These two equations can be combined to relate the stress intensity factor per unit load to the rate of change of the compliance with crack length

$$\left(\frac{K_I}{P}\right)^2 = \frac{E'}{2B}\frac{\mathrm{d}C}{\mathrm{d}a} . \tag{7}$$

By combining Equation 3, which came from an elastic solution of the particular specimen geometry, with Equation 7 which again is obtained from elasticity, we find

$$\frac{\mathrm{d}C}{\mathrm{d}a} = \frac{18\, l^2\, a\, Y^2}{E'B\, W^4} \tag{8}$$

and so by integration

$$C = \frac{18\, l^2}{E'BW^2}\int \frac{a}{W^2}\, Y^2\, \mathrm{d}a + C_1 \tag{9}$$

where C_1, the integration constant, is the compliance of the uncracked beam. From the literature [5], this can be found to be given by the expression

$$C_1 = \frac{l}{2BW^3E'}\left[4l\,(3g + l) + 3W^2(1 + \mu)\right] . \tag{10}$$

If Equation 9 is written in the form

$$C = C_1 + C_2 \tag{11}$$

then

$$C_2 = \frac{18\, l^2}{E'BW^2}\int \frac{a}{W^2}\, Y^2\, \mathrm{d}a$$

$$C_2 = \frac{18\, l^2}{E'BW^2} \times f\left(\frac{a}{W}\right) \tag{12}$$

where $f(a/W)$ can easily be calculated.

From Equations 2, 4, 5, 10, 11 and 12 we have

$$U = \frac{G_c BW^2}{18aY^2 l}\left\{18lf\left(\frac{a}{W}\right) + \frac{1}{2W}\left[4l\,(3g + l) + 3W^2(1 + \mu)\right]\right\} \tag{13}$$

The comparison of this expression with the results of experiment will be discussed in Section 4.

It is valuable to examine the two limits of Equation 13, namely when the crack is short hence $C_1 \gg C_2$ and when the crack is long and W is large so $C_2 \gg C_1$. Normally, the second term in Equation 10 is small, so if we ignore it and also assume that $C_1 \gg C_2$ and that $g = 0.267\, l$ (true for the Hounsfield H.20 plastics impact tester) we have

$$U \approx \frac{G_c BW\, 7.21l}{Y^2\, a\, 36}$$

but in this limit $Y \approx 2$ and the area of one new surface formed, $A = WB$,

$$\therefore U \approx \frac{G_c Al}{20a} \qquad a \ll W \ll l. \qquad (14)$$

This shows that for short cracks the energy-to-break varies as the inverse of the crack length, and as the area of specimen to be fractured. This predicts that the energy to break an "unnotched" specimen will vary as its cross-sectional area if the initial flaw size is constant and so it is not possible to distinguish between initiation and propagation in an unnotched specimen. If these initial flaws come from machining, this assumption is reasonable and so unnotched impact strength, defined as energy loss over cross-sectional area, is constant if the tester dimensions are fixed. This latter requirement is important and has not in general been stressed.

For long cracks and wide specimens, C_2 is dominant and we get

$$U = G_c \frac{BW^2 f(a/W)}{a\, Y^2 (a/W)} \qquad (15)$$

$$\therefore U = G_c \frac{A\, Wf(a/W)}{a\, Y^2 (a/W)}$$

which for $a/W = 0.5$ can be found to give

$$U \approx \frac{G_c A}{3}. \qquad (16)$$

It is interesting to note that Equation 15, unlike 14, is independent of the length of the specimen but its dependence on the other dimensions is more complex. It is assumed in this model that there is sufficient elastic energy in the specimen at the moment of fracture to propagate the crack through the specimen. The energy required to do this is given by model A to be $G_c A$ if G_c is independent of crack speed. It is immediately apparent from Equation 16 that when the specimen has a long initial crack there will only be sufficient strain energy present if G_c drops considerably with crack speed so that the energy required to form the new surfaces is less than $G_c A/3$. If this is not the case one would expect model B to predict the impact energy for short cracks but to give too small an energy for long cracks.

3. Experimental results

Measurements have been made of the energy to break an unsaturated ended urethane, polycarbonate of bisphenol A, amorphous polyethylene terephthalate (PET) of three molecular weights, high molecular weight crystalline PET, and ABS. For each material a number of samples were taken with different crack lengths and constant, W, and in the urethane four different values of W were also used. The samples were cut to rectangles then slowly notched using a razor blade. It is not claimed that this notching technique gives the "true" impact energy as it causes considerable plastic flow and crazing round the crack tip but it does give a consistent form for the crack tip.

The results are shown in Figs. 2 to 8 where the impact energy is plotted either against the area of the break, A, or against U/G_c calculated from model B.

If the graphs of measured U against calculated U/G_c are straight lines, then model B is being followed and G_c is given by the reciprocal of the gradient of the graph.

4. Discussion

From Figs. 2 and 3 it can be seen that in the urethane specimens the experimental results fit model B much better than A, although there is a tendency at the low energies (long cracks) for the measured energy to be larger than that calculated. This might be caused by G_c failing to drop sufficiently with crack speed, as has already been discussed. Alternatively, it might be because the kinetic energy stored in the specimen at the moment of fracture was ignored in the analysis. From these results, one can calculate the value of G_c and hence, knowing E, a value of K_{IC}. This is found to be 2.0 MN m$^{-3/2}$ assuming $E' = 3.2 \times 10^9$ N m^{-2}, a value obtained at low strain rates, which is to be compared with a value of 1.7 MN m$^{-3/2}$ obtained from a notched tensile test.

The notched PET and polycarbonate specimens were all found to fracture in the impact tests with very little microscopic deformation round the crack. The brittle nature of the failure is caused by the high speed of the impact test; Foot and Ward [6] found that in slow speed notch tensile tests specimens of medium and high molecular weight amorphous PET were not notch brittle.

The results of PET shown in Figs. 4, 5 and 6 can be seen to fit model B fairly well. This is

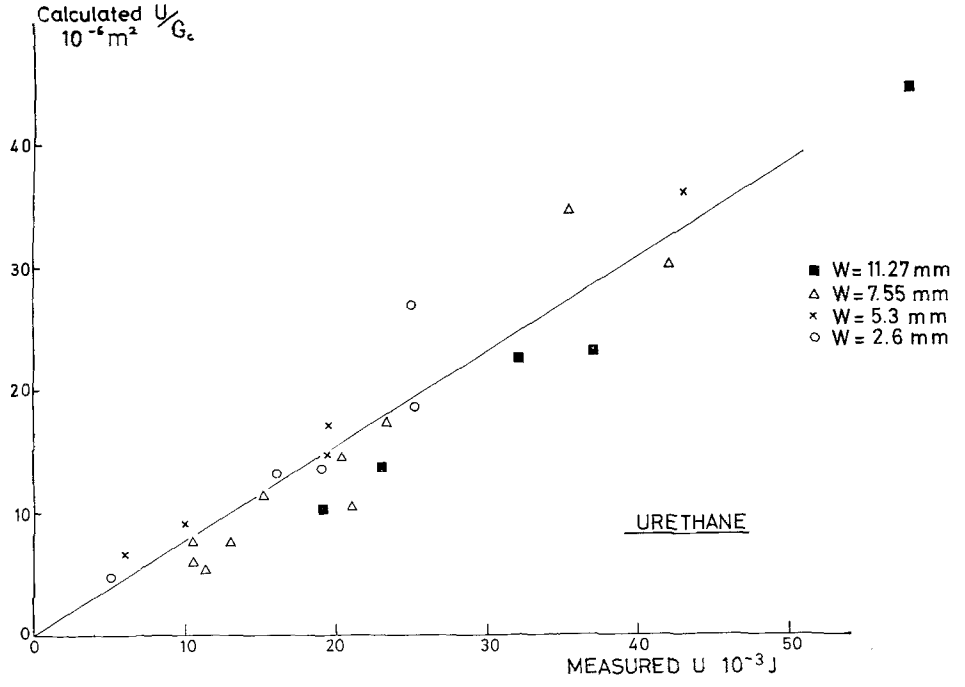

Figure 2 Calculated versus experimental fracture energy in vinyl-urethane.

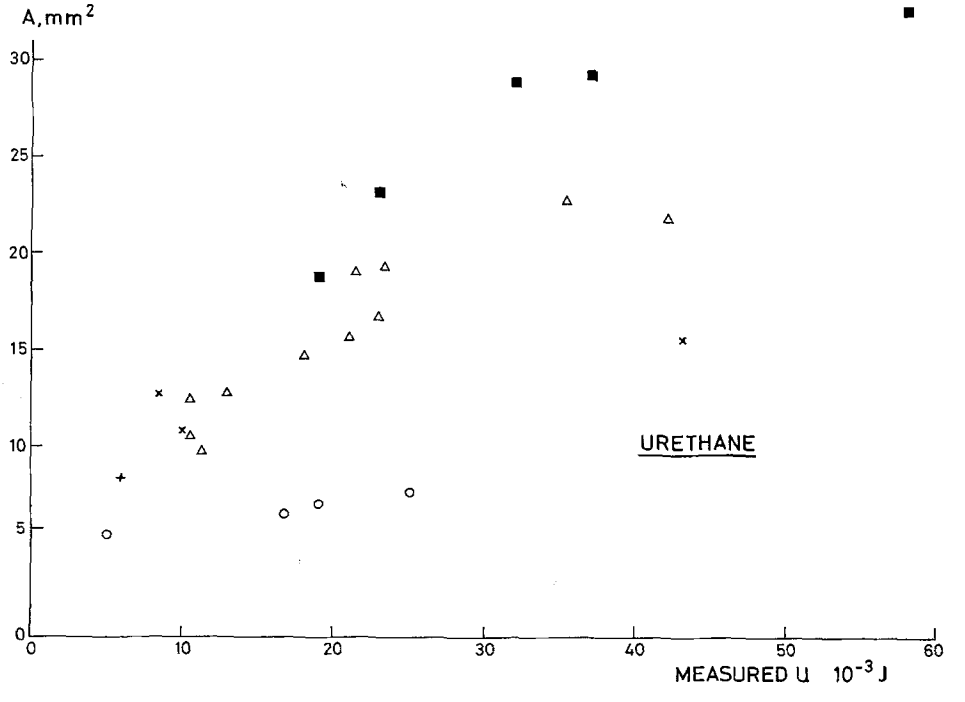

Figure 3 Experimental fracture energy versus specimen area, A, in vinyl-urethane.

perhaps not surprising as fracture mechanics in general applies well to brittle failure. The values of G_c are of the same order as those found by Foot and Ward [6] when they induced the materials to fail in a brittle manner. They found K_{Ic} in the range 1 to 5 MN $m^{-3/2}$ which

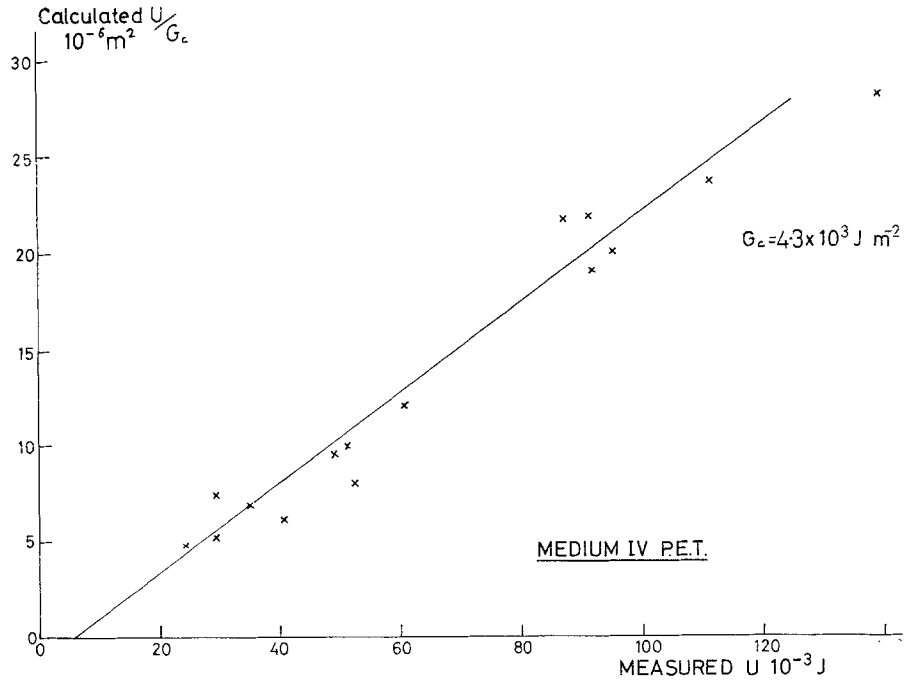

Figure 4 As Fig. 2 for medium I.V. PET.

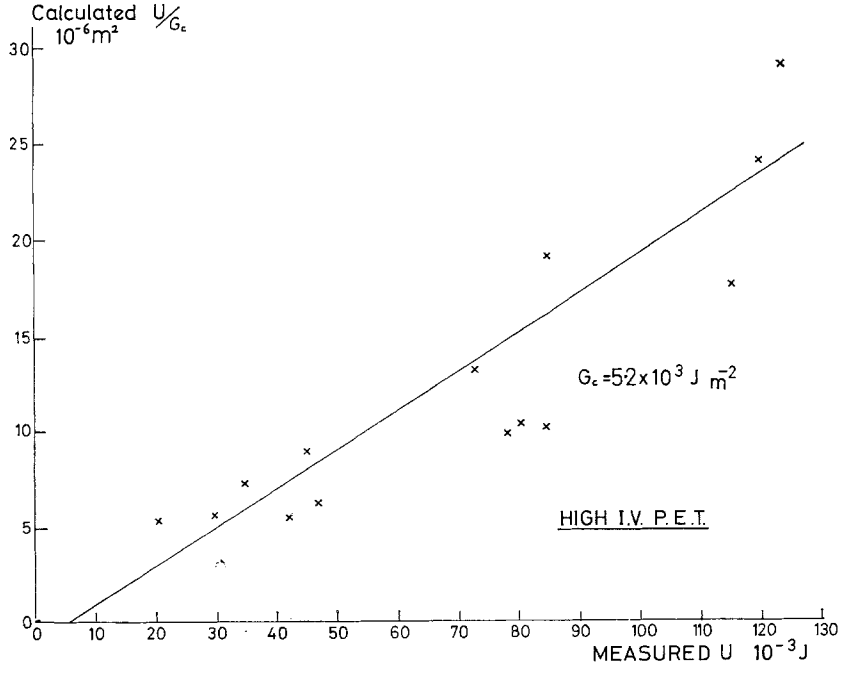

Figure 5 As Fig. 2 for high I.V. PET.

gives G_c equal to 0.4 to 10 kJm^{-2}. The trend of increasing G_c with increasing molecular weight agrees with that observed by Foot and Ward [6] as does the observation that crystallization has no great effect on G_c. The latter observation is interesting as the unnotched impact energy is much less in crystalline than in amorphous PET, in fact the amorphous

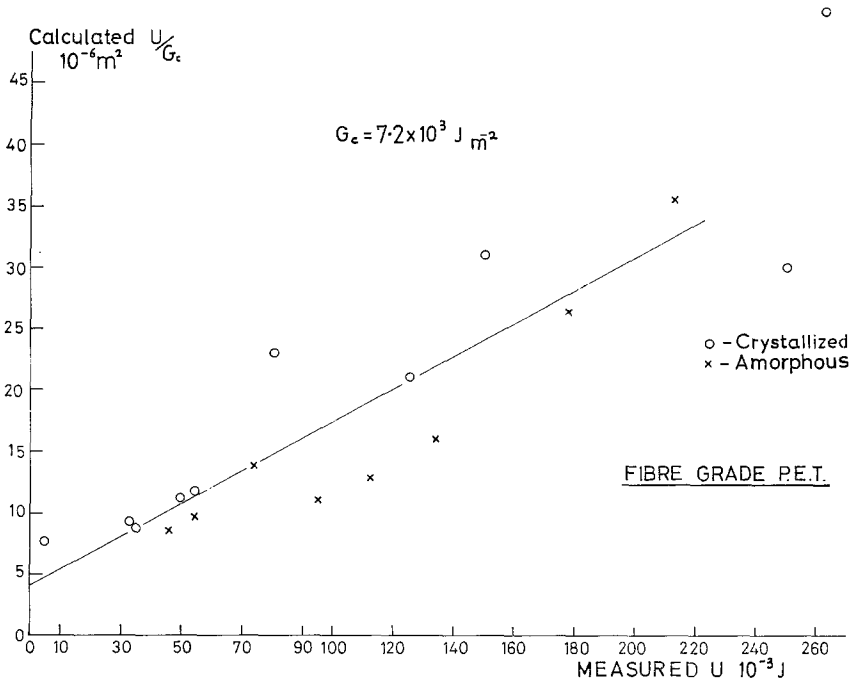

Figure 6 As Fig. 2 for fibre-grade PET.

material does not break unnotched in this test, so clearly the flaws are much greater in the crystalline material. It is very difficult to propagate a slow crack using a razor-notching technique in crystalline material as the crack tends to take off, so the energy to propagate high speed cracks must be extremely low.

It is interesting to note that G_c found in polycarbonate (Fig. 7) lower than that found in any of the grades of PET. This is perhaps surprising as polycarbonate is often said to be the toughest of the homopolymers but this is just a demonstration of the fact that toughness, meaning impact strength, is very dependent on the notch tip radius [1].

It is well known that polycarbonate has a "brittle-ductile" transition that can be caused by alterations of notch tip radius or temperature [7] or by annealing [8]. In this test the material is on the brittle side of the transition due to the sharp notch tip and the results obtained have no relevance to the situation on the ductile side of the transition. It is its high impact strength in the ductile failure mode that has given polycarbonate its reputation for toughness.

It is perhaps surprising that crack initiation is so important in razor-notched terylene and polycarbonate because this implies that the notched ASTM or BS impact strengths, which are measured in specimens with relatively blunt notches, must depend purely on crack initiation. This is in disagreement with the view expressed by other authors. [1]

It can be seen from Fig. 8 that ABS, unlike the homopolymers, fits model A reasonably well and so the energy is mainly expended in propagating the crack through the specimen. This is in agreement with the observations that hinge breaks are fairly common in this material and also that if too small a pendulum is used the crack will propagate a short distance then stop [9]

As model B fits the experimental results for such a range of glassy polymers, impact strength could be used as a quick and convenient method of measuring G_c which does not require the use of a tensile tester or other expensive equipment.

5. Conclusions

Two models for the impact strength of polymers have been discussed and their predictions compared with experimental results on a number of systems. A model which assumes that all the energy goes into elastically straining the material before the crack begins to move was shown to fit the experimental results on a number of

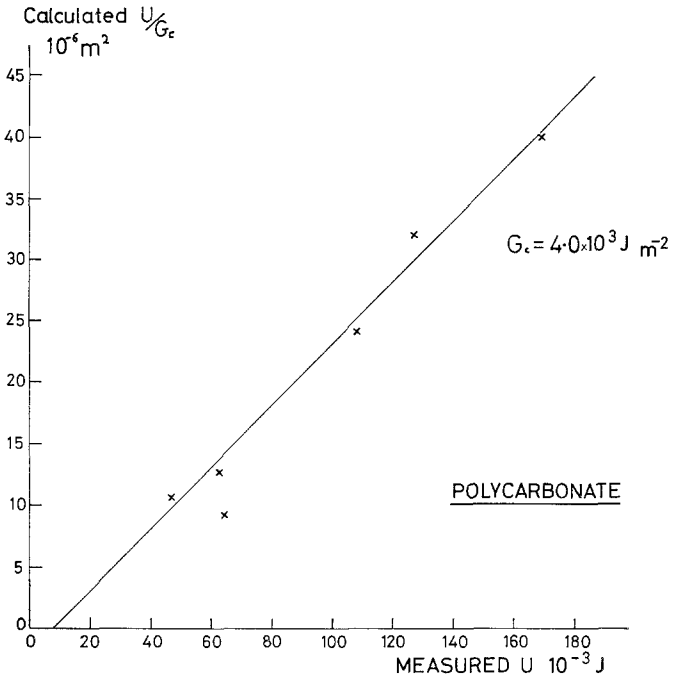

Figure 7 As Fig. 2 for polycarbonate.

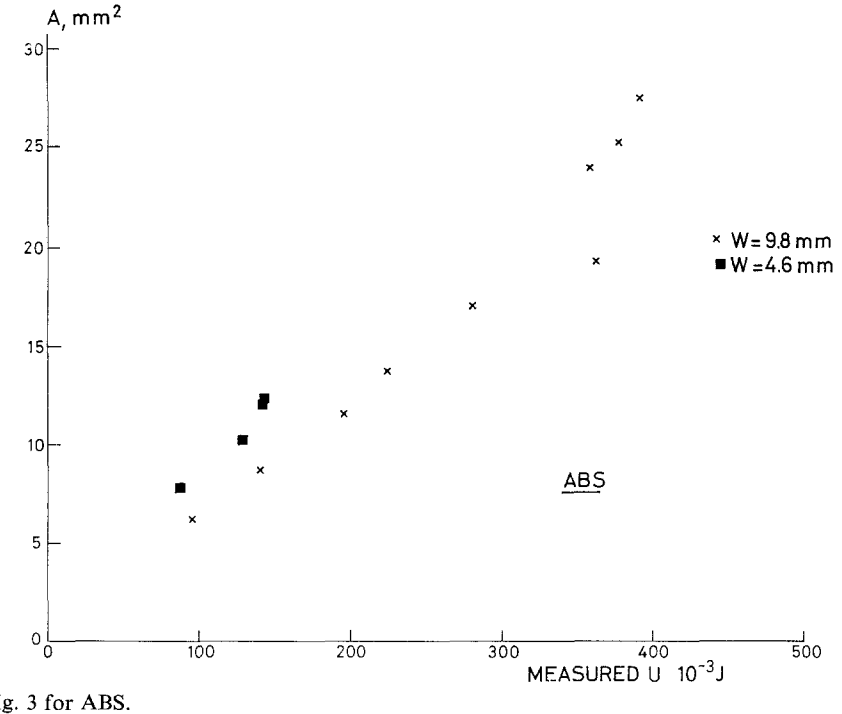

Figure 8 As Fig. 3 for ABS.

glassy homopolymers. On the other hand, in the "composite" material ABS the energy was shown to go into propagating the crack through the specimen.

It is concluded that the notched impact strengths normally quoted for polymers are a measure of the difficulty in initiating a moving crack.

Note

When this work was completed, the author learned of very similar work being conducted by Dr G. P. Marshall and his colleagues at the Department of Mechanical Engineering, Imperial College.

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