Effect of Geometric Constraints on Toughness of Polycarbonate

K. SEHANOBISH, D. H. BANK and C. P. BOSNYAK
Dow Chemical Company, Freeport, Texas, USA

ABSTRACT

Polycarbonate(PC) has been observed to undergo a transition in failure mechanism from ductile to brittle with increasing thickness. However, little effort has been made to explain this behavior quantitatively. The difficulty in pursuing this problem was discovered to arise from the presence of intrinsic defects in the specimens which overshadowed the effects of thickness. A statistical approach has been developed relating the probability of breaking strength (brittle or ductile) with the distribution of defect size. Further by introducing a well characterized notch (extrinsic defect), one can distinguish the thickness contribution to the failure mechanism in PC. The results reveal that PC fails by varying proportions of two dominant deformation mechanisms. An experimental method is presented here for the estimation of the amount of energy dissipated in each mechanism.

KEYWORDS

Toughness; Polycarbonate; Crack; Surface energy; Brittle; Ductile.

INTRODUCTION

Polycarbonate deforms in a ductile manner at room temperature (Brinson, 1970; Hyakutake and Nishitani, 1985; Donald and Kramer, 1981a,b) although a ductile to brittle transition was shown to depend on the rate of loading (Parvin and Williams, 1975), temperature (Hyakutake and Nishitani, 1985; Pitman and Ward, 1980; Martin and Gerberich, 1976) and the thickness of the specimen (Pitman and Ward, 1979, 1980; Haddaoui et. al., 1985). Furthermore, for notched specimens of constant thickness, brittle fracture was observed under monotonic loading for sharply notched specimens and fully ductile failure was noted for large notch root radii (Hyakutake and Nishitani, 1985; Dekkers and Hobbs, 1987). Historically, the embrittlement of materials with increasing thickness has been explained by the concept of transition from a plane stress to plane strain state (Pitman and Ward, 1979, 1980; Williams, 1983; Irwin, 1960; Pisarski, 1981). However, little effort has been made to quantitatively account for the presence of defects and relate fracture toughness to the failure mechanisms. Realizing that the stress field as well as the characteristics of the damage zone ahead of the crack tip (Dekkers and Hobbs, 1987; Kitagawa, 1982; Lee et al., 1987) play an important role in the resistance of polycarbonate to fracture (i.e. "toughness"),this paper discusses the concept of relating fracture toughness to the associated damage mechanisms.

EXPERIMENTAL

Pellets of PC, Calibre-22, Dow Chemical Co., were dried in a forced air oven for four hours at 120°C. The dried pellets were first injection molded to 1/8" thickness. Then sample plaques of thicknesses 1/4", 1/8" and 1/64" were compression molded in a Tetrahedron MTP-14 programmable molding machine from the injection molded plaques. Dumbell shaped specimens were machined from the plaques according to ASTM D638 for tensile testing. Single edge notched (SEN) specimens were also prepared from the same plaques for fracture toughness tests.

All notched and unnotched tensile tests were done in an MTS servo-hydraulic machine and the microscopic analyses were performed on a Zeiss optical microscope.

RESULTS AND DISCUSSIONS

Figure 1 describes the effect of thickness and strain rate on the breaking

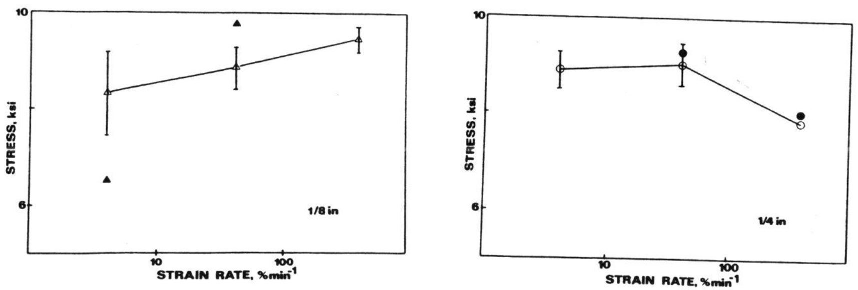
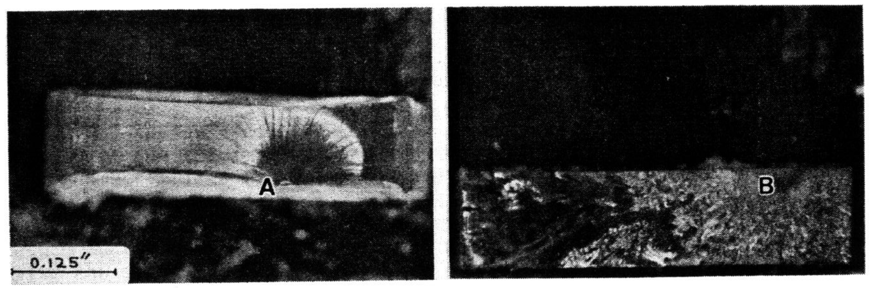


Fig. 1. Breaking Stress as a function of strain rate at two different thickness. Open symbols represent ductile failure and closed symbols represent brittle failure.

stress of PC. Within the strain rate range tested, the breaking stress does not correlate well with the applied strain rate. The open symbols represent situations where the specimen underwent yielding and necking prior to failure, while the filled symbols represent catastrophic failure without yielding. Failure mechanisms are inconsistent in both 1/8" and 1/4" thick

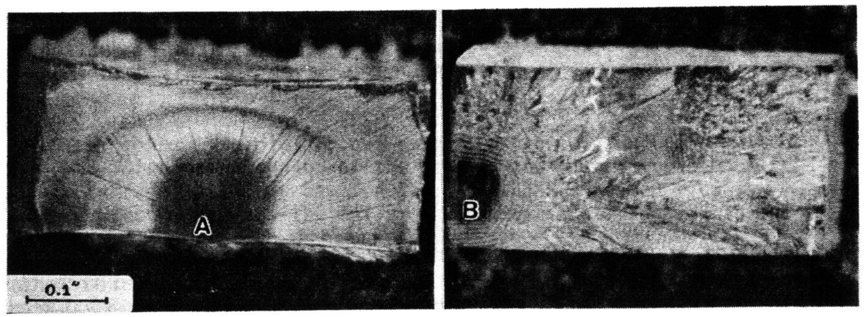
samples , i.e., some of the specimens yielded while a few failed catastrophically. In order to explain this effect, we examined the fracture surfaces. Figure 2a shows the fracture surface of a 1/8" thick specimen that failed by yielding. Note the distinct features of flow in the yielded sample associated with thinning. The fracture surface in figure 2b displays random features where crack propagated at various planes creating complex surfaces. Discontinuous crack propagation in each of these planes are



a b

Fig. 2. Fracture surface of 1/8" thick PC samples (a) Ductile and (b) Brittle.

marked by periodic striations. In both situations, one can trace the failure initiation site marked "A" in figure 2a and marked "B" in figure 2b. These phenomena were also observed in 1/4" thick specimens (Figures 3a and 3b). The key recognition gained here is failure is always controlled by a defect in these tests. Thus, it is necessary to develop a statistical tool to predict the uncertainties of failure based on defect size distribution.



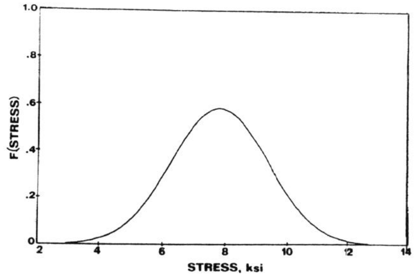
a b

Fig. 3. Fracture surface of 1/4" thick samples (a) Ductile and (b) Brittle.

In order to establish such an approach, one needs to describe the probability of breaking stress by an appropriate probability density function such as shown in equation 1 (Chudnovsky and Kunin, 1987):

$$f(\sigma) = \frac{1}{\sqrt{2 \pi \phi^2}} \exp^{-}(\sigma-\mu)^2/2\phi \qquad (1)$$

where, μ is the mean stress and ϕ is the variance of the breaking stress. In order to establish the validity of the distribution shown above, one needs to run a statistical set of samples. Figure 4a displays the breaking stress distribution. In order to calculate the critical defect size



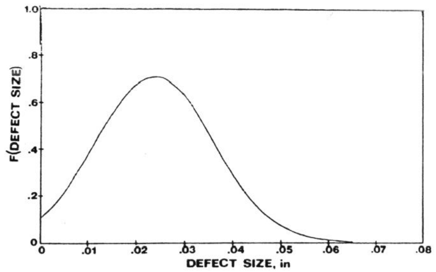


Fig. 4. Statistical distribution of (a) Breaking Stress and (b) Critical Defect size.

distribution from the breaking stress distribution, one must first consider a proper failure initiation criterion. Observing the initiation site as indicated by markers A and B in Figures 2a,b and 3a,b, it appears that for both brittle and ductile fracture, failure initiation involved creation of two virtually mirror-like surfaces followed by propagation of failure through yielded or unyielded material. Thus, the criterion proposed by Griffith which states that the critical energy release rate is equal to twice the surface free energy (γ) seems appropriate. This is described mathematically as

$$\frac{\sigma^2 \pi r}{E} = 2\gamma \qquad (2)$$

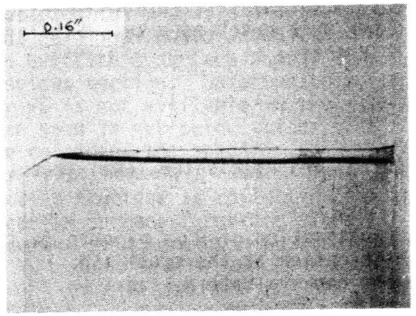
where σ is the applied stress, r is the defect size and E is the Young's modulus. Using equations 1 and 2, one can derive the defect size distribution. The distribution of critical defect size evaluated from the above equation is shown in figure 4b. γ has been assumed to be 10 J/m^2 in these calculations only to demonstrate the usefulness of this approach. The validity of this assumption is considered in the next section. One must note that the critical defect size corresponding to the peak of the distribution curve (Figure 4b) is similar to the defect sizes marked in

Figures 2a and 3a. This justifies the validity of a brittle initiation criterion. Thus if the distribution of the maximum defect size is known, one can reconstruct the probability of breaking stress using the failure criterion in equation 2.

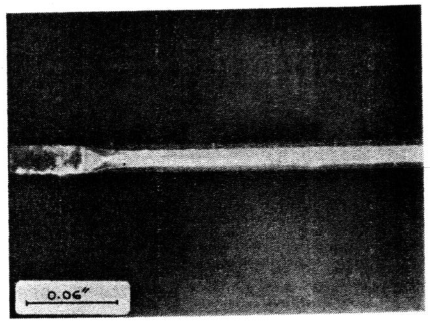
In order to distinguish the thickness contribution, an extrinsic defect, i.e., a notch larger than all possible intrinsic defects, was introduced so that failure is initiated at the defect site. Monotonic tensile load was applied to the single edge notched specimen at a displacement rate of 100 in/min. Since the load displacement relationship is practically linear at this rate, one may evaluate fracture toughness, Jlc,

$$J_{1c} = \frac{\sigma_\infty^2 \pi a}{E} f^2(a/W) \qquad (3)$$

where, σ_∞ is the critical remote stress, "a" is the crack length, and W is the width of the specimen. $f^2(a/W)$ is the correction factor applied for finite geometry (Tada et al., 1973). J_{1c} has been evaluated as 24.3 ± 0.8 kJ/m^2 for 1/64" thick samples and 1.3 ± 0.2 kJ/m^2 for 1/4" thick samples. In order to explain the toughness variation due to thickness changes, fracture surfaces of the two specimens were examined. Figures 5a and 5b display the side view and the fracture surface of a 1/64" thick specimen.



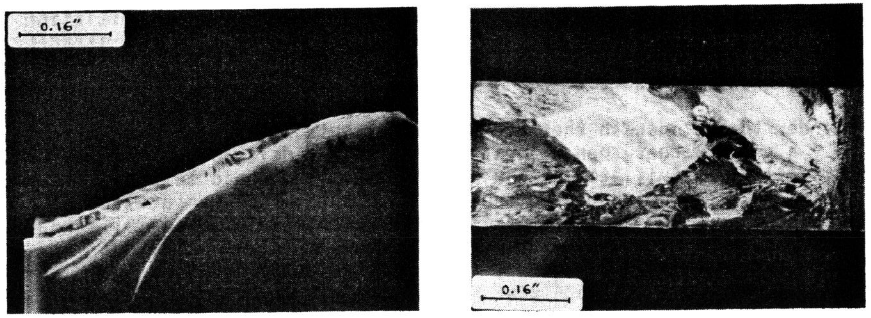
a



b

Fig. 5. Optical micrographs of cracked PC specimens (1/64") (a) Side view (b) Fracture surface.

At this thickness, yielding (flow) appears as the primary deformation mechanism. Distinct thinning of the yielded zone is noticeable on the fracture surface in Figure 5a. In the 1/4" thick specimen flow is completely restricted and the failure occurs by shattering (explosion-like), resulting in the creation of fragments. A side view devoid of these fragments and a fracture surface highlighting the random nature of the brittle failure are illustrated in figures 6a and 6b. Thus, thickness a transition in deformation mechanism and the toughness parameter J_{1c}, defined by fracture mechanics, needs to be expressed as a product of two independent variables. One of the variables (γ^*) reflects the specific energy associated with material transformation around the crack tip and the other (R) expresses the total amount of material transformed during crack growth. When the deformation mechanism involves volumetric transformation by



a b

Fig. 6. Optical micrographs of cracked PC samples (1/4") (a) Side view (b) Fracture surface.change causes

yielding, J_{1c} may be expressed as

$$J_{1c} = \gamma_1^* R_{1c} \tag{4}$$

where, γ_1^* is the energy per unit volume of material transformed and R_{1c} is the total volume of material transformed per unit crack advance. Similarly, one may express J_{1c} for pure brittle fracture as

$$J_{2c} = \gamma_2^* R_{2c} \tag{5}$$

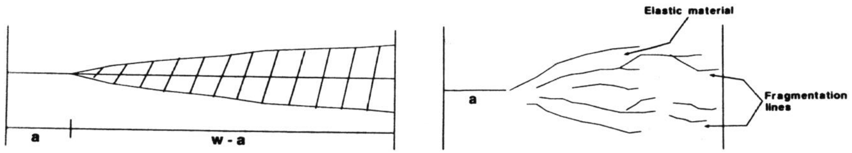
where γ_2^* is the Griffith type surface energy and R_{2c} is the total amount of surface created.

A simple methodology is proposed here for estimation of R_{1c}. Figure 7a is a schematic describing the volumetric transformation at the crack tip. Knowing the area of the shaded zone, A, one can evaluate R_{1c} as

$$R_{1c} = \frac{A\ t}{(W-a)\ t_0} \tag{6}$$

where t is the average thickness of the yielded material, t_0 is the original thickness, a is the crack length and W is the width of the specimen. γ_1^* was estimated as 60 J/g for yielding of PC (Haddaoui et al., 1985; Bosnyak et al., 1988). Brittle fracture as evidenced in 1/4" thick samples leads to surface creation as opposed to volumetric transformation during yielding. A crude estimate of the surface area involves measuring the total length of the fragmentation lines, L, as illustrated in figure 7b. It is assumed that the material between the fragmentation lines remain unaffected, so that the surface area may simply be expressed as

$$A = L\ t_0 \tag{7}$$



a b

Fig. 7. (a) Shaded zone shematically represents yielded material in front of crack (1/64") and (b) Schematic of Fragmentation lines (1/4").

Consequently, R_{2c} can be derived as

$$R_{2c} = \frac{L}{(W-a)} \tag{8}$$

Thus equation 8 should provide us an estimate of R_{2c}, although more careful consideration should account for the complex surfaces. γ_2^* is analogous to the Griffith's surface energy term. Thus, γ_2^* should be the same as in the previous section. Unfortunately, the magnitude of Griffith's surface energy for PC is not available in the literature. However a γ_2^* of 10 J/m^2 has been used in our calculations. This number seems to be quite acceptable if one considers the specific energy of crazing of amorphous PS (Haddaoui et al., 1983; Andrews, 1988). Thus the loading history dependence for J_{1c} for ductile fracture is reflected through R_{1c}. For pure brittle fracture that leads to surface formation alone, the parameter R_{2c} is more dependent on material structure than on loading history.

CONCLUSIONS

At a given thickness, the size and distribution of defects control failure in PC. A statistical approach has been developed to predict the probability of failure from the defect size distribution. Thick specimens of polycarbonate primarily respond to the defect by fragmentation, while thin specimens of polycarbonate respond to the defect through localized yielding. The fracture toughness of the specimen at a given thickness can be expressed as a product of the specific energy associated with the failure mechanisms (brittle or ductile) and the amount of deformed material (in volume or surface area).

ACKNOWLEDGEMENT

The authors would like to gratefully acknowledge Allison Nixon, Jim Wilchester and Ralph Guerra for their help with the experiments and Prof. A. Chudnovsky and Che Kao for useful discussion and critique.

REFERENCES

Andrews, E. H. (1988). Queen Mary College, London, Personal Communications.
Bosnyak, C. P., J. K. Rieke, V. Krongauz and A. Chudnovsky (1988). Evaluation of a Spiropyran as a Molecular Probe for the Large Strain Deformation of Polycarbonate. 7th Int. Conf. Def. Yield & Fract. of Polymers, Cambridge, U.K., 14/1-14/4.
Brinson, H. F. (1970). The Ductile Fracture of Polycarbonate. Experimental Mechanics, 10, 72-77.
Chudnovsky, A. and B. Kunin (1987). A Probabilistic Model of Brittle Crack Formation. J. Appl. Phys., 62, 4124-4129.
Dekkers, M. E. J. and S. Y. Hobbs (1987). Ductile Tearing Instability in some Engineering Thermoplastic Blends. Polym. Eng. Sci., 27, 1164-1169.
Donald, A. M. and E. J. Kramer (1981a). Plane Stress Deformation Zones at Crack Tips in Polycarbonate. J. Mater. Sci., 16, 2967-2976.
Donald, A. M. and E. J. Kramer (1981b). Micromechanics and Kinetics of Deformation Zones at Crack Tips in Polycarbonate. J. Mater. Sci., 16, 2977-2987.
Haddaoui, N., A. Chudnovsky and A. Moet (1983). Determination of the Specific Enthalpy of Fracture of Polystyrene. Polym. Mater. Sci. & Eng., 49, 117-123.
Haddaoui, N., A. Chudnovsky and A. Moet (1986). Ductile Fatigue Crack Propagation in Polycarbonate. Polymer, 27, 1377-1383.
Hyakutake, H. and H. Nishitani (1985). Condition for Determining the Static Yield and Fracture of a Polycarbonate Plate Specimen with Notches. Eng. Fract. Mech., 22, 359-368.
Irwin, G. R. (1960). Fracture Mode Transition. J. Bas. Eng., 82, 417-425.
Kitagawa, M. (1982). Plastic Deformation and Fracture of Notched Specimens due to Bending in Glassy Polymers. J. Mater. Sci., 17, 2514-2524.
Lee, L. H., J. F. Mandell and F. J. McGarry (1987). Fracture Toughness and Crack Instability in Tough Polymers under Plane Strain Conditions. Polym. Eng. Sci., 27, 1128-1136.
Martin, C. C. and W. W. Gerberich (1976). Fracture Toughness Predictions in Polycarbonate. J. Mater. Sci., 11, 897-902.
Parvin, M. and J. G. Williams (1975). Ductile-Brittle Fracture Transitions in Polycarbonate. Int. J. Fract., 11, 963-972.
Pisarski, H. G. (1981). Influence of Thickness on Critical Crack Opening Displacement(COD) and J values. Int. J. Fract., 17, 427-440.
Pitman, G. L. and I. M. Ward (1979). Effect of Molecular Weight on Craze Shape and Fracture Toughness in Polycarbonate. Polymer, 20, 895-902.
Pitman, G. L. and I. M. Ward (1980). The Molecular Weight Dependence of Fatigue Crack Propagation in Polycarbonate. J. Mater. Sci., 15, 635-645.
Tada, H., P. C. Paris and G. P. Irwin (1973). The Stress Analysis of Cracks Handbook. Del Research Co., Hellertown, Pennsylvania.
Williams, J. G. (1983). Fracture Mechanics of Polymers, pp. 132-143. Halsted Press, New York.