JOURNAL OF APPLIED POLYMER SCIENCE VOL. 19, 3355–3359 (1975)

Fracture Properties of Rigid Polyurethane Foams

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Synopsis

The fracture behavior of rigid polyurethane foams has been investigated and is shown to obey the Griffith criterion for fracture in so far as the predicted behavior of tensile strength on the size of artificially introduced cracks is concerned. The energy for crack propagation (fracture surface energy) has been measured as 91.4 J/m^2. From this result, the intrinsic flaw size of the material is calculated. This value was found to be within the range of cell dimensions of the material.

INTRODUCTION

Recent increases in raw material prices have rendered the reduction costs of polymer products of prime importance in the polymer industry. One of the most effective ways of cheapening a product is to produce the material in expanded form. This process is particularly useful in packaging applications in which the improved thermal insulation properties of the expanded material are an additional advantage, especially in the packing of perishable goods. The modulus and tensile strength in foamed polymer systems are, however, lower than in unexpanded materials. A decrease in modulus may well be desirable to enable products to be firmly held in the packaging matrix. However, the decrease in tensile strength of the material may result in brittle fracture in foamed polymer systems such as expanded polystyrene and rigid polyurethane foam.

Fracture properties of brittle polymers have been widely studied using the methods described by Berry.[1,2] Whittaker[3] has investigated the fracture properties of foamed urethane rubbers, but little work appears to have been carried out on the fracture properties of brittle cellular plastics. The present work describes a method for the characterization of the fracture properties of brittle polymeric foams.

THEORY

The present work is based on the energy balance analysis of fracture carried out by Griffith.[4] In plane stress conditions, for a test piece containing an artificially introduced crack of length c, this leads to the following criterion for fracture:

$$\sigma_f = \left(\frac{2ES}{\pi c}\right)^{1/2} \tag{1}$$

where σ_f is the tensile stress at break, E is Young's modulus, and S is the surface energy of the material. This equation has been applied to polymeric materials by Berry[1] who found that although the dependence of tensile stress on crack length, i.e., $\sigma_f \propto (1/c)^{1/2}$ was followed, the value of S determined in these experiments (1.7×10^3 J/m^2 for polystyrene) was greatly in excess of the expected value.

The surface energy which would be expected from even close-packed atomic arrangements would be in the region of 1 J/m^2. This discrepancy may be explained by noting that in polymeric materials a large amount of energy will be dissipated in local plastic deformation of material ahead of the crack tip, whereas the figure of 1 J/m^2 is arrived at by considering chain scission only. It is usual, therefore, to modify the theory by replacing S with a quantity γ, the characteristic energy for crack propagation, where γ includes both S and the plastically dissipated energy, so that eq. (1) becomes

$$\sigma_f = \left(\frac{2E\gamma}{\pi c}\right)^{1/2}. \tag{2}$$

Broutman and McGarry[5] have shown that the value of γ for poly(methyl methacrylate) may be decreased significantly by crosslinking, which reduces plastic flow. Brittle urethane foams are themselves highly crosslinked, so that plastic flow should be relatively unimportant and the Griffith criterion should be followed to yield a value of γ significantly lower than for brittle polymers containing no crosslinks.

EXPERIMENTAL

Polyurethane foams were prepared from a commercial two-component system, Propcan MR49 and Isocon M, supplied by Lankro Chemicals Limited.

In order to prevent contamination with water vapor from the atmosphere, the components were initially split into small batches which were opened only immediately prior to the preparation of the foams. This precaution was taken as it was thought that water was likely to reduce the reactivity of the components. This would in turn reduce the crosslink density of the polyurethane and so affect the fracture surface energy.

To fabricate the foam, equal volumes (20 cm^3) of each component were metered by syringe and thoroughly mixed by hand, prior to the onset of expansion. The mixture was then poured into open trays where expansion took place. After preparation, the foams were conditioned for one week at room temperature before any testing was carried out. This method of fabrication was found to give foams of reproducible physical properties with a mean density of 35 kg/m^3 and mean Young's modulus of 1.67 MN/m^2.

Specimens for tensile testing were cut from the foams using a band saw to give rectilinear test pieces of approximate dimensions 5 × 35 × 150 mm. Evidence of anisotropic structure was shown by the elongation of the cells in the rise direction. The effect of anisotropy on the fracture properties of brittle polymers has been reported by Broutman and McGarry[5] and Curtis.[6] For this reason, care was taken to ensure that crack propagation took place in one direction only, i.e., the rise direction; and so the 35 mm dimension was always cut in this rise direction. Edge cracks of various lengths up to 10 mm were

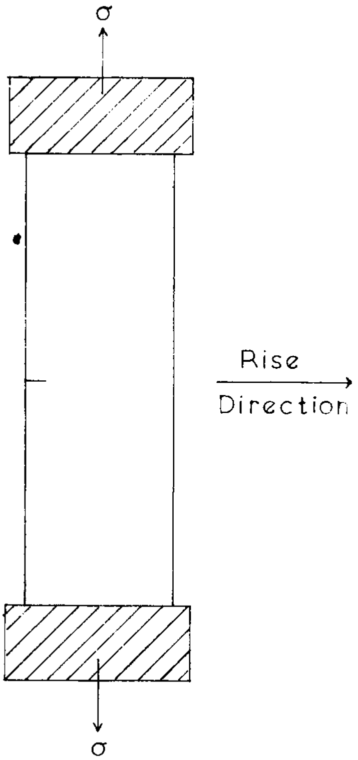

Fig. 1. Edge-crack test piece.

then introduced into the specimens using a scalpel blade. The length of these cracks was measured using a traveling microscope. The form of the test pieces is shown in Figure 1.

The tensile strength of the test pieces was measured using a Monsanto Tensometer Type E testing machine. Sliding wedge grips were used at a cross-head speed of 25 mm/min. The tensile strength of unnotched material was also determined using dumbbell-shaped test pieces with paralled sided region of width 12.5 mm. It was found necessary to use dumbbells rather than parallel sided test pieces because the latter type frequently failed in the grips.

RESULTS

A plot of tensile strength against $(1/c)^{1/2}$ is shown in Figure 2. The graph gives a good straight line, the coefficient of variation of the gradient being 5.5%. By substitution of the value for the modulus of material into eq. (2), the value of γ may be calculated from the gradient of the line. This procedure gave a value of 91.4 J/m^2 for the fracture surface energy.

Results from specimens with edge cracks of length below 1 mm have not been used in the above calculations since Berry[1] in his work on polystyrene found that below a certain crack length, eq. (2) was not satisfied. The tensile strength of specimens with very short artificial flaws was much lower than expected. This behavior is explained by the presence of natural flaws in the

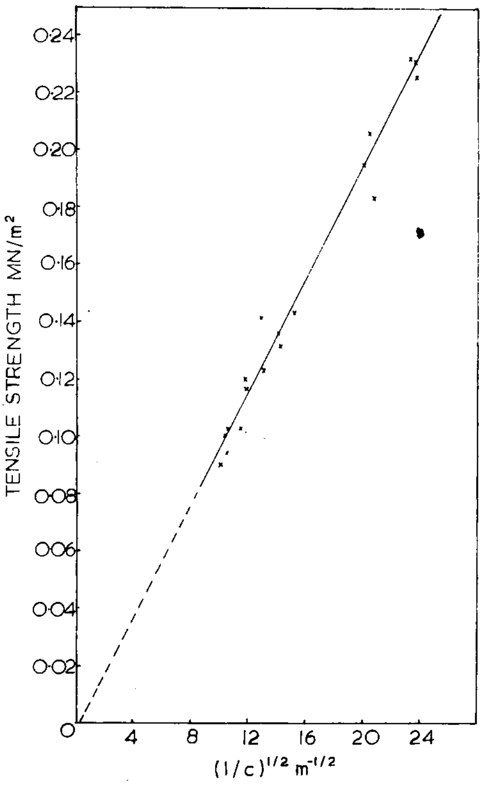

Fig. 2. Tensile strength vs. $(1/c)^{1/2}$.

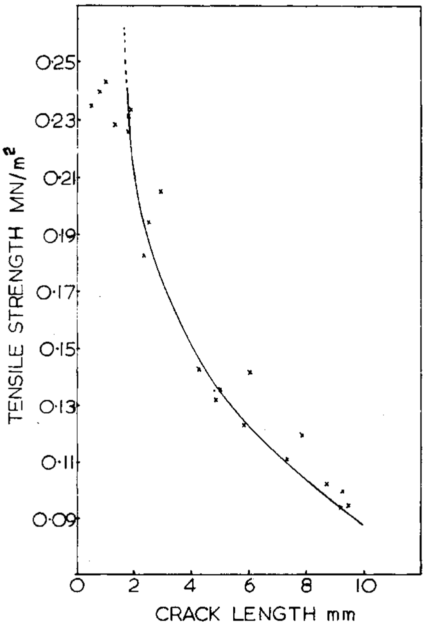

Fig. 3. Tensile strength vs. crack length.

materials which initiate fracture at a lower overall stress level than would be predicted from the length of the artificially introduced flaw.

This effect is illustrated in Figure 3, where tensile strength is plotted against crack length. At crack lengths below about 1 mm, the tensile strength ceases to follow the expected upward trend, indicating that the intrinsic flaw length has been reached. Using the tensile strength of unnotched test pieces together with the value of γ of 91.4 J/m^2, the intrinsic flaw length may be calculated from eq. (2). The intrinsic flaw length was found to be 0.73 mm. In a foamed material, it might be expected that the intrinsic flaw size is governed by the size of broken cells at the edge of the test piece. The mean cell diameter of the foams in the rise direction was found to be 0.47, with a maximum size of 1.3 mm.

CONCLUSIONS

The fracture behavior of rigid polyurethane foams is in close agreement with the Griffith criterion for fracture in so far as the tensile strength is proportional to $(1/c)^{1/2}$. The value for fracture surface energy of 91.4 J/m^2, although being a factor of ten less than for other brittle polymeric materials, is still considerably higher than would be expected from considerations of chain scission. This suggests that there must still be a considerable amount of plastically dissipated energy involved in the fracture process.

Calculation of the intrinsic flaw size gives a value which lies within the range of experimentally measured cell dimensions.

The author would like to thank Mr. T. Flanagan of the Department of Polymer Technology, Manchester Polytechnic, for his assistance and advice on the preparation of foam samples.

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Received January 14, 1975