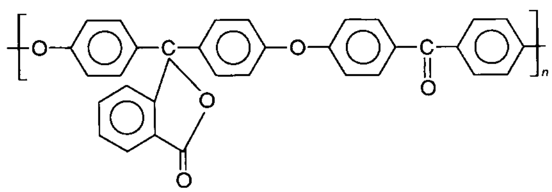
Ductile Tearing Instability in Phenolphthalein Poly(ether ketone)

YANCHUN HAN,* YUMING YANG, BINYAO LI, and ZHILIU FENG

Changchun Institute of Applied Chemistry, Academia Sinica, Changchun, 130022, People's Republic of China

SYNOPSIS

Phenolphthalein poly(ether ketone) (PEK-C)



can fail by tearing instability when the elastic contraction is greater than the plastic extension due to crack growth. Tearing instability (TIS) theory developed by Paris and coworkers describes the effect of specimen geometry on the ductile fracture properties of polymers. The stability of crack growth in three-point bend specimens depends on the specimen's dimensions. In this article, tearing instability theory is applied to describe the ductile tearing instability of PEK-C at different temperatures. The temperature dependence of tearing modulus T_m, $dJ/d\Delta a$, and the relationship between $dJ/d\Delta a$ and yield stress are discussed. © 1994 John Wiley & Sons, Inc.

INTRODUCTION

Polymeric materials are increasingly being used for load-bearing structural applications and, therefore, understanding of their fracture properties is becoming more important. Linear elastic fracture mechanics (LEFM) has successfully described the fracture properties of brittle polymers, e.g., polypropylene (PP), below their glass transition temperature.[1,2] However, many polymers of great engineering interest are ductile and predictions of failure based on LEFM can be extremely conservative for these materials, such as for polyethylene and PP. PP is above its applied tearing modulus (Ta). Significant progress has been made in characterizing the initiation of crack growth in ductile materials, primarily metals, in terms of the J integral concept discovered by Rice[3] and suggested by Begley and Landes[4] as a fracture criterion that has resulted in a standard procedure.[5] Recently, Hodgkinson and Williams[6] successfully described the initiation of ductile fracture in polyethylene by the J integral.

The J integral may be interpreted in two ways: (a) the intensity of the elastic–plastic deformation and stress field in the crack tip region or (b) the change in energy of the cracked body due to a small extension of the crack. However, since sustained crack growth in ductile materials requires additional energy, even the J integral value for initiation is conservative in describing the fracture resistance and only describes the initiation of crack growth. Therefore, the J resistance curve (J vs. change in crack length) is commonly used to characterize the crack extension behavior of ductile materials.

Paris et al.[7] extended this approach to characterize the conditions for crack instability in a ductile material in terms of a nondimensional parameter, T_m, the tearing modulus. This concept postulates

* To whom correspondence should be addressed.

Journal of Applied Polymer Science, Vol. 53, 19–22 (1994)
 CCC 0021-8995/94/010019-04

that instability occurs if the elastic shortening of the system exceeds the corresponding plastic lengthening required for crack extension.

The ductile tearing instability is the sudden transition from slow, controlled ductile crack growth to fast, uncontrolled brittle crack growth. It is generally recognized that this instability results from a lack of balance between the applied driving force for crack growth and the material's resistance to crack growth. In the past decade, much effort has been devoted to describing ductile crack growth and crack instability in metals because of the need to design highly critical structures for use in the ductile failure range. Relatively little attention has been paid to these phenomena in polymers, although many polymers and polymer blends are ductile and predictions of failure based on linear elastic fracture mechanics can be very conservative.

This article presents the results of an investigation into the temperature dependence of the ductile tearing instability in phenolphthalein poly(ether ketone) (PEK-C). Crack growth in a single-edged notched (SEN) specimen was studied in three-point bending over a wide range of temperatures. Attention was focused on the effects of temperature on crack growth stability.

Tearing Modulus

Paris and co-workers developed the tearing modulus concept to describe the stability of ductile crack growth in terms of elastic–plastic fracture mechanics. This concept postulates that instability occurs if the elastic shortening of the system exceeds the corresponding plastic lengthening required for crack extension. In this theory, a nondimensional parameter called the tearing modulus, T_m, that has the form

$$T_m = \frac{E}{\sigma_y^2} \frac{dJ}{d\Delta a} \tag{1}$$

where E is the Young's modulus; σ_y; the yield stress; J, the J integral; and Δa, the crack growth length.

The condition for stable crack growth was then given as

$$T_a = 2b^2 S / W^3 < T_m \tag{2}$$

where T_a was defined as the applied tearing modulus, depending on the specimen size and configuration. T_a's for different specimen geometries are given in Table I.

Table I T_a Applied for Various Specimens

Specimen	Loading	T_a
1. Centre crack strip	Tension	$2L/W$
2. Double-edged notched	Tension	$12L/W$
3. Three-point bend	Bending	$2b^2S/W^3$
4. Compact tension specimen	Tension	Negative —hence, always stable

L is the length; W, the thickness; b, the ligament of the specimen; and S, the span.

EXPERIMENTAL

The materials used were phenolphthalein poly(ether ketone) (PEK-C) supplied by Xu Zhou Engineering Plastic Co. China. The specimens were made by injection molding. Test specimens were three-point bend bars of dimensions of $B = 8$ mm, $W = 16$ mm, and length, $L = 80$ mm. The single-edge initial notch of the specimen was sharpened with a fresh razor blade. The notch length, a, of 8 mm ($a/W = 0.5$) was given. Tests were performed on an Instron testing machine. A span-to-width ratio of 8 was used ($S = 4W = 8B$). At each temperature at a crosshead speed of 5 mm/min, a series of identical specimens were loaded to various deflections corresponding to different amounts of crack extensions and then unloaded. Measurements of the crack extension Δa were made on the fracture surface using a traveling microscope.

RESULTS AND DISCUSSION

Temperature Dependence of $dJ/d\Delta a$

The resistance of the material to stable crack growth should be expressed by the values of the slope of the J_R curve, $dJ/d\Delta a$. The dependence of $dJ/d\Delta a$ on temperature is shown in Figure 1. It shows that decreasing the temperature increases $dJ/d\Delta a$. So, we can conclude that the fracture mode becomes more unstable with lower temperature. $dJ/d\Delta a$ may be correlated with the yield stress of the material, as shown in Figure 2.

Temperature Dependence of T_m

According to eq. (1) and the values of $dJ/d\Delta a$, E, and σ_y at different temperatures (Table II), T_m was obtained as a function of temperature, as shown in

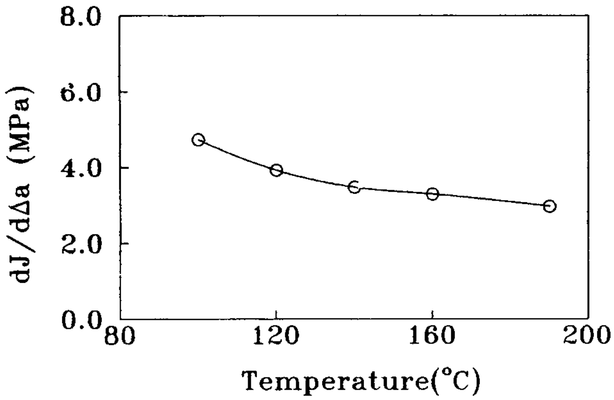


Figure 1 $dJ/d\Delta a$ of PEK-C as a function of temperature.

Figure 3. It increases with increasing temperature. This indicates that decreasing temperature is less resistant to tearing instability because the plastic deformation mechanisms (crazing, shear yielding, or both) occurring in front of the running crack tip are less effective in dissipating energy and stabilizing crack growth.

For a three-point bend, $W = 16$ mm, $W - a = 8$ mm, and $S = 64$ mm, according to $T_a = 2(W - a)^2 S/W^3$ $T_a = 2.00$ is obtained; when temperature is lower than 160°C, $T_a > T_m$, predicting unstable tearing, and when temperature is higher than 160°C, $T_a < T_m$, predicting stable tearing, as observed. In general, both T_m and T_a are not constant in a given test, but change in value as crack growth proceeds.[8,9]

In recent years, significant effort has been devoted to experimentally verifying the tearing modulus concept. Under conditions of plane-strain and full yielding of the remaining ligament of the cracked section before crack extension, the determination of T_m and T_a is rather straightforward and the instability criterion in eq. (2) has been shown to predict successfully instability in both metals and polymers.[10–15] However, sometimes, the conditions of plane-strain and full yielding are not met. First, the relative specimens used do not ensure plane-strain conditions. Second, yielding before crack extension is most often confined to the zone around the notch, so the remaining ligament is not fully plastic but partly elastic.

Table II E, σ_y, J_{IC}, T_m, and $dJ/d\Delta a$ as a Function of Temperature

T (°C)	100	120	140	160	190
E (GPa)	1.49	1.41	1.39	1.36	0.80
σ_y (MPa)	71.12	59.83	55.75	46.78	31.60
$dJ/d\Delta a$ (MPa)	4.72	3.92	3.47	3.29	2.96
T_m (l)	1.38	1.54	1.55	2.05	2.37
J_{IC} (kJ/m^2)	2.17	2.70	2.88	3.55	3.75

Although tearing instability (TIS) theory[7] (this theory postulates that instability occurs if the elastic shortening of the system exceeds the corresponding plastic lengthening required for crack extension) explains most of the results, two of the results need further study and clarification:

(1) T_m was not independent of geometry, and

(2) unstable fracture occurred at the higher deformation rates even though T_m was greater than T_a.

Some results showed that T_m determined from data obtained from CT specimens was two times that obtained from three-point bend (TPB) speci-

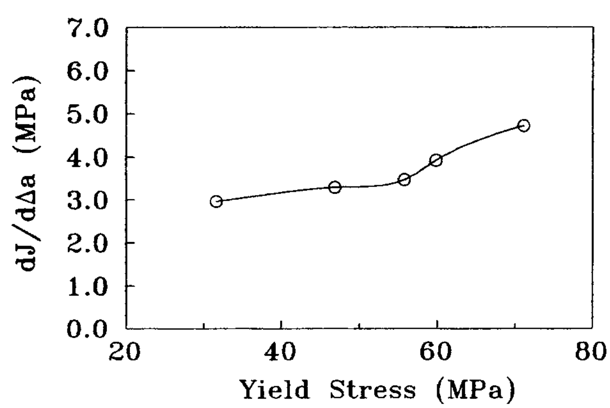


Figure 2 $dJ/d\Delta a$ vs. yield stress of PEK-C.

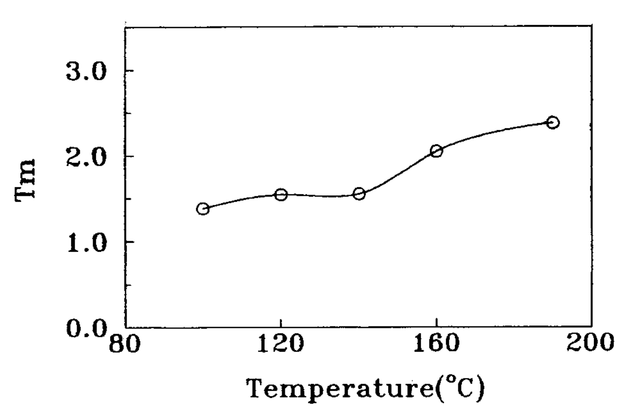


Figure 3 T_m of PEK-C as a function of temperature.

mens. J_{IC} was the same for both specimens and, therefore, is considered independent of geometry. The source of this difference may be due to the different plastic zones in the two specimens, and the extent of plastic deformation as a function of load could be different and thus account for the different T_m. Other investigators studying metals have made similar observations and, consequently, T_m is not independent of the geometry of the specimen and may not be a material property.

At the higher strain rates, some specimens failed in an unstable way even though T_a was less than T_m, requiring some comment. A possible explanation is that as the deformation rate increases the yield strength increases and, therefore, the size of the plastic zone decreases and the remaining ligament was not fully plasticized. The values of T_a calculated were based on the assumption that full plasticity in the remaining ligament effectively increases T_a and unstable crack growth becomes probable. Another manifestation of the effect of deformation rate on the extent of plasticity is the decrease of both J_{IC} and T_m with an increasing deformation rate. Because the extent of plasticity is reduced, the critical crack driving force and the resistance to crack growth T_m are decreased. These are not unexpected results.

Overall, the theory of TIS explains many of the observations regarding the ductile fracture previously unexplained and provides a basis for better characterization of the ductile fracture properties of polymers.

CONCLUSIONS

1. J_{IC} was independent of specimen geometry, but T_m was not.
2. J_{IC} and T_m increased with temperature; $dJ/d\Delta a$ decreased with temperature.
3. $dJ/d\Delta a$ may have a correlation with the yield stress of the material.
4. TIS theory can also explain temperature dependence of crack growth stability.

REFERENCES

1. K. Friedrich, *Prog. Coll-Polym. Sci.*, **64,** 103 (1977).
2. K. Friedrich, *Prog. Coll-Polym. Sci.*, **66,** 299 (1979).
3. J. R. Rice, *J. Appl. Mech.*, **35,** 379 (1968).
4. J. A. Begley and J. D. Landes, *The J-Integral as a Fracture Criterion'*, ASTM STP 514, American Society of Testing and Materials, Philadelphia, PA, 1972, PP1–20.
5. ASTM Standard E813, 1982 Annual ASTM Standard, Part 10, PP822–840.
6. J. M. Hodgkinson and J. G. Williams, *J. Mater. Sci.*, **16,** 50 (1981).
7. P. C. Paris, H. Tada, A. Zahoor, and H. Ernst, *Elastic-Plastic Fracture*, ASTM STP 688, American Society of Testing Materials, Philadelphia, PA, 1979, p. 5.
8. P. C. Paris and R. E. Johnson, in *Elastic-Plastic Fracture: Second Symposium*, ASTM STP 803, II-5 (1983).
9. H. A. Ernst, in *Elastic-Plastic Fracture: Second Symposium*, ASTM STP 803, II-133 (1983).
10. T. Vu-Khanh, *Polymer*, **29**(11), 1979 (1988).
11. C.-B. Lee and F.-C. Chang, *Polym., Eng. Sci.*, **32**(12), 792 (1992).
12. S. Hashemi and J. G. Williams, *J. Mater. Sci.*, **26,** 621 (1991).
13. I. Narisawa, *Polym. Eng. Sci.*, **27**(1), 41 (1987).
14. S. Hashemi and J. G. Williams, *Plast. Rubber Process. Appl.*, **6**(4), 363 (1986).
15. J. G. Williams, *Fracture Mechanics of Polymers*, Wiley, Chichester, UK, 1984, p. 233.

Received September 13, 1993
Accepted December 13, 1993