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Fracture from inherent flaws in polymers

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Fracture data on polymethyl methacrylate, polyvinyl chloride, polyacetal and polypropylene in both tension and bending are given for a range of temperatures using unnotched specimens. It is concluded that all the failures occur at the gross crazing or yield stress and are thus stress-controlled, and the flaw size has little effect. The propagation behaviour of the flaws under these conditions is discussed in terms of stability criteria.

1. Introduction

At first sight it is surprising that so little has been published on the consideration of failure from inherent flaws in polymers. After all, fracture mechanics was founded on such an analysis of inorganic glasses [1], and early work on the fracture of polymethyl methacrylate (PMMA) by Berry [2] included the calculation of inherent flaw size in unnotched specimens. He did point out, however, that the calculated values were unreasonably large and varied with temperature. It was concluded that these were not genuine, pre-existing flaws but more probably crazes which developed during loading prior to final fracture. Indeed the analysis of failures in inorganic glass and ceramics is far from simple, with the flaw shape and different types of crack growth complicating any size assessment [3]. While these problems exist for polymers they are not the major ones, and we shall consider the basic solution for an infinite cracked plate in order to illustrate the mechanisms involved. For a crack of length a the failure stress p is derived from

$$p^2 \pi a = K_c^2 \tag{1}$$

where K_c is the critical stress intensity factor, related to the energy per unit area of crack G_c by $K_c^2 = EG_c$. In tests to determine the toughness large values of a are used to give brittle fractures from which p is measured and hence K_c found. As a decreases then p at failure increases, and the limit on this is when p reaches the yield stress p_y of the material. When this occurs the flaw is nullified by plastic flow and failure will in general be by ductile tearing, controlled by the yielding condition. A lower limit on the flaw size for which brittle failure may be expected can thus be calculated from

$$\bar{a} = \frac{1}{\pi}\left(\frac{K_c}{p_y}\right)^2 \tag{2}$$

For inorganic glasses K_c is quite low ($\sim 0.7\,\mathrm{MPa\,m^{1/2}}$) and p_y is very high ($\sim 4\,\mathrm{GPa}$) so that a is extremely small ($\sim 10\,\mathrm{nm}$) and almost any inhomogeneity is sufficient to produce a brittle failure. For polymers the yield stresses are much lower, and indeed their greater toughness is mostly attributable to this fact. $\bar{a}$ is thus considerably larger and typically $K_c \sim 2\,\mathrm{MPa\,m^{1/2}}$ and $p_y \sim 70\,\mathrm{MPa}$ so that $\bar{a} \sim 260\,\mu\mathrm{m}$. This is quite large and usually considerably greater than debris and dust particles (~ 10 to $30\,\mu\mathrm{m}$) although some inclusions can be of this size and, of course, some damage can be greatly in excess of this figure. It is worth noting that Berry's calculation is essentially that given above so that his flaw sizes are, in fact, $\bar{a}$.

Much of the use of fracture mechanics has been confined to large cracks. It is particularly useful for characterizing toughness using pre-cracked specimens, and the results of this have been employed in design to consider the failure of damaged parts. When such damage is absent then the inherent flaws are usually sufficiently small that failure is by a ductile process, either

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shear yielding or crazing, and the failure loads may be predicted from yielding limit loads. Behaviour in this yielded state is not always the same, however. Ideally one would hope for the stable growth of the flaw with a gradual decrease in load but this does not always occur. In many cases quite small changes in rate and temperature can result in the ductile tearing changing to brittle, unstable fractures which are potentially dangerous. The dominance of crazing over shear yielding will often induce such a transition, and this is important since the "safe" state of ductile yielding is rendered unsafe and brittle failures result.

It is clear, therefore, that brittle failures from small inherent flaws do occur even at gross yielding, and from a practical design viewpoint it is important to have some knowledge of their nature and also to be able to predict how failures progress from them. To this end a series of experiments was performed on several polymers in which failures in unnotched specimens were observed, and this paper describes their interpretation in terms of fracture mechanics.

2. Experiments and calculations

A very large number of tests were performed in both tension and bending, varying both strain rate and temperature. The specimens were machined from sheet and the edges carefully polished to remove machining marks. In fact the majority of failures, even in tension tests, did occur from the surfaces (both machined and cast) and no degree of polishing seemed able to stop this. PMMA specimens soaked in water, which gave surface plasticization, did give almost all internal failures but they showed no marked difference in stress level or surface appearance from the edge failures. Details of the various methods employed will be found elsewhere [4], but here it is sufficient to note the types of failure obtained. The materials tested were PMMA, uPVC*, rubber-modified PVC, polyacetal and PP†, and they were tested in the temperature range −170° C to +20° C. In all cases the same general pattern of load–deflection curves was observed. At higher temperatures the load peaked and then decreased steadily as stable tearing proceeded across the section; this will be termed *ductile failure*. At lower temperatures the load peaked again but soon afterwards an unstable failure occurred with a sudden load decrease, resulting in a typical brittle fracture surface together with distortion in some cases, termed here a *semi-brittle failure*. At lower temperatures still the unstable failure occurred just prior to maximum load and there was no distortion, termed here a *brittle failure*. Examples of all three types are shown in Fig. 1 for uPVC tested in bending. Not all of these types occurred in all the materials, and increasing strain rate tended to promote brittle behaviour.

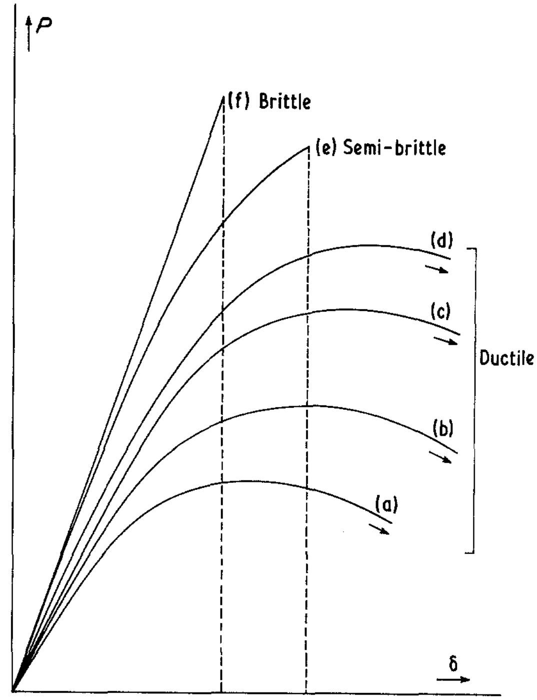

Figure 1 Typical load–deflection curves for unnotched uPVC specimens at temperatures of (a) +20° C, (b) −20° C, (c) −60° C, (d) −80° C, (e) −100° C, and (f) −140° C (D = 6.25 mm) in the flexure test.

In all tests the surface of the fracture was viewed using microscopy [5] and an estimate of the size of the initial flaw was made. This is a very problematical process and it involved much subjective judgement. In PMMA at 20° C, for example, the surface failures emanated from a rather rough region but outside this there is a distinct arc which appears to be a transition to rapid crack growth, and this was taken as the flaw size at failure. For internal failures the

*uPVC = unplasticized PVC.
†PP = polypropylene copolymer.

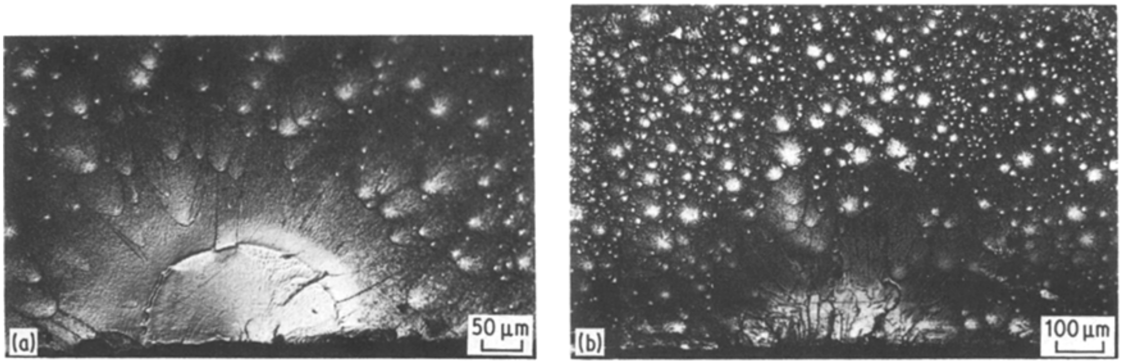

Figure 2 (a) PMMA at 20° C, surface failure; (b) PMMA at −40° C, surface failure.

transition lines could be seen but the initiation point was indeed a point and no increase in resolution could assign a size to it. At lower temperatures the distinct line was not present and the size was taken as the limit of the roughened region. In other materials such as PVC and polyacetal there were a number of internal failures, and here a well-defined inclusion could be identified and measured (usually a piece of unprocessed material) but there was no clear transition with the surface becoming progressively rougher. The size of the transition to gross roughening was taken as that corresponding to final failure. Figs. 2 and 3 show some examples of failure sites and illustrate the difficulties of making such measurements. It should also be noted that the shape of the flaw can be important in size calculations, and some estimate of degree of ellipticity can be made from the surfaces.

The failure stresses in simple tension were calculated in the usual way as load at fracture over area. For the bend tests there are several possibilities to consider, depending on the degree of yielding and the position of the failure in the section. If the fracture is judged to be elastic then the elastic solution assuming a linear stress distribution should be used

$$p_e = y\frac{3PS}{BD^3} \tag{3}$$

where y is the distance from neutral axis, P is the fracture load, S the span, B thickness and D depth, of the specimen (for three-point bending).

If, however, the section has yielded then the stress is constant in each half of the section and is given by

$$p_p = \frac{PS}{BD^2} \tag{4}$$

For a surface failure $y = D/2$ and $p_e/p_p = 3/2$.

For the ductile and semi-brittle failures it is clearly reasonable to use Equation 4 since the failures are post-plastic yielding, but for the brittle cases the situation is not obvious. The load increases by a factor of 1.5 from first yield to full collapse and it is not clear how much yielding has occurred prior to maximum load. Fig. 4 shows some data on brittle fractures in

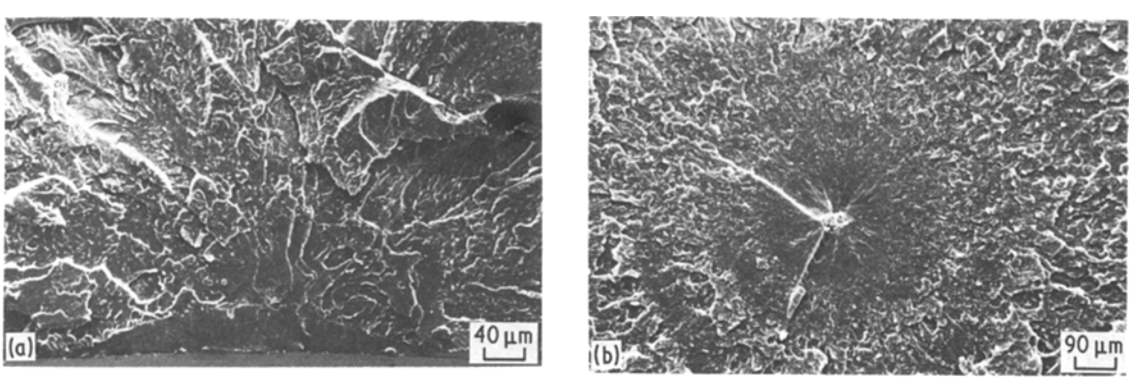

Figure 3 (a) uPVC at −120° C, surface failure; (b) uPVC at −120° C, internal failure.

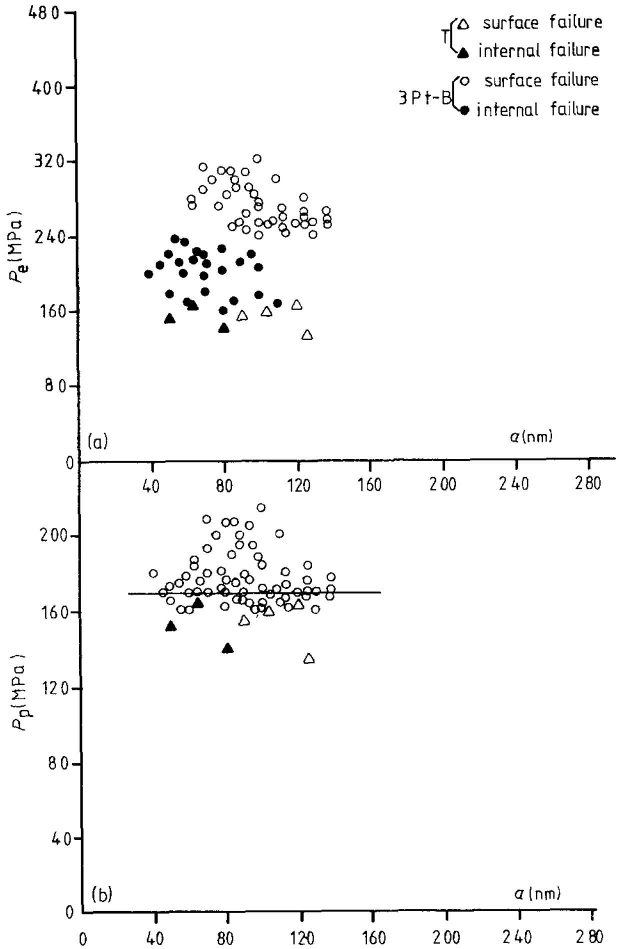

Figure 4 Variation of the fracture stress with slow crack growth radii: (a) elastic, (b) plastic collapse for uPVC at -120°C. T = tension tests, 3Pt-B = three-point bending.

uPVC plotted as failure stress against surface feature size calculated using the two limiting assumptions, and it is clear that the plastic case gives quite good agreement even though the tension data are rather lower and internal flaws are below surface ones. The use of the fully plastic bending solution gave similarly good agreement for all the materials for all the types of failure, although a discrepancy remained in PMMA at low temperatures.

This result is important since it implies that all the failures are occurring at, or close to, the fully yielded condition even when they are brittle, and that an apparent factor of 1.5 between bending and tension is due to this. The main motivation for testing in bending was to consider the statistical effects of flaw size distributions as described by the Weibull theory [6] which predicts differences between bending and tension because of the different stress states. There is no significant evidence of such an effect in the data reported here. It should also be noted in Fig. 4 that the failure stress is essentially constant even though the surface flaws varied from 40 to 140 μm, again suggesting that the fracture is stress-controlled. The calculated size limit $\bar{a}$ is 194 μm (K_c = 4.2 MPa m$^{1/2}$, p_y = 170 MPa) which confirms this condition. This was true of observations made on the brittle fractures in all the materials tested.

A further series of tests was performed in which K_c was measured as a function of temperature for all the materials using notched three-point bend tests. The details of the test-method can be found elsewhere [4] and will not be given here. The results will be given, however, since they are used to compute the flaw sizes from the fracture stresses and are a measure of the inherent toughness of the material.

3. Results and discussion

The results obtained are summarized in Figs. 5

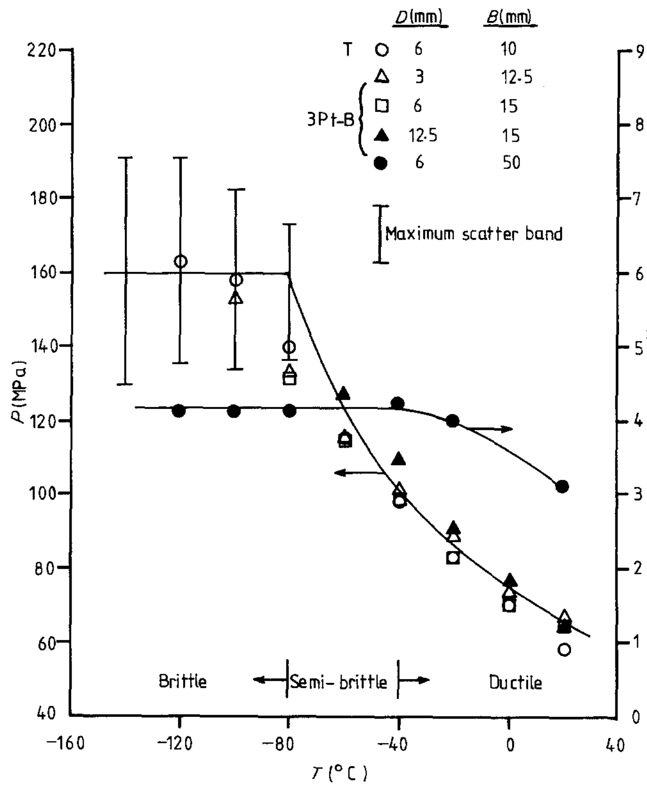

Figure 5 Fracture data on uPVC. T = tension tests, 3Pt-B = three-point bending.

to 9 and the general pattern is remarkably similar. In the ductile and semi-brittle region there is very little scatter in the data, and close agreement between tension and bending data. For PMMA at $T < 20°$ C the failures were semi-brittle and the tension data were above those for bending (Fig. 9). The domination of crazing over shear yielding in this material may account for this difference, and the elastic solution in Fig. 9 shows that the failure is most likely at some intermediate state prior to crazing across the whole section. (At most temperatures 10 to 15% of the depth is at the craze stress.) The onset of brittle fractures gives much more scatter ($\pm 20\%$) in the failure stresses but the agreement is generally good on the basis of full plastic yielding. As pointed out previously, however, there is no evidence of the failure stress varying systematically with observed flaw size. The calculated critical flaw sizes $\bar{a}$ (as computed from the K_c values) remain quite large ($> 100\,\mu$m) even down to the lowest temperatures, and are considerably greater than the observed origins of the fractures. For all three types of failure, therefore, the fracture is *stress-controlled* at the yield (or craze) stress but clearly the nature of the fracture process is different.

The most likely explanation appears to be that the controlling factor is the *stability* of the fracture. In all cases the initial flaw is sufficiently small that the fracture condition for it is only reached when the section is mostly at the yield condition, though the larger scatter in the brittle failures would suggest instability before full plasticity. If the flaw grows and is always stable then we have the ductile fractures noted previously. If they grow and then reach a critical size then we have the semi-brittle case, but if they are unstable immediately they start to grow then we would have a brittle failure. In the latter case this would result in the instability preceding full plasticity in bending, thus giving increased scatter.

Some insight into the mechanisms may be obtained by considering the "tearing modulus" theory of plastic fracture instability [7, 8]. This is based on the simple notion that the instability occurs when the elastic unloading is sufficient to effect propagation, and is expressed in terms of the applied modulus T_a and the material value T_m. T_a is found simply from the elastic recovery of the plastically loaded specimen and is approximately equal to L/D for simple tension (L is the specimen length). The material property

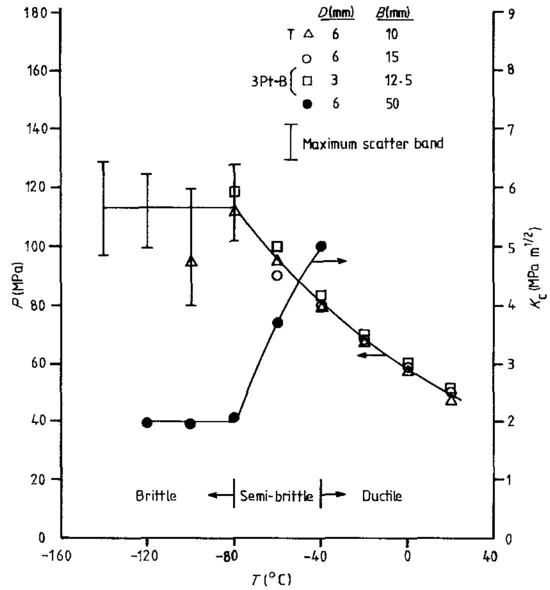

Figure 6 Fracture data on modified PVC. T = tension tests, 3Pt-B = three-point bending.

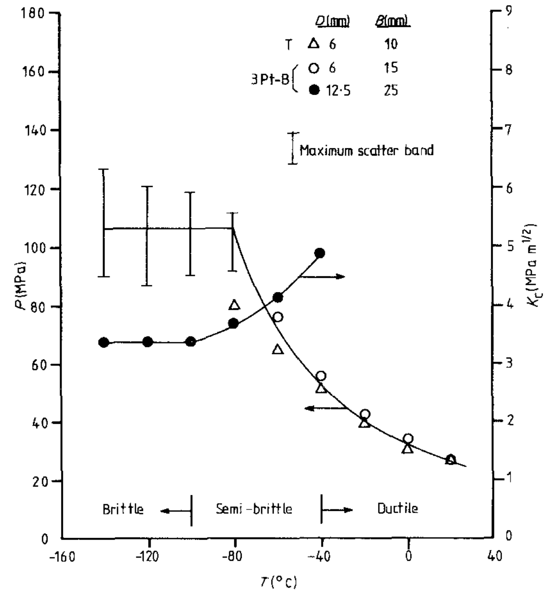

Figure 8 Fracture data on PP. T = tension tests, 3Pt-B = three-point bending.

is given by

$$T_m = \frac{E}{p_y^2}\frac{dJ_c}{da} \quad (5)$$

where J_c is the ductile fracture criterion and $J_c = G_c$ for elastically controlled failures. dJ_c/da is the slope of the resistance curve, i.e. the increase in J_c as the crack grows so that a large T_m represents a high resistance to crack propagation. For ideally brittle failures $dJ_c/da \simeq 0$ and stable failures are only obtained for special, constant G configurations. For the simple flawed sheet as tested here we can only achieve stable crack growth for positive dJ_c/da values, since the driving forces T_a are positive. This is equivalent to the stability induced by viscoelasticity where $dJ_c/da > 0$ to give stable growth [8]. In PMMA, for example,

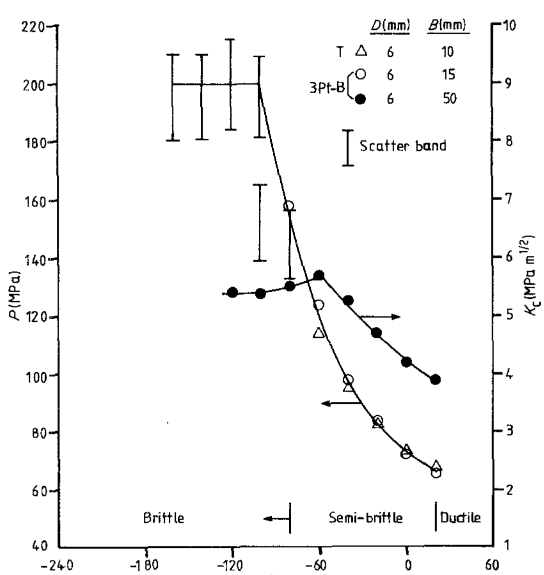

Figure 7 Fracture data on polyacetal. T = tension tests, 3Pt-B = three-point bending.

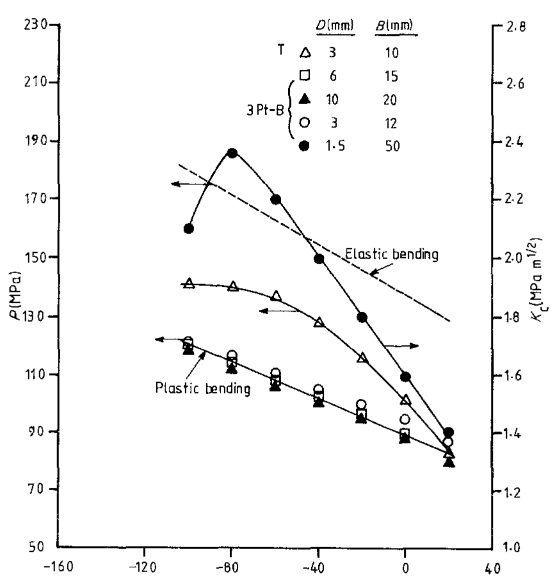

Figure 9 Fracture data on PMMA. T = tension test, 3Pt-B = three-point bending.

the craze stress is less than that for shear yielding below room temperature but there is a strong β transition giving stable craze growth prior to fracture and hence semi-brittle behaviour. In the other materials tested here there is substantial shear yielding at temperatures greater than about $-80°$ C, and stability arises from ductility effects. Below this temperature crazing is dominant but without any significant visco-elasticity, so that $\mathrm{d}J_c/\mathrm{d}a$ is small and the fractures are unstable.

4. Conclusions

It is apparent from the results given here that for this rather wide range of polymers the behaviour is remarkably similar. For specimens which do not contain substantial flaws, i.e. none greater than about 250 μm, then the failure is controlled, not by the flaw size, but by the stress level of the deformation mechanism; either crazing or shear yielding. Thus the failures are *stress*-controlled and not determined by the flaws present. In bending this controlling stress level acts as a plastic collapse condition and there is no evidence of a flaw-size distribution effect.

The major difference in the behaviour of the failures is not in the controlling criteria but in the stability of the failure which is governed by an additional material parameter, the resistance or R curve. For stable failure we need an increasing resistance to fracture and a steep R curve which can be characterized by T_m, the tearing modulus, through $\mathrm{d}J_c/\mathrm{d}a$. In these unnotched tests the driving force T_a is positive, so that if T_m is low then the failure will tend to be unstable giving abrupt failure. This pertains at low temperatures, but at higher values the presence of viscoelastic transitions and shear flow leads to an enhanced resistance to crack propagation, a higher T_m, and hence the failure is likely to be stable and classified as ductile.

In these tests the toughness was characterized only by K_c (or J_c) and no measures of $\mathrm{d}J_c/\mathrm{d}a$ were made. It is believed that this additional parameter is essential to an understanding of the observations made here on unnotched samples, and work is in hand to develop such data [9].

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