

Tear strength of oriented crystalline polymers

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Three crystalline polymers; high- and low-density polyethylene and *trans*-polyisoprene; were prepared as sheets with varying degrees of molecular orientation. Measurements of the energy G_c required to propagate a tear along and across the direction of orientation are described. The tear energy was found to depend linearly upon the thickness, t , of the sheet, in all cases. This dependence is attributed to plastic yielding at the crack tip in a zone having a cross-sectional area proportional to t^2 . The dependence on thickness was greater for highly oriented sheets torn across the orientation direction. This is attributed to a greater extent of plastic yielding in the orientation direction, as indicated by model experiments. By extrapolation to zero thickness of sheet, values were obtained of threshold tear energy in the absence of large-scale plastic yielding. However, these values were still relatively large (1 to 50 kJ m⁻²), and they were directly proportional to the extension ratio imposed across the tear direction at the time of crystallization. Thus, for tearing parallel to the orientation direction, both the threshold strength and the additional contribution from plastic yielding were small, whereas for tearing across the orientation direction both were large. Values of the effective diameter of the tear tip for unoriented materials were deduced from the threshold strengths. They were about 35 μm for HDPE and about 90 μm for LDPE and TPI; about five times the average spherulite diameter.

1. Introduction

Although partially crystalline polymers are widely used (for example as packaging films) because of their high strength and toughness, relatively little quantitative work has been published on their resistance to tearing and its relationship to molecular structure, crystallinity and orientation. In a previous paper [1], the effect of the thickness of a moulded polyethylene sheet upon the fracture energy for propagating a tear was shown to be quite large. Indeed, the fracture energy was found to increase almost in direct proportion to the sheet thickness [2]. This variation was attributed to a dependence of the volume of the plastic zone at the crack tip upon t^2 , where t is the thickness torn through. A major part of the fracture energy, in other words, was apparently consumed in yielding processes in a zone having a width of the same order as the thickness, t .

By extrapolation to zero thickness of sheet, values of fracture energy were deduced in the absence of large-scale plastic yielding, referred to as threshold values. They were found to be still relatively large, however, 1 to 20 kJ m⁻², and it was inferred that local yielding or other dissipative processes were still taking place in a thin zone, comparable in thickness to the individual semi-crystalline units (spherulites) of which the moulded sheets were composed [1].

We now turn to other aspects of the strength of semi-crystalline polymers. It is well-known that oriented materials show much lower resistance to tear propagation for a tear running in the direction of the orientation. Anderton and Treloar [3] observed a decrease for high-density polyethylene from about

80 kJ m⁻² in the unoriented state down to about 1 kJ m⁻² in the highly oriented state, and a similar marked reduction for low-density polyethylene. Sims [4] found similar effects and similar values to hold for polypropylene sheets.

Now it should be noted that these results were obtained with cold-drawn samples of uncrosslinked polymers. The exact state of molecular orientation is therefore somewhat uncertain. Also, no attempt was made to hold the sheet thickness constant and therefore some part of the observed decrease in strength with orientation may well have been due solely to the reduced thickness of oriented specimens in comparison with unoriented ones. Furthermore, no results were given for tearing at right angles to the direction of orientation. Thus, there is an obvious need for further examination of the tear strength of semi-crystalline polymers, especially in the oriented state. A study has now been carried out for three such materials: high density polyethylene (HDPE), low-density polyethylene (LDPE), and *trans*-polyisoprene (TPI). They were chosen for experimental convenience and, in particular, because it is possible to crosslink them lightly in the molten state to form elastic solids, rather than viscous liquids. Deformations can be imposed subsequently without any viscous flow taking place. The corresponding state of molecular orientation can then be inferred with some confidence.

After crystallization in the stretched state was complete, the sheets, now relatively rigid, were removed from the stretching device and torn in the direction of orientation or at right angles to it. Values of the fracture energy in each case are reported here as a

function of the degree of orientation for the three materials examined.

Highly oriented materials are presumably similar to fibres in structure and properties. The present results are therefore assumed to apply also to fibres.

2. Experimental details

2.1. Materials

Three partially crystalline materials were used: *trans*-polyisoprene (TPI), supplied by Polysar Limited, Canada, denoted TP-301, and having a molecular weight M_w of approximately 8×10^5 ; low-density polyethylene (LDPE) containing a few per cent of dicumyl peroxide, supplied by Nisseki Chemical Company, denoted W 2040, with a melt index of $1 \text{ g } 10 \text{ min}^{-1}$ and a density at room temperature of about 0.920 Mg m^{-3} ; and high-density polyethylene (HDPE) supplied by Asahi-Kasei Industries, denoted R340P, with a melt index of $7 \text{ g } 10 \text{ min}^{-1}$ and a density at room temperature of about 0.955 Mg m^{-3} .

Trans-polyisoprene was mixed with 1% by weight of dicumyl peroxide and 1% by weight of Antioxidant 2246 (American Cyanamid Company). The mixture was pressed into sheets and lightly crosslinked by heating for 1 h at 150°C . The crosslinked sheets were then cooled to room temperature and a 10 mm reference grid was drawn on the surface of each sheet. They were clamped between the jaws of an extending device, reheated to 95°C , well above the melting temperature, for 15 min and then rapidly stretched to a predetermined level and placed in a warm water bath at 40°C for about 20 h, until crystallization was virtually complete [5]. Samples were prepared in this way that had crystallized at 40°C whilst held at various extensions. The exact extension was determined by measuring the separation of grid lines on the crystallized sheets.

The low-density polyethylene material, containing dicumyl peroxide, was pressed into sheets and crosslinked by heating for 30 min at 140°C followed by 30 min at 160°C . Again, oriented sheets were prepared by extending the crosslinked sheets rapidly in the molten state and then allowing them to crystallize in the stretched state during rapid cooling (about $1^\circ \text{C sec}^{-1}$) to room temperature.

High-density polyethylene was mixed with 1% by weight of Antioxidant 2246 and then pressed into sheets as for LDPE. In this case the material was not crosslinked but, as shown later, the effects of orientation were found to be quite similar to those for the crosslinked materials.

Measurements were made of the latent heat of fusion and melting temperature for each sample using differential scanning calorimetry, with a heating rate of $0.17^\circ \text{C sec}^{-1}$. The degree of crystallinity, C , was calculated from the ratio of the observed heat of fusion to the reported values for 100% crystallinity: 44.5 cal g^{-1} (186 J g^{-1}) for TPI [6] and 69 cal g^{-1} (288 J g^{-1}) for polyethylene [7, 8]. The results are plotted in Fig. 1 as a function of the extension imposed during crystallization. No significant changes were observed in the measured melting temperatures or degree of crystallization with orientation imposed during crystallization. For TPI and LDPE, both

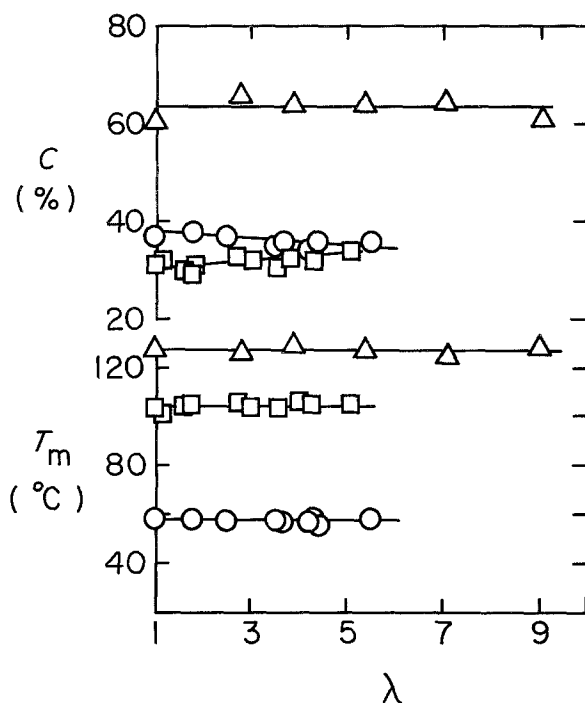


Figure 1 Effect of imposed extension ratio, λ , upon the degree of crystallinity, C , and melting temperature, T_m , for TPI (O), LDPE (\square) and HDPE (Δ).

crosslinked, the degree of crystallization was about 35%. For HDPE it was about 65%.

The yield stress also did not change significantly with the degree of orientation imposed during crystallization, being about 10 MPa for TPI, 10 MPa for LDPE and 25 MPa for HDPE. However, the maximum extension that samples would undergo before breaking was sharply reduced by orientation, decreasing from about 10 to about 1 for HDPE and from about 3.5 to about 0.5 for TPI as the extension imposed during crystallization was increased from zero up to the maximum level attainable before the melted samples broke, about ten times for HDPE and about three times for TPI.

2.2. Measurement of tear strength G_c

Testpieces, in the form of strips about 100 mm long and 20 mm wide, were cut from the oriented sheets both parallel and perpendicular to the direction of orientation. They were first scored along the centre line to a depth of about one-half of the total thickness, using a sharp razor blade, so that about one-half of the total thickness was left to be torn through. This procedure prevented gross plastic yielding of the entire testpiece in place of tearing, and also minimized deviation of the tear from a straight path.

Measurement of the tear force, F , was carried out as shown in Fig. 2, at a rate of tear propagation of about 0.4 mm sec^{-1} and at room temperature, about 23°C . Values of the fracture energy, G_c , were calculated from the average values of the tear force, F , and tear path width, t (measured subsequently by microscopy), using the equation [9]

$$G_c = 2F/t \quad (1)$$

Energy expended in bending or stretching the testpiece legs has not been taken into account in this equation

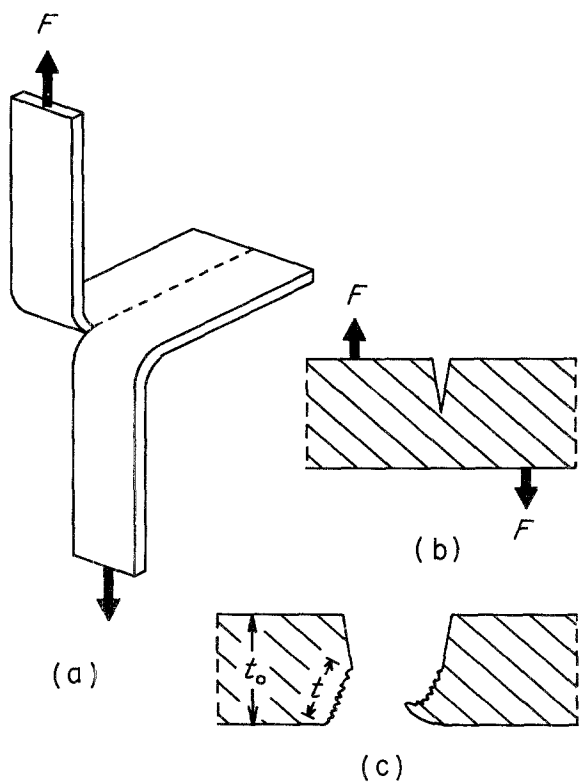


Figure 2 (a) Tear test. (b) Cross-section of testpiece. (c) Cross-section after tearing.

for G_c . Measurements of the bending contribution, reported elsewhere [10], indicate that it is a relatively small component of the total force for the testpieces discussed here.

3. Results and discussion

3.1. Effect of orientation on tear energy

Values of tear energy, G_c , for TPI samples are plotted in Figs. 3 and 4 against the extension ratio, λ , imposed during crystallization. Logarithmic scales are used for both axes. The results shown in Fig. 3 are for tearing parallel to the direction of orientation and those shown in Fig. 4 are for tearing at right angles to the

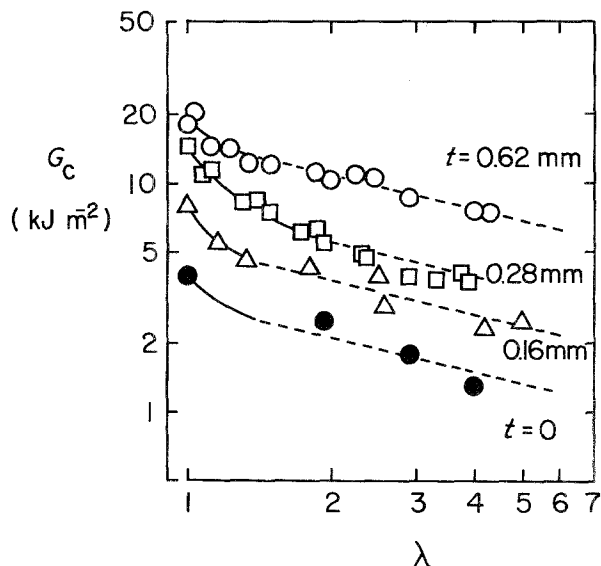


Figure 3 Fracture energy, G_c , for tearing TPI parallel to the direction of orientation, for various widths, t , of tear path. The results for $t = 0$ were obtained by extrapolation, Fig. 5.

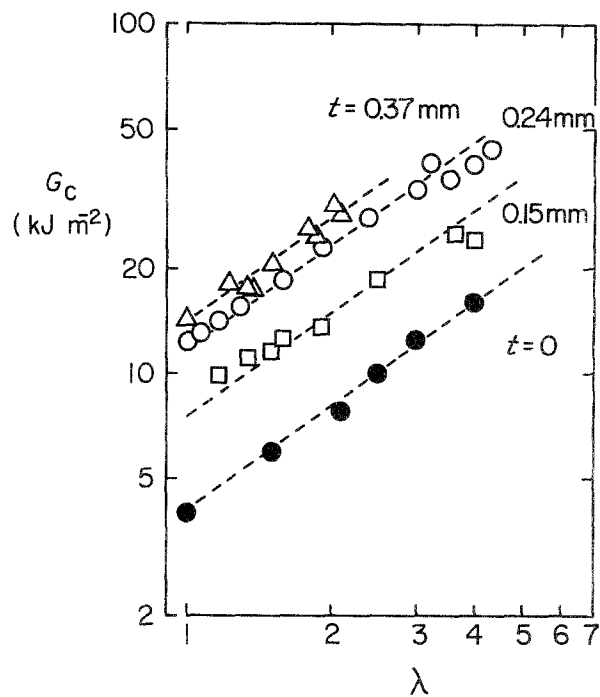


Figure 4 Fracture energy, G_c , for tearing TPI at right angles to the direction of orientation, for various widths, t , of tear path. The results for $t = 0$ were obtained by extrapolation, Fig. 6.

orientation direction. In the first case the tear energy decreases as the degree of orientation increases, as might be expected, and in the second case the tear energy increases strongly.

In both cases it should be noted that the fracture energy is calculated with reference to the oriented state, i.e. using the tear path width, t , in Equation 1 measured on the oriented strip. Referred to the unoriented state, the thickness, t_u , would be given by $\lambda^{1/2} t$ for tearing parallel to the orientation direction, and the torn length, L_u , by L/λ , where L is the length torn through in the oriented state. Thus, the area torn through, referred to the unoriented state, would be smaller by a factor $\lambda^{-1/2}$ [11]. The broken lines in Fig. 3 have been drawn with a slope of $-1/2$ and appear to describe the effect of orientation, at least for extension ratios greater than about 1.5, with reasonable success.

Similarly, for tearing at right angles to the direction of orientation, the area torn through is larger in the unoriented state by a factor of λ in comparison with the oriented state. The apparent tear energy referred to the oriented state would then be expected to increase by the same factor if there is no other effect than this simple geometrical change in the torn cross-section [11]. The broken lines in Fig. 4 have been drawn with a slope of $+1$, corresponding to a direct proportionality between G_c and λ . Again, they appear to account satisfactorily for the observed variation of the tear strength with orientation. Thus, the major changes in tear strength along and across the direction of orientation as the degree of orientation is increased are approximately accounted for in terms of corresponding changes in the area torn through with respect to the unoriented state. The intrinsic strength of the material does not seem to change substantially, at least for the levels of orientation employed here.

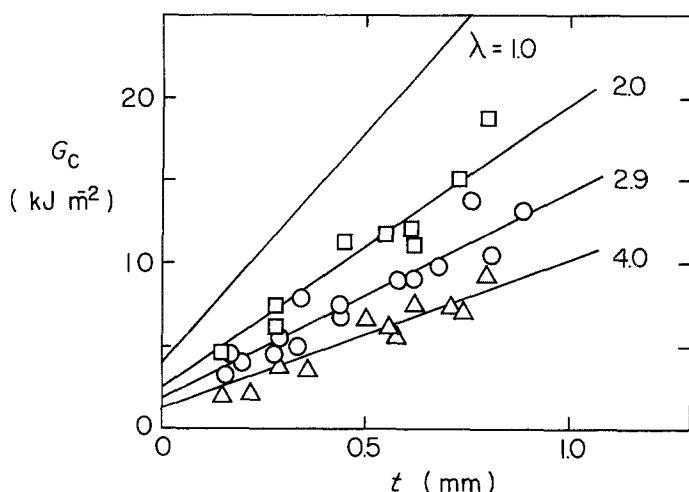


Figure 5 Fracture energy, G_c , plotted against tear path width, t , for TPI at various levels of orientation (extension ratio λ) parallel to the tear direction.

3.2. Effect of tear path width, t

Results are shown in Figs. 3 and 4 for testpieces of different thickness and, hence, tear path width, t , Fig. 2. At any level of orientation, thinner testpieces were found to tear at lower values of fracture energy. (A similar observation was made previously for unoriented samples [1].) In Figs. 5 to 8, the fracture energy is plotted against the tear path width, t , for samples of TPI and HDPE prepared at various orientations. In all cases a substantially linear dependence of fracture energy, G_c , upon t was found to hold. Thus, by extrapolation to zero thickness a threshold value of fracture energy could be determined, denoted $G_{c,0}$. These values are thought to represent the tear strength of the material in the absence of large-scale plastic yielding [1], and also in the absence of any contribution to the measured tear force arising from bending of the testpiece legs during tearing [10]. However, they are still relatively large, 1 to 15 kJ m^{-2} for TPI and 2 to 60 kJ m^{-2} for HDPE, so that a considerable amount of energy dissipation in processes other than rupture of a plane of C-C bonds can be inferred. (The latter process would only require about 1 J m^{-2} for a rigid material [12] and about 50 J m^{-2} for a highly extensible elastomeric material [13], in the absence of dissipative processes.)

It thus appears that the observed fracture energy

can be represented by the equation

$$G_c = G_{c,0} + (dG_c/dt)t \quad (2)$$

where both the intercept $G_{c,0}$ and slope dG_c/dt depend upon the degree of orientation of the sample. These two important quantities characterizing the tear strength of ductile materials will now be discussed separately.

3.3. Threshold tear energy

Values of the threshold fracture energy, $G_{c,0}$ are plotted in Fig. 9 against the extension ratio, λ , across the tear direction. For tearing parallel to the direction of orientation, the results have been plotted against (imposed extension ratio) $^{-1/2}$; thus, these points are represented by values of λ of less than unity. For all of the experimental materials, linear relations are obtained between $G_{c,0}$ and λ , passing through the origin. These relations are consistent with the simple geometrical factor governing area changes, discussed previously in terms of affine displacements. For example, the number of molecular chains crossing unit area would be expected to change in this way with an imposed deformation. Although the measured fracture energy is many times greater than the energy required for molecular rupture, nevertheless, it may

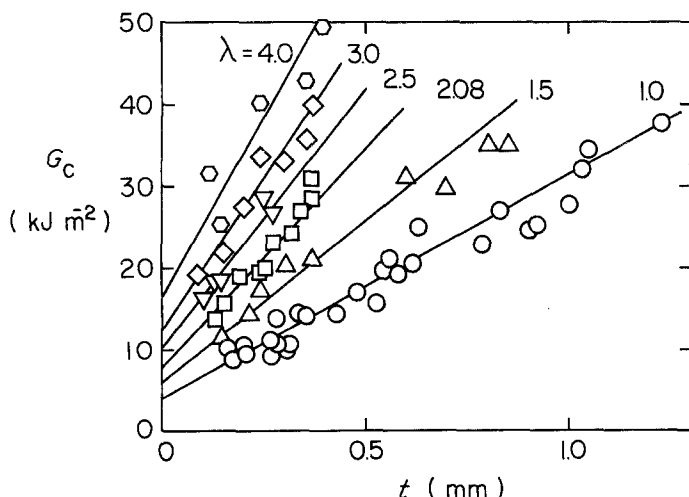


Figure 6 Fracture energy, G_c , plotted against tear path width, t , for TPI at various levels of orientation (extension ratio λ) across the tear direction.

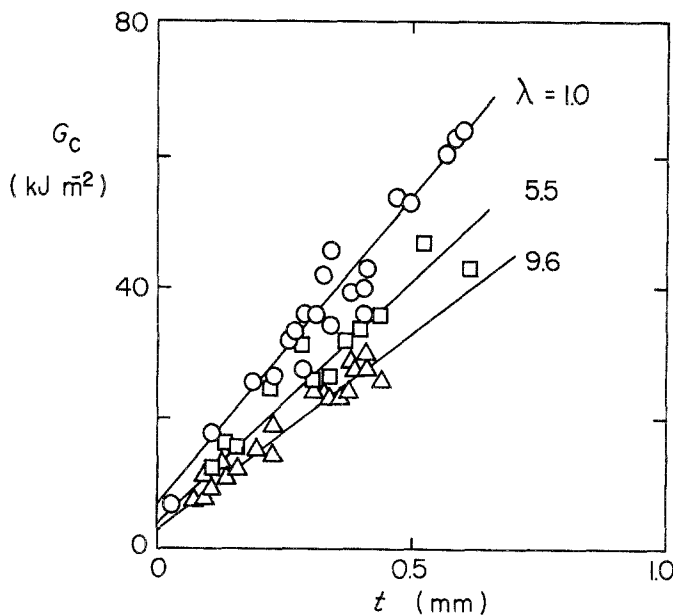


Figure 7 Fracture energy, G_c , plotted against tear path width, t , for HDPE at various levels of orientation (extension ratio λ) parallel to the tear direction.

well depend directly upon the number of chains crossing the fracture plane.

It is interesting to note that the fracture energies for the three materials; TPI, LDPE and HDPE; are rather similar in magnitude. They can be interpreted in terms of a microscopic tensile rupture process taking place at the tear tip, so that

$$G_{c,0} = d U_b \quad (3)$$

where d is the effective unstrained diameter of the tear tip and U_b is the work required to break unit volume of the material in tension [14]. Measurements were made of the work-to-break, U_b , using small dumb-bell tensile specimens of gauge length 15 mm stretched to break at a strain rate of about 1 sec^{-1} . The results are given in Table I, together with values of $G_{c,0}$ for the unoriented specimens and values of the effective tear tip diameter, d , calculated from them by means of Equation 3. The inferred values of d are about $90 \mu\text{m}$ for unoriented samples of TPI and LDPE and about $35 \mu\text{m}$ for HDPE.

It is interesting to compare these values with the diameters of spherulites observed in unoriented samples. For TPI, the spherulites were found to have an average diameter of about $15 \mu\text{m}$. For LDPE only

vague spherulites could be seen, with diameters in the range 10 to $30 \mu\text{m}$. For HDPE, small, well defined spherulites were visible, having diameters in the range 5 to $8 \mu\text{m}$. Thus, in all cases the estimated values of tear tip diameter are about five times the observed spherulite diameter, suggesting that the extrapolated fracture energy at "zero" thickness of test specimen is really the fracture energy for a specimen having a minimum thickness of material on either side of the fracture plane involved in dissipative yielding processes of the same kind as that observed in thicker specimens. This minimum layer thickness is apparently 1 to 3 spherulite diameters.

Values have been deduced previously for the effective diameter of the tear tip in low-density polyethylene, of about $300 \mu\text{m}$ [3], and in polypropylene of about $600 \mu\text{m}$ [4]. These are really measures of the extent of plastic yielding around the tear tip, however, because they were obtained from measurements on relatively thick specimens, 0.3 to 1.0 mm thick. They are much larger than those inferred here for thin testpieces, by a factor of about ten times. The extent of plastic yielding during tear propagation is discussed in more detail in the following sections.

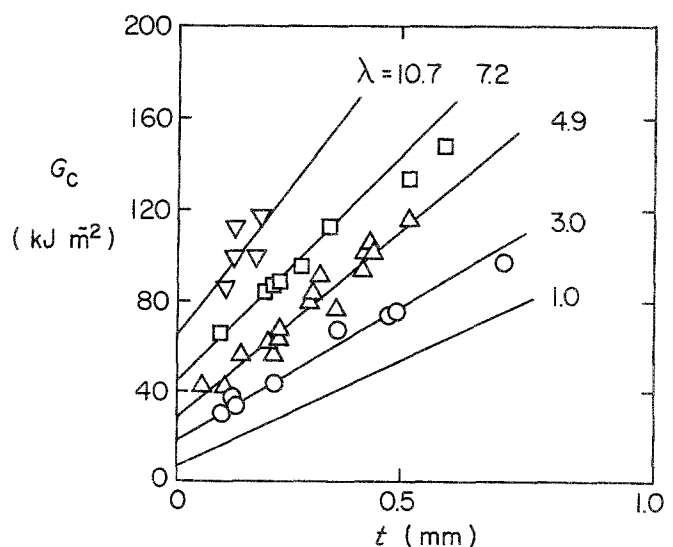


Figure 8 Fracture energy, G_c , plotted against tear path width, t , for HDPE at various levels of orientation (extension ratio λ) across the tear direction.

TABLE I Estimated tear tip diameter, d for unoriented sheets of TPI, LDPE and HDPE

Material	$G_{c,0}$ (kJ m ⁻²)	U_b (MJ m ⁻³)	d (calculated from Equation 3) (μ m)
<i>Trans</i> -polyisoprene (TPI)*	4.0 ± 1.0	45 ± 6	90 ± 35
LDPE*	6.0 ± 1.5	65 ± 5	90 ± 30
HDPE	6.5 ± 2.0	180 ± 20	35 ± 15

* Crosslinked with dicumyl peroxide.

3.4. Plastic zone sizes

The linear relations observed between the fracture energy, G_c , and tear path width, t (Figs. 5 to 8), can be interpreted in terms of a direct proportionality between the effective size of the plastic zone at the tear tip and the width of the tear path:

$$dG_c/dt = \alpha U_b \quad (4)$$

where α is the ratio of the effective diameter of the plastic zone to the tear path width and U_b is the work expended plastically per unit volume, assumed here to be approximately equal to the work-to-break [1]. Values of dG_c/dt taken from Figs. 5 to 8 are listed in Tables II and III and are shown graphically as functions of the extension ratio λ across the tear direction in Figs. 10 and 11. They are seen to depend strongly upon the extension ratio λ , indicating a much greater dependence of the fracture energy upon thickness of the tear specimen for highly oriented samples, torn at right angles to the orientation direction, than for unoriented ones. This dependence also appears to be more pronounced for HDPE than for TPI. It does not reflect changes in the work-to-break, U_b , with orientation, however, because, as shown in Figs. 10 and 11, that quantity was found to decrease with orientation rather strongly, reflecting the lower extensibility of oriented materials [15]. Thus, when values of the ratio α are calculated by means of Equation 4 from the measured quantities dG_c/dt and U_b , they are found to increase sharply with orientation (Fig. 12), indicating much larger plastic zone sizes for highly oriented samples, with effective diameters of 3 to 4 times the tear path width. Conversely, for tearing parallel to the orientation direction the effective diameter of the plastic zone is inferred to be quite small at high

TABLE II Tear energy, G_c , work-to-break, U_b , and ratio, α , of effective diameter of plastic zone to width, t , of tear path for oriented TPI

Extension ratio, λ , across tear direction	$G_{c,0}$ (kJ m ⁻²)	dG_c/dt (MJ m ⁻³)	U_b (MJ m ⁻³)	α
0.50	1.3 ± 0.3	9.2 ± 1	70	0.13
0.59	1.8 ± 0.3	12.0 ± 1	60	0.20
0.71	2.5 ± 0.5	17.0 ± 1	51.5	0.33
1.00	4.0 ± 1.0	27.5 ± 2	43	0.64
1.50	6.0 ± 1.0	40.0 ± 2	39	1.02
2.08	7.8 ± 2.0	53.0 ± 9	34	1.55
2.50	10.0 ± 2.0	63.0 ± 7	31	2.05
3.00	12.5 ± 3.0	73.0 ± 10	29	2.50
4.00	16.0 ± 4.0	90.0 ± 15	26.5	3.40

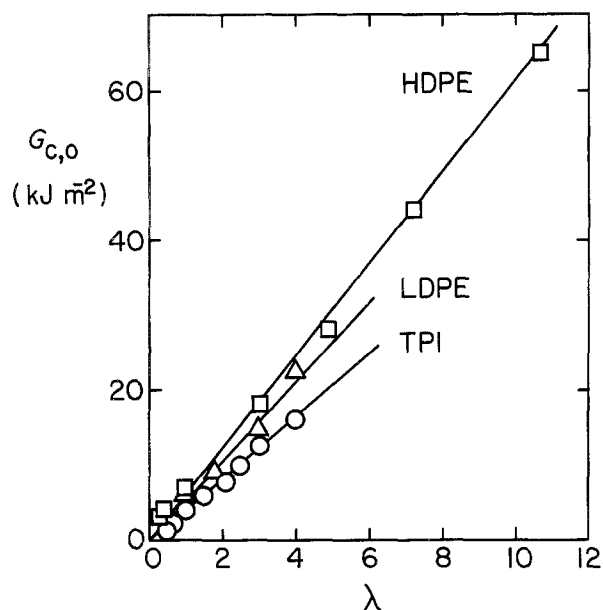


Figure 9 Threshold fracture energy $G_{c,0}$ plotted against the degree of orientation (extension ratio λ) across the tear direction.

orientations, only of the order of 10 to 20% of the tear path width.

Some direct observations of the extent of plastic yielding around the tear tip are now described.

3.5. Plastic zone sizes in edge-cut tensile specimens

Up to this point the extent of plastic yielding around a propagating tear has been a rather hypothetical quantity. Although values have been deduced for the effective diameter of the plastic zone from the marked dependence of the fracture energy upon the width of the tear path, no direct observation of the yielded zone has been made. Indeed, it is difficult to observe plastic zones during tearing. An attempt was therefore made to model the tear tip, using tensile specimens with cuts made in both edges, as shown in Fig. 13. These specimens had an overall width of 9 mm and the edge cuts

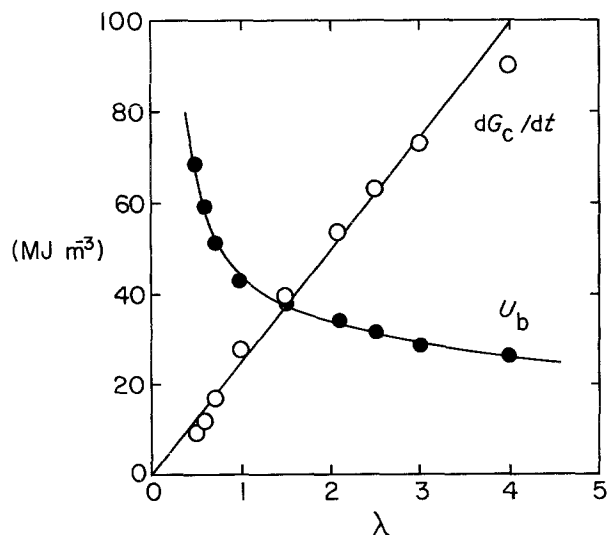


Figure 10 Values of dG_c/dt for TPI, taken from Figs. 5 and 6, and corresponding values of the tensile work-to-break U_b , plotted against the degree of orientation (extension ratio λ) across the tear direction. Note that in determining U_b stretching took place in the same direction as the prior extension λ .

TABLE III Tear energy G_c , work-to-break U_b , and ratio, α , of effective diameter of plastic zone to width, t , of tear path for oriented HDPE

Extension ratio, λ , across tear	G_{c0} (kJ m^{-2})	dG_c/dt (MJ m^{-3})	U_b (MJ m^{-3})	α
0.32	3 ± 1.5	60 ± 6	~ 300	0.20
0.43	4 ± 1.5	74 ± 6	~ 250	0.30
1.00	7 ± 2	94 ± 8	180	0.52
3.05	18 ± 2	120 ± 10	130	0.92
4.90	28 ± 5	165 ± 16	105	1.57
7.20	44 ± 6	200 ± 30	84	2.38
10.70	65 ± 10	255 ± 50	62	4.10

were 2.25 mm deep, so that one-half of the original cross-section remained to be ruptured as in the scored tear testpieces.

Measurements were made of the length, l' , of the plastically yielded region between the two cuts as the specimen was slowly stretched at a strain rate of about 0.1 sec^{-1} . Maximum values were recorded just before rupture and were used to calculate effective values of the ratio $\alpha' = l'/t'$, where t' is the width of the specimen between the two edge cuts, Fig. 13.

The quantity α' represents the effective size of the plastic zone. It can be compared with corresponding values of the parameter α , defined in Equation 4. This comparison is made in Fig. 14. As can be seen, there is reasonably good agreement between the two measures of the size of the plastic zone, suggesting that it is, indeed, plastic work which accounts for the strong dependence of the fracture energy upon the sheet thickness, and that the larger effect shown by highly oriented materials is due to larger plastic zones for a given thickness.

It should be noted that simple plastic yielding was only observed at intermediate levels of orientation, with λ lying between about 0.5 and 8. Highly oriented specimens showed stress-whitening rather than plastic flow (Fig. 15). The length of the whitened zone was taken as the appropriate measure of the softened zone size in these cases. Specimens that had been highly oriented in the transverse direction, i.e. in the direction of the initial cuts, so that the extension ratio, λ , across the tear plane and in the tension direction was small, much less than unity, showed no plastic yielding.

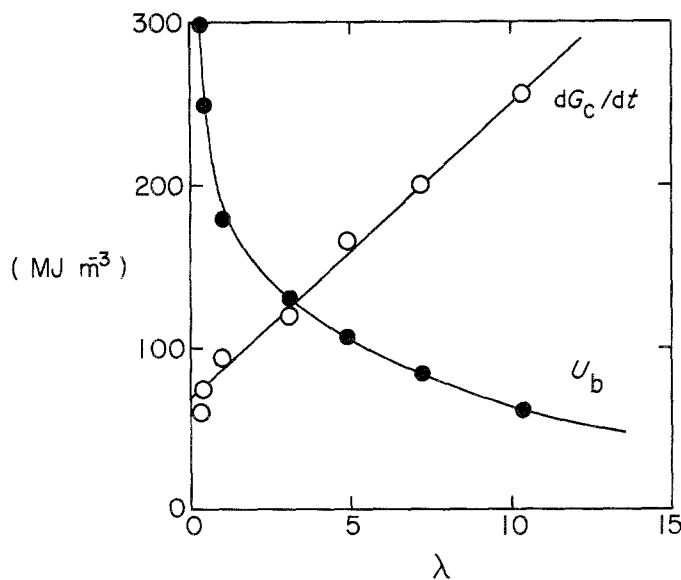


Figure 11 Values of dG/dt for HDPE, taken from Figs. 7 and 8, and corresponding values of the tensile work-to-break, U_b , plotted against the degree of orientation (extension ratio λ) across the tear direction. Note that in determining U_b stretching took place in the same direction as the prior extension λ .

Instead, a thin band of material between the initial cuts broke apart into fine fibrils on stretching (Fig. 15). In these cases, the width of the fibrillated band was taken as a measure of the length, l' , of the softened zone. The value of α' deduced in this way was rather small, about 0.2.

Measurements on edge-cut samples of TPI gave values of α' of about 0.6 in the unoriented state, in good agreement with the value deduced for α of 0.65 from measurements of tear strength (Fig. 12).

3.6. Strength of highly oriented (fibrous) materials

The observations made here suggest that highly oriented, partially crystalline materials have a high resistance to tear propagation across the direction of orientation for two main reasons. When the tear width is extremely small, so that plastic yielding is localized in the torn surfaces, then the threshold tear strength is applicable. This parameter increases in direct proportion to the extent of prior orientation (Fig. 9), in the same way that the density of molecular chains crossing the fracture plane would do for an affine deformation. Although the threshold strength is far greater than the work of rupture of molecular chains, by a factor of about $\times 100$, it appears to change in a similar way with prior orientation.

When the tear path width is larger, then an additional contribution to the fracture energy arises from plastic yielding in a zone around the tear tip. In oriented materials, this zone extends further into the material in the direction of prior orientation and the contribution from plastic yielding is then greatly enhanced (Figs. 14 and 15).

The reason for this increase in size of the yielded zone is obscure. It does not reflect a change in the yield stress, which is hardly altered by a prior extension [15]. On the other hand, the maximum degree of plastic deformation is much reduced by a prior orientation, and it may be that this feature is mainly responsible for the increased amount of material that becomes softened.

For a tear propagating in the direction of prior orientation, the energy required is correspondingly

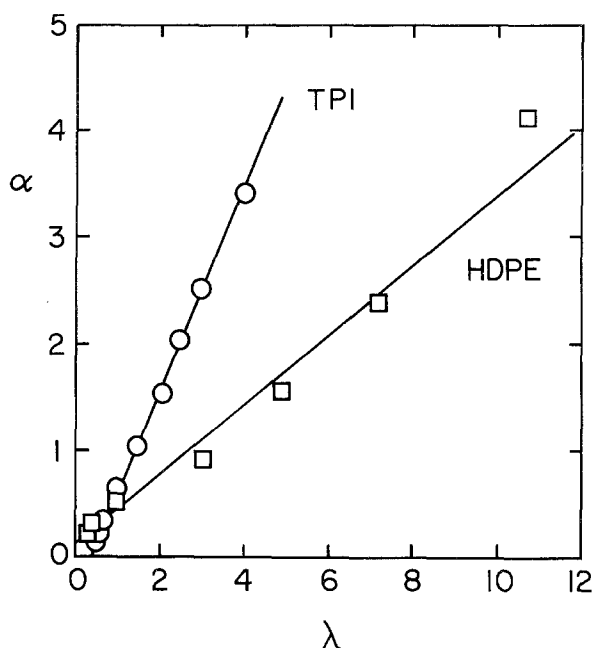


Figure 12 Values of the ratio, α , of effective diameter of the plastic zone to width, t , of the tear path plotted against the degree of orientation (extension ratio λ) across the tear direction.

reduced, for the same reasons. When the degree of orientation is sufficiently high, however, the specimens no longer exhibit plastic yielding but break open by fibrillation. Under these circumstances, the fracture energy is not enhanced by plastic work, whatever the width of the tear path.

4. Conclusions

The following conclusions are obtained:

1. The tear strength (fracture energy) for three partially crystalline materials increased linearly with increasing thickness of the test sheet. This is attributed to an increasing volume of material that becomes deformed plastically during tear propagation. Thus, the three polymers all tore in the same way, with a great deal of energy expended in plastic work.

2. By extrapolating to zero thickness of the sheet, threshold values were deduced for the fracture energy in the absence of gross plastic yielding. The values obtained were still relatively large, however, 4 to 7 kJ m⁻². They were surprisingly similar for the three materials (TPI, LDPE and HDPE), having quite different tensile properties. Expressed in terms of the inherent sharpness of the tear tip, they correspond to effective diameters of 35 to 90 μ m, about five times the observed spherulite diameters in all cases.

3. The threshold fracture energy varied strongly with the degree of orientation of the material at the time of crystallization, changing from 1 to 60 kJ m⁻² as the extension ratio λ across the tear direction was increased from less than one-half to ten times. This dependence upon λ was found to be approximately linear, passing through the origin. It is consistent with a simple geometrical change in the number of molecular chains crossing the tear plane when an affine deformation is imposed, although the measured fracture energies are much greater than chain rupture energies.

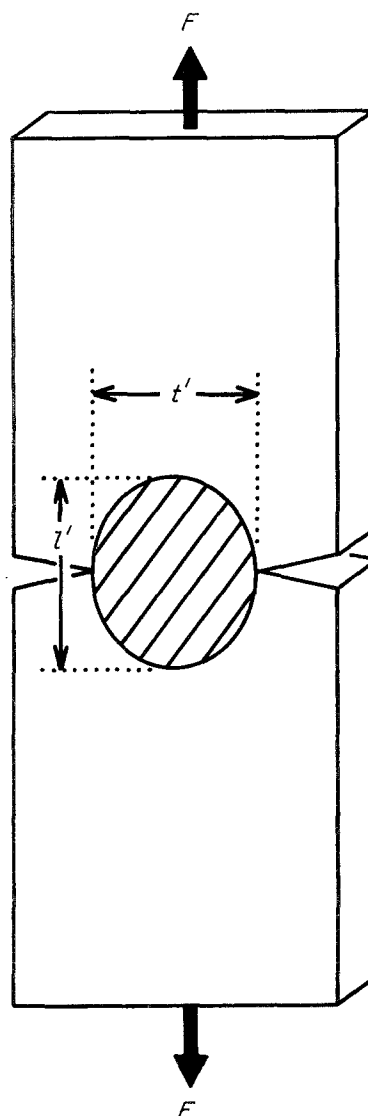


Figure 13 Sketch of plastic zone for edge-cut tensile specimen.

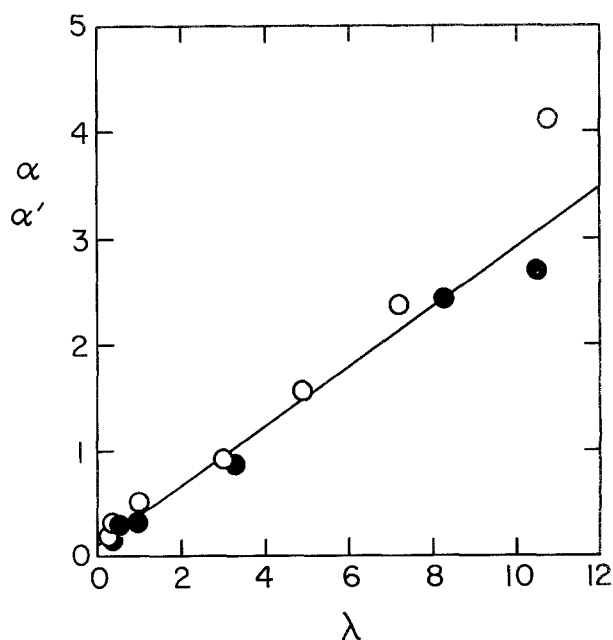


Figure 14 Values of the ratio, α' , of effective length, l' , to width, t' , of the plastic zone in edge-cut tensile specimens of HDPE, (Fig. 13) plotted against the degree of prior orientation (extension ratio λ) across the tear direction. The open circles represent values of α deduced from the dependence of tear energy, G_c , upon tear path width, t , taken from Fig. 12.

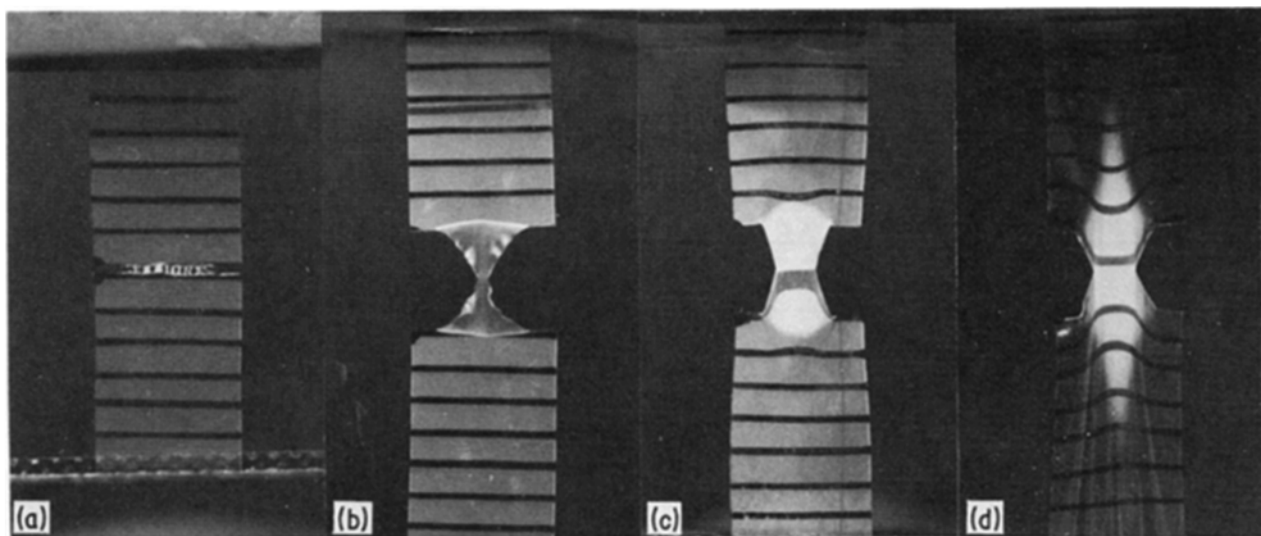


Figure 15 Edge-cut specimens of HDPE under tension. (a) Fibrillation, $\lambda = 0.34$. (b) Plastic yielding, $\lambda = 1.0$. (c) Plastic yielding, $\lambda = 4.6$. (d) Stress-whitening, $\lambda = 10.7$.

4. The dependence of fracture energy upon sheet thickness also increased greatly with increasing orientation of the material across the tear direction. This is attributed to an increasing length of the plastically yielded zone (or stress-whitened zone at high orientations) in the direction of orientation. Thus, contributions to the observed fracture energy from plastic yielding become much greater at higher orientations, approaching the fibrous state. Model experiments with edge-cut tensile specimens showed a corresponding increase in the length of the highly deformed zone with prior orientation of the material in the tensile direction.

It should be noted that all of the experiments described here were carried out with sheets scored initially about half-way through the thickness so that only about one-half of the thickness remained to be torn through (Fig. 2). Plastic yielding then occurred in the highly stressed region near the tear tip but it did not occur far away, where the stresses were lower. If the sheets were not scored initially then the stresses imposed were more homogeneous and large amounts of the material deformed plastically before rupture. Thus, in the absence of stress-concentrating cuts or grooves the present materials do not undergo simple tearing but exhibit the characteristic high ductility and strength of partially crystalline polymers.

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