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The adiabatic fracture of thermoplastic fibres

R. N. HAWARD*
Department of Chemistry, UMIST, P.O. Box 88, Manchester M60 1QD, UK

The development of methods for measuring true stresses and strains in thermoplastics and of models for representing the results, makes it possible to predict polymer performance in a number of ways. Recently this method was used to study the stability of the tensile deformation of high-density polyethylene under adiabatic conditions. It was proposed that at high strain rates, thermomechanical softening would render the plastic deformation process unstable, promoting localised deformation and fracture. In this paper, the isothermal extension process measured at different temperatures is assumed to be stopped and then restarted after different draw ratios have been attained, as in the drawing of a fibre. In this way the effect of draw ratio on fibre tensile properties can then be predicted. It is shown that, with fast deformation under adiabatic conditions, the softening effect due to the increase in temperature exceeds the opposing influence of strain hardening so that the nominal stress is predicted to fall continuously with increased strain. This leads to a ductile fracture process, which, in a fibre, can generate mushroom shaped blobs of polymer at the broken ends. This effect has previously been reported by Hearle and co-workers.
The applicability of the model to different type of fibre is considered. © *2003 Kluwer Academic Publishers*

1. Introduction

The influence of thermomechanical heating on processes leading to fracture in thermoplastics has been recognised for some time. For example Vincent [1] investigated the tensile properties of PVC and concluded that when tensile deformation took place under adiabatic conditions, the nominal tensile stress would fall continuously as the extension ratio increased, leading to fracture. This process was later demonstrated by Cross and Haward [2]. Similarly the role of thermomechanical softening in reducing the energy required for crack propagation was demonstrated some time ago by Doll and co-workers [3, 4] and by Williams [5]. More recently Leevers [6], and Leevers, Douglas, Chong and Williams [7] have developed a thermal decohesion model to explain this effect.

Other workers have shown that the temperature increases during fast fracture can be quite large. For example Haward and Brough [8] found evidence of raised temperatures and melting in polystyrene fracture surfaces. With more direct measurements Dickenson *et al.* [9] recorded temperatures of 600 K and above during the fracture of polystyrene and Fuller, Fox and Field [10] observed temperature rises of 400 and 450 K during fast crack propagation with polystyrene and polymethyl methacrylate respectively. Even with slow craze propagation experiments, Karger-Kocsis and Moskala [11] observed increases of 3–5 K.

With thermoplastic fibres high strain rates have also been shown by Hearle *et al.* [12] to have very substantial effects. At slow rates a V notch grows in a stable ductile process from the fibre surface and this continues until a catastrophic fracture is initiated [13]. The slow growth stage turns out to be similar to that seen by Cornes and Haward [14] and by Walker, Hay and Haward [15] with test pieces taken from ductile thermoplastic sheets which show substantial post yield drawing and orientation. As the extension proceeds these materials developed either V notches at the edge of the test piece or characteristic "diamond" cavities elsewhere. Both features showed stable growth until a point was reached where fast fracture intervened. With fibres (which have already been oriented by a drawing process), a pendulum may be used to generate a fast fracture. Under these conditions, the fibre behaves in quite a different way and after fracture, the broken ends left behind swell and assume a mushroom shape. Hearle and coworkers [12] concluded that localised adiabatic deformation had taken place (Fig. 1), and that after separation, the terminal blobs of heated polymer expanded in response to entropic forces released within the oriented fibre.

2. The process of fast fracture

2.1. The effect of temperature on true stress strain curves

While the explanation outlined above is qualitatively convincing it would be desirable to develop the discussion in a more quantitative way. Similar mushroom ends have not been recorded in conventional experiments with bulk polymers and it would be interesting to know

*Correspondence to "East Barn", Old Ditch, Westbury-sub- Mendip, Somerset BA5 1HN, UK.

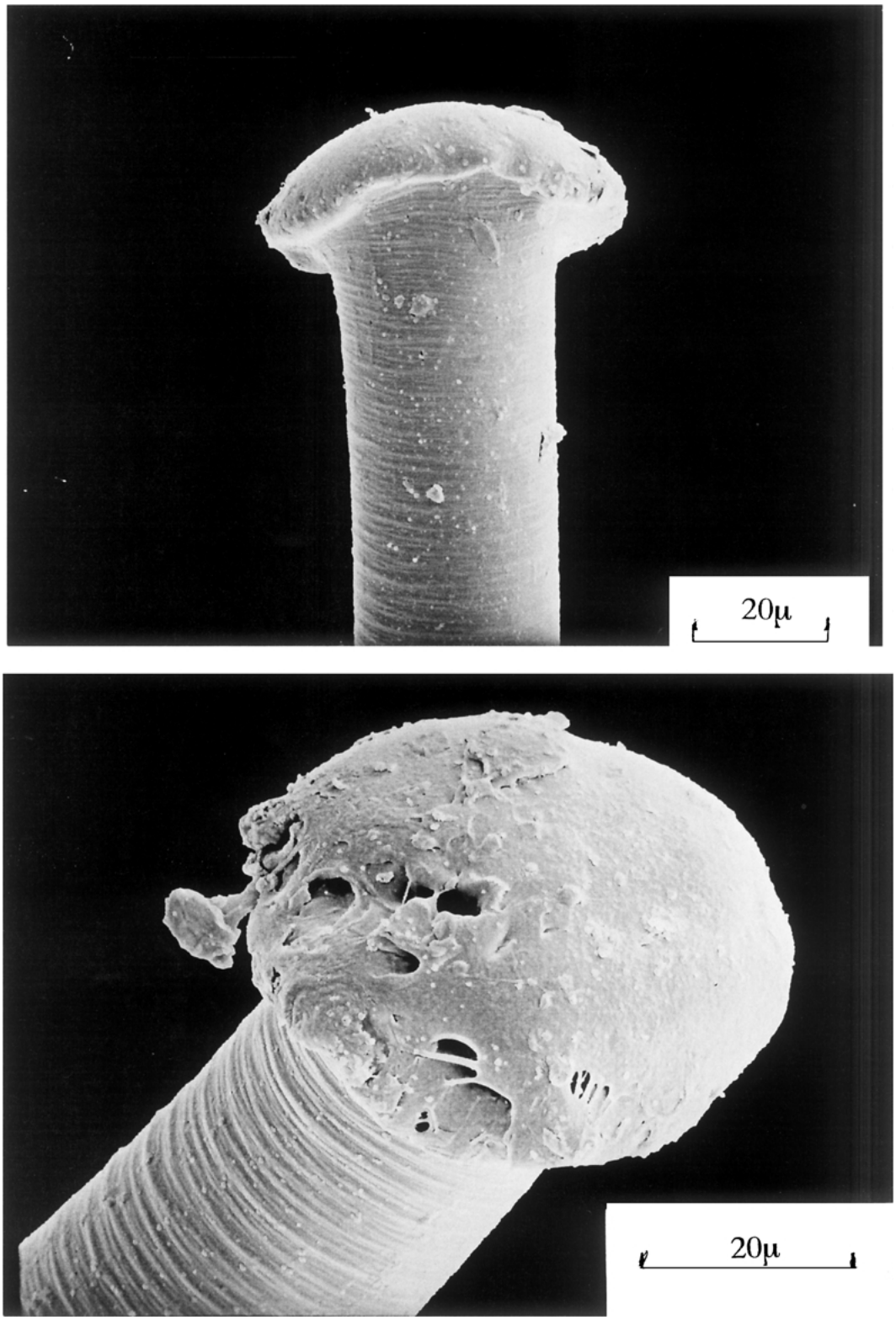

Figure 1 Mushroom ends from a Polyethylene terephthalate (PET) fibre broken at high strain rate (Hearle *et al*. [12]). As the heated material expands to form the mushroom some cavities may develop which can be bridged by fibrils extended in the plane perpendicular to the orientation of the fibre. (Lower picture). (Published by permission of Prof. J. W. S. Hearle).

how the special features present in the fibre affect the fracture process. For this purpose it is useful to take advantage of developments in the measurement of true stress strain curves initiated by G'Sell and Jonas [16, 17] and by Hope, Ward and Gibson [18]. Both groups used a waisted (hour glass) test piece and measured deformation only at the point of lowest cross section. G'Sell and Jonas also controlled the strain rate at this point through a feed back system and so were able to measure true stress and strain under a condition of constant true strain rate. This made it possible to provide an improved quantitative treatment of the plastic deformation process. For this purpose a mathematical model has to be employed to describe the large strain process. This may be either purely empirical [16, 17], or, the semi-empirical HT model proposed by Haward and

Thackray [19] can be used. The latter proposes that strain hardening during the deformation process is due to changes in the conformation and entropy of the polymer chain similar to those described in the theories of rubber elasticity [20]. With either model it is possible to give a quantitative account of the necking process in a conventional isothermal tensile test (see Boyce and co-workers [21, 22], G'Sell *et al.* [23] and others [24, 25]).

In view of the success of this work it was decided to investigate the use of published true stress strain curves to predict adiabatic effects. For this purpose it is necessary to employ experimental measurements covering a wide range of temperatures at a particular strain rate. From a small number of available results those published by Hiss and Strobl [26] with a high density polyethylene covering the range of 297 to 401 K (24–128°C) at constant strain rate (in this case 10^{-2} sec^{-1}), were particularly suitable. The results from these workers, who employed techniques similar to G'Sell and Jonas, also had the advantage of complying with the HT model in its simplest form. This form of the model predicts that after yield the true stresses in a section of material extending at a constant true strain rate, may be represented by the equation

$$\sigma_t = Y + Gp(\lambda^2 - 1/\lambda) \quad (1)$$

where σ_t is the true stress, Gp a strain hardening constant, Y the yield stress and λ the extension ratio. In developing an adiabatic relation, which includes temperature changes [27], a combined relation has been derived to describe the results over a range of temperatures. For this purpose it was necessary to take into account the temperature variation of Y and Gp and the following equation was derived to represent the results over the temperature range 297–361 K [27].

$$\sigma_t = (19300/T - 40.6) + \exp(3414/T - 9.9) \times (\lambda^2 - 1/\lambda) \quad (2)$$

where T is the absolute temperature. When required $\sigma_{n,t}$ the true nominal stress is estimated as σ_t/λ. These equations, though derived from a physical model may equally be regarded as simply an empirical representation.

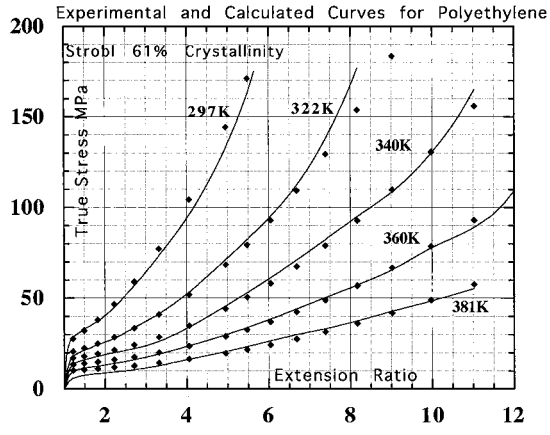

Figure 2 The true stress strain curves measured by Hiss and Strobl [26] are reproduced together with points calculated from Equation 2.

In Fig. 2 the results calculated from this equation, represented by solid points, are compared with the experimental curves represented as continuous lines over the temperature range 24–108°C. The agreement of the model with experiment is considered very satisfactory. It is the object of the work described below to show that with oriented polymer the amount of adiabatic heating in a tensile experiment should increase with the initial draw ratio of the sample.

3. The adiabatic process

3.1. Stress changes during deformation

In reference 1 Vincent proposed that if a test piece (or in this case a fibre) was extended under tension, its behaviour depended qualitatively on the changes in nominal stress with extension ratio. For example if, at the beginning of the large strain process (yield) $d\sigma_n/d\lambda < 0$ i.e., negative then necking was predicted (Case I: Considere condition).

For necking to be stable, with more material being drawn into the necking process from the matrix, then $d\sigma_{n,t}/d\lambda$ must increase to zero and become positive (Case II: 2nd Considere condition).

On the other hand if $d\sigma_{n,t}/d\lambda < 0$ and remained negative at high values of λ fracture is predicted (Case III). However, where the displacement due to the plastic deformation is large compared with the elastic deformation of the test piece the fracture process may be relatively stable if a constant rate of extension is applied in the conventional procedure. This type of behaviour occurs with the ductile fracture of plastics [14, 15] but can also be observed with pure metals as described by Hull [28].

In a previous publication [29] it was shown that nearly all polyethylenes following the Gaussian Equation 1 above, would also meet the 2nd Considere condition (Case II). Later, when adiabatic conditions were introduced into an isotropic high density polyethylene having the true stress strain properties reported by Hiss and Strobl [26] then Case III behaviour was predicted i.e., strain localisation and failure. Of course these estimates only apply directly to the constant strain rate of 10^{-2} sec^{-1} used in [26], whereas adiabatic conditions would only be attained at much higher strain rates. However it is clear that at higher strain rates stresses would be higher and adiabatic effects correspondingly increased. The prediction of failure would therefore be reinforced as argued in [27]. It is of interest also to note that thermomechanical heating has been reported to cause strain localisation also with metals [30].

At this stage it is useful to present information on isothermal changes in stress as represented by $d\sigma_{n,t}/d\lambda$. For this purpose it is necessary to employ the quantities Y and Gp as derived from Equation 1 and this is shown in Fig. 3a using the isothermal curve at 297 K from [26]. The relation for $\sigma_{n,t}$ then becomes:

$$\sigma_{n,t} = Y/\lambda + Gp(\lambda - 1/\lambda^2) \quad (3)$$

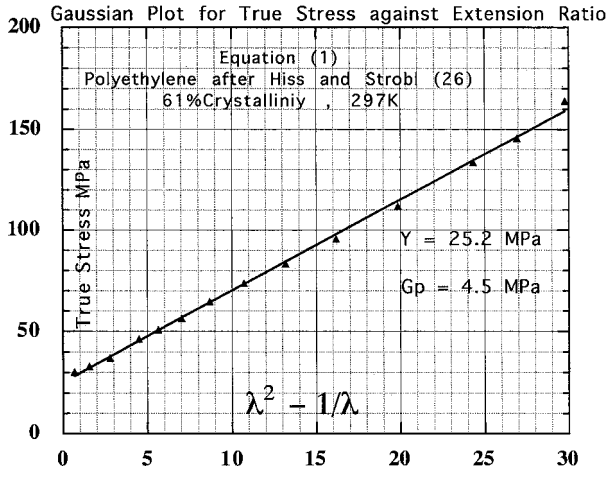

Figure 3 (a) Experimental points derived from Hiss and Strobl's true stress strain curve [26] at 297 K are plotted according to Equation 1.

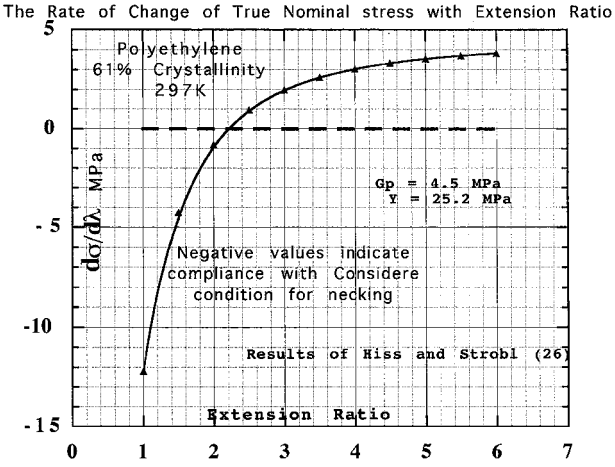

Figure 3 (b) The differential variation of $\sigma_{n,t}$ with extension ratio under isothermal conditions. Below $\lambda = 2.2$ $d\sigma_n/d\lambda$ is negative but it becomes positive when $\lambda > 2.2$. This correlates with the normal necking behaviour of HD polyethylenes.

and

$$d\sigma_{n,t}/d\lambda = Gp + 2Gp/\lambda^3 - Y/\lambda^2 \qquad (4)$$

Equation 4 is then plotted in Fig. 3b. It will be seen that the differential (slope) is initially negative (the material necks) but later rises through zero at an extension ratio around 2.2. It then increases towards the constant value given by Gp at high values of λ. This means that material which has already been extended beyond an extension ratio of 2 does not neck when extended further. Thus polymers that are capable of appreciable extension after passing through the neck will extend uniformly. This makes it possible to measure the strain hardening constant Gp in a limited number of cases with a conventional tensile test using a prenecked material e.g., low density polyethylene [31] or polyaryl ether ether ketone [32]. Extrapolation of this conclusion would suggest that oriented fibres when further extended should do so uniformly, which is, of course, quite different from what happens with most isotropic materials. However, this behaviour is not easy to measure experimentally, as a ductile fracture process intervenes which appears to be controlled by shear strains [12, 14, 15]. In the circumstances it seemed appropriate to estimate the effect of draw ratio on the adiabatic deformation of the model polyethylene which had been previously extended over the range of experimentally accessible draw ratios of 2–5.

3.2. The adiabatic deformation of predrawn polyethylene

In their true stress strain experiments with polyethylene G'Sell and Jonas [17] showed that if the extension process was discontinued and the applied stress allowed to return to zero, the measured stress returned to the original value if the extension was subsequently restarted, in the same way as before. This type of behaviour, which is confirmed by many other measurements, makes it realistic to treat the properties of predrawn polymer as a simple continuation of the original stress strain relation (see Fig. 4). In this way it is possible to adapt the previous methods [27] to the deformation and fracture of drawn polyethylene. The initial isotropic segment of polymer ($\lambda = 1$) can be drawn to any convenient extension ratio λ_o and the process restarted at this draw ratio. We then apply a constant true nominal stress $\sigma_t/\lambda_o = \sigma_{n,t,o}$ leading to an initial strain rate of 10^{-2} sec^{-1}. Under isothermal conditions when $\lambda > 2.2$ the rate of strain at constant load would fall ($d\sigma_{n,t,o}/d\lambda > 0$ Fig. 3) though at values of $\lambda < 2.2$ it would increase. This corresponds to normal necking behaviour as previously reported by Coates and Ward [33] where under a constant load the local rate of strain initially increases and then falls.

In our case the work done per unit volume of material startimg the extension at λ_o will be $\sigma_{n,o}(\lambda - \lambda_o)$ Jcm^{-3} (MJ m^{-3}) leading to a temperature rise of $\sigma_{n,t,o}(\lambda - \lambda_o)/\rho \cdot Cp$, where ρ is the density and Cp the specific heat. This temperature rise can then be estimated as in [27] allowing for the change in Cp with temperature. These estimates are given in Fig. 5 for values of λ_o from 2–5. It will be seen that the rate of increase in temperature with extension ratio rises in line with the starting value of the draw ratio λ_o.

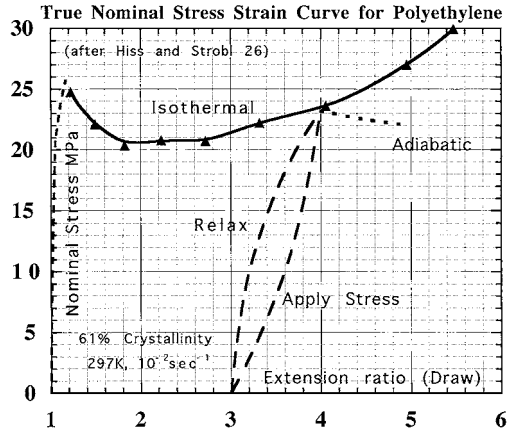

Figure 4 A isothermal plot of σ_t/λ against λ is relaxed and restarted at $\lambda = 4$ and the resulting stresses are demonstrated diagrammatically (dotted lines). Under isothermal conditions the stress continues along the original curve when the extension is restarted.

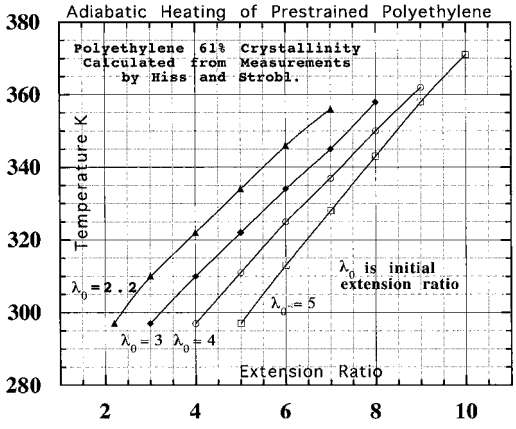

Figure 5 Temperature changes are estimated for a theoretical adiabatic extension under a constant nominal stress characteristic of the draw ratio on the isothermal curve.

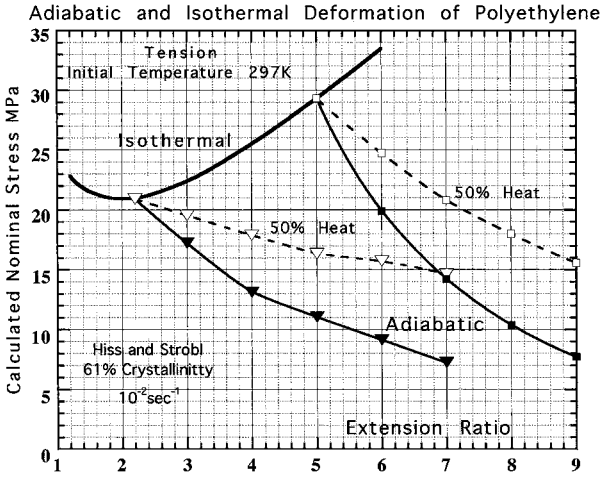

Figure 6 Under adiabatic conditions the rise in temperature causes the nominal stress to fall continuously as λ increases. At the higher draw ratio the effect becomes greater. The dotted lines are for a 50% conversion of the work done into measurable heat.

Using the temperature and extension ratios from Fig. 5 it becomes possible to calculate the stresses under adiabatic conditions at a strain rate of 10^{-2} sec^{-1} according to Equation 2. The predicted results are presented in Fig. 6 where the estimated adiabatic stresses fall continuously below the initially applied values of $\sigma_{n,t,o}$, and the magnitude of the fall is greatest at the initial highest draw ratio λ_o. This implies that if the stress were maintained as $\sigma_{n,t,o}$ as in the model, there would have to be very large and continuous increases in strain rate which would maintain adiabatic conditions and lead on to fracture in accordance with case III. This occurs, in spite of the stability of isothermal plastic deformation where $d\sigma_{n,t,o}/d\lambda > 0$ (in this case $\lambda > 2.2$). So the applied nominal stress increases continuously with draw ratio and along with the scaling factor which determines the generation of heat—$\sigma_{n,t,o}/\rho \cdot Cp$. This causes the difference between isothermal and adiabatic deformation to increase with the draw ratio λ_o. Although this calculation has been entirely based on one sample of polyethylene, it is expected that the effect of draw ratio on yield stress and therefore on adiabatic sensitivity should apply generally to other oriented plastics and fibres [32].

TABLE I Estimated scaling factors for HDPE , Nylon 66 and PET [33 and 34]

Property	HDPE	Nylon 66	Polyethylene terephthalate (PET)
Medium tenacity σ_t draw ratio 4.3	105 MPa	580 MPa	490 MPa (Draw ratios of PET. not known)
High tenacity σ_t draw ratio 5.0	135"	750"	800"
Density $g \cdot cm^{-3}$	0.95	1.14	1.38
Specific heat *Cp* J/g	2.0	2.1 (ave value)	1.13 (293 K)
Melting point K	403	523	533
Medium tenacity scaling factor K	49	240	310
High tenacity scaling factor K	71	310	510

4. Application to thermoplastic fibres

It has been shown above that the prior drawing of high density polyethylene should increase the adiabatic heating effect when it is suddenly extended later. Because this increase in temperature is mainly controlled by the increase in yield stress[1] and its consequent effect on the scaling factor, it is possible to predict the occurrence of adiabatic fracture in other types of film or fibre by estimating the scaling factor. In this way, using published properties for Nylon 66 polyamide [33] and polyethylene terephthalate (PET) [34] which have been shown to exhibit mushroom fractures as in Fig. 1, it is possible to equate tensile strength (tenacity) measurements with true yield stress (σ_t) and to estimate approximate values of the scaling factors for adiabatic deformation (Table I).

The scaling factor gives the rise in temperature K for unit increase in extension ratio for terephthalate may have been slightly increased by the use of a low temperature specific heat instead of an average value between 293 and 533 K. The high scaling factor is also on line with Russian studies in which a temperature rise of 170 K was observed during the fracture of PET with a draw ratio of 4 [37].

It is suggested that these estimates show that adiabatic heating effects in Nylon 66 and Polyethylene terephthalate fibre should be larger than with the polyethylene, and sufficient to soften the polymer significantly in spite of their higher melting points. It is concluded that under fast fracture conditions an adiabatic type of fracture may be expected, as observed by Hearle *et al.* [12].

It should be noted that true stress strain curves, both for Nylons 6 and 66 and for two different polyethylenes, have been measured by G'Sell and Jonas [17] at ambient temperatures and found to follow Equation 1.

[1] In deforming a fibre the draw ratio based on the isotropic material may not be available so that the extension ratio has to be based on the drawn length and the measured stress is the true stress. It should be noted that at a draw ratio λ_o the product $\lambda_{t,o}$ (Extension ratio based on extended length) = $\sigma_{n,o}$ (extension ratio based on original length).

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