The effect of temperature on the fracture of rubber modified polystyrene

M. PARVIN
Mechanical Engineering Department, Aria-Mehr University, Teheran, Iran
J. G. WILLIAMS
Mechanical Engineering Department, Imperial College, London, UK

Brittle fractures were obtained in rubber modified polystyrene over the temperature range -120 to 20°C using the surface notch method. When compared with single edge notch data, a thickness effect was apparent and this was described in terms of plane stress and plane strain fracture toughness values. The plane strain value agreed closely with that of the polystyrene matrix indicating that the constrained region showed no toughening effect. The relaxation process of the rubber was apparent in the plane stress toughness.

1. Introduction
Polystyrene is one of the most important commercial polymers and in its rubber toughened forms (high impact polystyrene, HIPS, and acrylobutadienestyrene, ABS) is used widely in applications requiring substantial levels of toughness. The toughening process has been widely discussed (e.g. [1]) and may be described briefly as the rubber particles acting as initiation points for the production of energy absorbing crazes resulting in a considerable increase in the fracture toughness over the basic polymer. The viscoelastic properties of the rubber phase are considered to be a contributing factor in the toughening process [2]. This paper seeks to offer some clarification of these mechanisms by describing the toughness in fracture mechanics terms and using concepts developed with other polymers to elucidate the energy absorbing processes.

Previous work on both PMMA and polystyrene [3, 4] has shown that the presence of a viscoelastic relaxation process around room temperature results in stable slow crack growth which elevates the fracture toughness of the material. The \( \beta \) process at 20°C is quite strong in PMMA leading to a substantial effect but in polystyrene it is less pronounced. By testing over a wide temperature range, the influence of the \( \beta \) transition in PMMA was further illustrated [5] so that when the instability fracture toughness, \( K_C \), was plotted as a function of temperature, a curve of \( K_C \) increasing with decreasing temperature was produced over the \( \beta \) transition range. Similar work on polycarbonate showed the same effect [6] but here there was evidence of a pronounced thickness effect which was absent in both PMMA and polystyrene. By using plane stress and plane strain fracture toughness values (\( K_{C2} \) and \( K_{C1} \)), it was possible to describe the data in the same terms as PMMA. Some fracture mechanics studies have been made on toughened polystyrene at 20°C [7] and because of the large degree of crazing, the specimens used (50 mm wide) underwent gross deformation and it was necessary to invoke the ductile failure criterion \( J_C \) to describe the data. Attempts to produce brittle failures for which \( K_C \) is an appropriate description were unsuccessful.

In the study reported here polystyrene and a rubber toughened grade were fracture tested over the temperature range -120 to 20°C. Thickness effects were apparent in the toughened material of a similar nature to polycarbonate and by the use of the surface notch technique it was possible to produce brittle failure over the whole temperature range.

2. Some theoretical concepts
The fracture toughness used in this work is calculated from the usual result:

\[
K_C^2 = Y^2 \sigma^2 a 
\]
where $Y^2$ is a geometric correction parameter depending upon the specimen geometry. For a centrally notched infinite plate it is $\pi$ and for a wide single edge notched (SEN) plate, 4. Some of the data reported here were obtained using SEN specimens and the appropriate function for $Y^2$ used (see [3–6]). Tests were also performed using the surface notch (SN) geometry which is shown in Fig. 1. A dead sharp flycutter is used to produce a notch in the form of a circular arc in the surface of the specimen. The expression for $K_C$ must be corrected for the different crack front length to crack area ratio as compared with the usual through-thickness crack and this is done using:

$$K'_C = Y^2 \sigma^2 a$$  \hspace{1cm} (2)

where

$$Y^2 = \frac{1.21\pi}{\sigma - 0.212(\sigma/\sigma_y)^2}$$

$$\phi = \frac{m^2}{b} \left\{ 1 - \left( \frac{b^2 - a^2}{b^2} \right) \sin^2 \theta \right\}^{3/2} d\theta$$

and $\sigma_y$ is the tensile yield stress [9].

It will also be assumed here (see [6]) that the surface notch has an equivalent thickness $H'$ given by a crack of uniform depth $d$ and the same surface area. For the circular form with $b \ll R$ (the radius of curvature):

$$H' \approx \frac{b}{2}$$  \hspace{1cm} (3)

It should be noted that Equation 2 is deduced assuming a semi-elliptical form for which the result is $H' = \pi/2b$.

It has been found helpful previously [6] to assume that there are two fracture toughnesses for a polymer; one under plane stress conditions, $K_{C2}$, and one under plane strain, $K_{C1}$. The concept is well known in the metals field but here it is invoked to describe flat fracture or at least those with very small shear lips. The equating of the plane stress surface zone with a shear lip usually found in the metals literature appears to be unnecessary in the case of polymers, and indeed, shear lips are totally absent in polystyrene. It seems most likely that plane stress may be associated with bulk yielding while the constrained plane strain state results in localized yielding (shear bands) or the formation of a craze.

The extent of the plane stress zone from the surface is taken as the plastic zone size:

$$r_{y2} = \frac{1}{2\pi} \frac{K_{C2}^2}{\sigma^2}$$  \hspace{1cm} (4)

and when a specimen of thickness $H$ fractures, the apparent toughness $K_C$ is made up of $K_{C2}$ in the surface zones of width $r_{y2}$ with $K_{C1}$ acting in the central section.

$$HK_C = 2r_{y2}K_{C2} + (H - 2r_{y2})K_{C1}$$

and

$$K_C = K_{C1} + \frac{2r_{y2}}{H}(K_{C2} - K_{C1})$$  \hspace{1cm} (5)

Clearly, $K_C$ decreases with increasing $H$ tending to $K_{C1}$ for large thicknesses and for $H < 2r_{y2}$, $K_C = K_{C2}$. In fact, this condition is not usually reached in polymers undergoing the normal yielding process, since for $H < 4r_{y2}$ the central plane strain region is no longer sustained and ductile fracture occurs with substantial thickness reductions [6]. For toughened polystyrenes, however, the crazing process takes place without any appreciable lateral contraction (Poisson’s ratio $\approx 0$) and so it is possible to achieve $K_C = K_{C2}$ without through-thickness contractions. Although this problem may be avoided, the further one of plastic zone size is difficult with these materials. For the expressions for deriving $K_C$ to be valid, $r_{y2}$ must be less than the crack length and must also not be close enough to the edge of the specimen to cause the onset of gross yielding. Observed plastic zone sizes are typically around 10 mm in small specimens so that for 20 mm cracks in SEN tests the plastic zone is quite near the back edge in a 50 mm sample. This led to the problems of gross yielding mentioned earlier [7]. The use of large specimens seems to overcome this problem but the true zone sizes are about 3 mm indicating that the observed 10 mm sizes are probably augmented by net section stress effects. To obtain reliable values it seems easiest to test large thicknesses (e.g. $\geq 10$ mm) but these are difficult to produce. The effective thickness concept of the surface notch overcomes this successfully and is the reason for its inclusion here.