Assessment of Temperature-Dependent Fracture Behavior with Different Fracture Mechanics Concepts on Examples of Unoriented and Cold-Rolled Polypropylene

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ABSTRACT: Applicability of different fracture mechanics concepts, including linear elastic fracture mechanics (LEFM), equivalent energy concept, and elastic-plastic fracture mechanics (EPFM), to assessing the temperature-dependent fracture behavior was compared using examples of an unoriented and a cold-rolled polypropylene under quasistatic and under dynamic loading. Under quasistatic loading, the fracture toughness values were determined from the recorded load versus load-line displacement curves on compact tension (CT) specimens. Fracture toughness values under dynamic loading were determined from the recorded load versus deflection curves on single edge-notched bend (SENB) specimens. In spite of its simplicity as an engineering design parameter, on the basis of the LEFM concept, the stress intensity factor K can only be validly used in a limited temperature range. Instead, the EPFM parameters (i.e., the J integral and the crack opening displacement (COD) concepts) can be applied over a wider temperature range. © 1997 John Wiley & Sons, Inc. J Appl Polym Sci **66:** 1237–1249, 1997

Key words: cold rolled polypropylene; fracture mechanics parameters; temperature; strain rate

INTRODUCTION

Investigations of the temperature dependence of fracture toughness in polymeric materials are of special technical importance with respect to identifying the ductile–brittle transition temperatures as well as to setting the boundaries of operating temperatures.[1] Owing to viscoelastic behavior in polymeric materials, the molecular relaxation processes play a decisive role, as for instance in directly affecting the time (loading rate)–temperature-dependent fracture processes. Moreover, the relationship between toughness and relaxation behavior is of particular practical interest because, within the operating temperature range, the relaxation time necessary to build up the equilibrium state is frequently on the order of the loading duration.[2]

The application of fracture mechanics concepts to the evaluation of temperature-dependent fracture behavior is becoming increasingly important because fracture mechanics parameters are extraordinarily sensitive to structural changes. Linear elastic fracture mechanics (LEFM) is widely applied to brittle polymers[1,3–8] because the size of the plastic zone is much smaller than the in-plane specimen dimensions and the initiation of unstable fractures can accurately be described by the critical stress intensity factor K_{IC}. For tough polymers, the size requirement for a valid K_{IC} value is often so strict that fracture testing by using the LEFM concept becomes impractical, and hence the concepts of elastic-plastic fracture mechanics

Dedicated to Prof. Dr. H. Hoffmann for his 60th birthday.

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(EPFM)[9–11] have been introduced. Several investigations have been carried out under quasistatic loading[12–14] as well as under dynamic loading.[15–17]

Various temperature dependencies (with or without a local or global maximum, or both, in terms of stress intensity factor) have been reported for different polymeric materials. Frequently, the measured fracture mechanics parameters represented an average apparent value across the specimen thickness and hence included the thickness effect. The different crack-tip constraints at different temperatures may have intensified the thickness effect. The fracture toughness values were found to increase with increasing temperature[5,18,19] but a decrease with increasing temperature has also been reported.[7,8,10,20–24] Sometimes, the valid and invalid LEFM fracture mechanics values were used together to analyze the temperature dependence.[5] Nevertheless, the coupled effects from the losing crack-tip constraint and the increasing plasticity with increasing temperature, lead to great difficulties in assessing the temperature dependence of fracture behavior by using only one fracture mechanics parameter on specimens with a constant thickness. Moreover, the fracture behaviors and the fracture modes may vary largely in different temperature ranges. Because of these great difficulties, no established procedures to study fracture behavior over a wide temperature range exist. Several different fracture behaviors can be covered extending from brittle (unstable fracture), over brittle–ductile transition (slow crack growth prior to cleavage), to ductile (ductile tearing) fracture behavior. Furthermore, the overlapped main and secondary molecular relaxation processes with temperature make the situation in polymeric materials more complicated.

In this article, the applicability of the different fracture mechanics concepts to characterizing the temperature dependence on fracture toughness is critically examined and discussed with regard to the advantages and drawbacks of each concept. The influence of temperature on the crack resistance behavior has been assessed[2,25,26] and is thus not addressed in the present study. Therefore, only the resistance against unstable fracture is treated. Tensile and fracture tests were conducted under static loading as well as under dynamic loading at different temperatures ranging from −150 to 20°C.

Theoretical Background

Fracture mechanics is based on the assumption that the fracture of components and consequently of material is due to propagation of preexisting flaws. Different fracture mechanics concepts have been proposed to account for various types of nonlinear material behavior (i.e., plasticity, viscoelasticity, and viscoplasticity) as well as dynamic effects. All of these, however, are extensions of LEFM.[27]

The LEFM concept can be applied to measure the plane strain fracture toughness of brittle polymers such as those at low temperatures considered in this article. When a material behaves in a linear elastic manner prior to failure such that the plastic zone is small compared with the in-plane specimen dimensions, a critical value of the stress intensity factor K_{IC} may be an appropriate fracture mechanics parameter. According to ESIS TC4 and ASTM,[28,29] the fracture mechanics parameters K_C^{LEFM} for static loading and K_d^{LEFM} for dynamic loading can be evaluated via the following equations:

$$K_C^{\mathrm{LEFM}}; K_d^{\mathrm{LEFM}} = \frac{F}{B\sqrt{W}} f\left(\frac{a}{W}\right) \tag{1}$$

with

$$f\left(\frac{a}{W}\right) = \frac{2 + \frac{a}{W}}{\left(1 - \frac{a}{W}\right)^{3/2}} \left[0.886 + 4.64\left(\frac{a}{W}\right) - 13.32\left(\frac{a}{W}\right)^2 + 14.72\left(\frac{a}{W}\right)^3 - 5.60\left(\frac{a}{W}\right)^4\right] \tag{2a}$$

for CT specimens and

$$f\left(\frac{a}{W}\right) = \frac{3\frac{S}{W}\sqrt{\frac{a}{W}}}{2\left(1 + 2\frac{a}{W}\right)\left(1 - \frac{a}{W}\right)^{3/2}} \times \left[1.99 - \frac{a}{W}\left(1 - \frac{a}{W}\right)\left\{2.15 - 3.93\left(\frac{a}{W}\right) + 2.7\left(\frac{a}{W}\right)^2\right\}\right] \tag{2b}$$

for SENB specimens, where F is the actual load,

W is the specimen width, and S is the support span.

Under consideration of the stable crack growth prior to fracture, the LEFM concept will be extended to the small-scale yielding fracture mechanics concept. The initial crack length a in eq. (1) is replaced by

$$a_{\text{eff}} = a + a_s, \tag{3}$$

where a_s represents the extent of the slow crack growth, which is small in very brittle structures as well as at high loading speeds or at very low temperatures. The size requirement for a valid plane strain fracture toughness K value (K_{IC} for static loading and K_{Id} for dynamic loading) is very severe:[29]

$$B, (W - a), a > \alpha\left(\frac{K}{\sigma_y}\right)^2 \tag{4}$$

where α has a value of 2.5 and σ_y is the yield stress. This empirical constant value of 2.5 was obtained on the basis of experimental results from metals. Furthermore, it has been experimentally shown that this equation overestimates the size requirement for polymers.[12,30] Instead, Irwin's suggestion[31] has been demonstrated to be more suitable:

$$B > \frac{2}{\pi}\left(\frac{K}{\sigma_y}\right)^2. \tag{5}$$

More recently, α was found to be no general constant but a material specific constant and was empirically determined to be

$$\alpha = 3466\, K^{-1.73} \quad K \text{ in MPa}\sqrt{\text{mm}} \tag{6}$$

for different polymeric materials under static as well as under dynamic loading.[32] With the application of eq. (6), it will be possible to determine the geometry-independent fracture mechanics parameter K in polymeric materials. However, this possibility does not immediately imply that the stress intensity factor is an appropriate parameter for describing the increased deformation behavior in polymeric materials.

To extend the applicability of the LEFM concept to thinner specimens or ductile materials, Witt and Mager[33] proposed the equivalent energy concept on the basis of dimensional analysis. It was assumed that the nominal load versus load-line displacement curves (F/B^2 versus υ/B) of specimens with analogous geometries but different thicknesses can be described by a single master curve [Fig. 1(a)]. The area under the nominal load versus load-line displacement curve has a dimension of volumetric energy. For the experimental estimation of the fracture toughness K_C^E based on the equivalent energy concept, the load F in eq. (1) is replaced by a pseudoelastic load F_Q^*, which is related to the deformation energy A_G and the initial slope tan α of the load versus load-line displacement curve through the relationship [Fig. 1(b)]:

$$F_Q^* = \sqrt{2A_G \tan\alpha}. \tag{7}$$

By using this method, the fracture behavior of large-sized specimens with elastic material behavior can be estimated from that measured in small-sized specimens that show elastic-plastic behaviors. However, according to its principle this method still belongs to the LEFM concept.[34–36] Because LEFM is only valid as long as nonlinear material deformation is confined to a small region surrounding the crack-tip, its applicability is also very limited. The size requirement can be expressed as

$$B > \alpha_1 \frac{(K^E)^2}{E\sigma_y} \tag{8}$$

where α_1 ranges from 25 to 50 and E is the Young's modulus.[37,38]

In tough materials, such as those in the present study at higher temperatures, LEFM loses its validity, and alternative fracture mechanics concepts are therefore required. Among these, EPFM parameters (the crack opening displacement [COD] δ[39] and the J integral[40]) are frequently applied to materials that exhibit time-independent, nonlinear behaviors, and the size requirements of the J and COD concepts are much looser than that of LEFM. The critical COD value for CT specimens was estimated via the following equation:

$$\delta_C = \frac{V_C}{1 + n\left[\dfrac{a + z}{W - a}\right]} \tag{9}$$

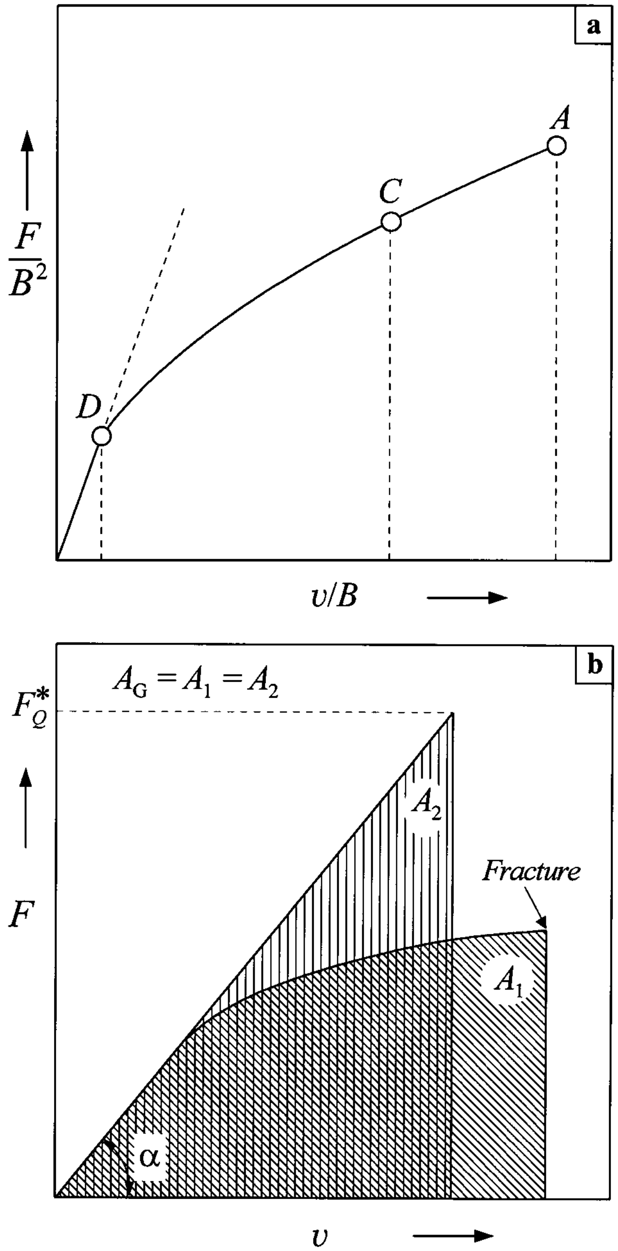


Figure 1 Equivalent energy concept according to Witt and Mager[33]: (a) nominal load versus load-line displacement curves (F/B^2 versus v/B) for geometrically similar specimens A, C, and D with different specimen thicknesses ($B_D > B_C > B_A$), where A, C, and D represent the respective fracture points; (b) determination of the pseudoelastic force F_Q^* with $A_G = A_1 = A_2$.

where V_C is the load-line displacement and z is the distance of the knife edge from the load-line. For single edge-notched bend (SENB) specimens the critical COD value was estimated through the following expression:

$$\delta_{dk} = \frac{1}{n}(W - a)\frac{4f_k}{S}, \tag{10}$$

where f_k describes the crack-tip deformation and is expressed as[41]

$$f_k = f_{\max} - f_B \tag{11}$$

with $f_{\max}$ as the deflection of the notched specimen and f_B as the deflection of an unnotched specimen. The rotational factor n is dependent on the loading level, and the rotation point moves to the crack-tip with further loading. In accordance with ESIS P2-92, the size requirement is related to the critical COD through the following expression:[42]

$$B \geq \xi\delta \tag{12}$$

with ξ at a value of 50. However, in the literature[32,35,43,44] we have shown that ξ is also a material specific constant for polymeric materials:

$$\xi = 3.6\ \delta^{-0.83} \quad \delta \text{ in mm} \tag{13}$$

The J integral was experimentally determined according to the approximate method introduced by Rice, Paris, and Merkle:[45]

$$J = \frac{\lambda(A_p - A_u)}{B(W - a_{\text{eff}})} \tag{14}$$

where A_p is the amount of work done on the notched specimens, A_u is the amount of work done on the unnotched specimens, and $\lambda = f(a/W)$ is the geometry-dependent calibration factor. The size requirement for J-controlling crack resistance behavior is given by[28]

$$B, (W - a), a \geq \epsilon\frac{J}{\sigma_y} \tag{15}$$

with ϵ at a value of 25. Likewise, it was recently found that for polymeric materials ϵ is also dependent on the material properties:[32]

$$\epsilon = 224\ J^{-0.94} \quad J \text{ in N/mm.} \tag{16}$$

EXPERIMENTAL

Materials

The material used was a commercial semicrystalline polypropylene (PP) from Chemiewerk (Eilenburg, Germany), with a mean relative molecular weight of 460,000 and a density of 904.5 kg m^{-3}. In view of methodology, only an unoriented polypropylene structure supplied in the form of as-received injection-molded sheets and a highly oriented polypropylene structure that was introduced by unidirectional cold rolling were tested.

The degree of chain orientation, denoted as f_x, was undetectable in the as-received structures, and that in the rolling direction after the cold rolling was estimated to be about 80% by means of an X-ray diffraction technique using (110)-peak,[46] which enabled identification of the c- as well as the a-texture. In the cold-rolled structure, the c-texture was predominantly observed. Undeformed spherulites were observed in the as-received structure, whereas the cold-rolled one exhibited a stretched spherulitic morphology along the rolling direction without significant change in crystallinity from that measured in the as-received unoriented structure. The crystallinity was determined to be about 53% through the X-ray and differential scanning calorimetry measurements.

The glass transition temperature T_g was measured by a dynamic mechanical testing method at a frequency of 1 Hz. With the tensile loading normal to the rolling direction, the cold rolling increased the T_g of the as-received structure from 6 to 12°C, whereas the loss factor tan δ was somewhat decreased.[30]

Fracture Mechanics Tests

Specimen geometries and dimensions as well as test conditions are summarized in Table I. For the unoriented structure, compact tension (CT) specimens were directly machined from the 10-mm thick as-received sheets, whereas SENB specimens for dynamic testing were milled down to a thickness of 4 mm symmetrically from the two-sided surfaces. The highly oriented structure for CT specimens was introduced by cold rolling a 34-mm thick supplied unoriented injection-molded sheet unidirectionally to a final thickness of 10 mm, and for SENB specimens by cold rolling a supplied unoriented sheet unidirectionally from an initial thickness of 10 mm to a final thickness of 4 mm. These two differently achieved, highly oriented structures were found to have similar spherulitic morphologies, crystallinities, and degrees of chain orientation.

Precrackings were performed at room temperature with a fly cutter, and subsequently the notch was sharpened by sliding a razor blade with a tip radius of 0.2 μm across the machined notch tip.

Table I Experimental: Specimen Geometries, Dimensions, and Test Conditions

Experimental	Tensile Tests		Fracture Tests
	E Modulus	Yield Stress	
	Quasistatic Loading		
Specimens	Tensile Thumbell $B = 4$ mm, $L_0 = 30$ mm		CT specimens $B = 10$ mm, $W = 40$ mm, $a = 20$ mm, $H = 48$ mm, $G = 50$ mm
Crosshead Speed	$v_T = 8.33 \times 10^{-5}$ m/s		$v_T = 8.33 \times 10^{-5}$ m/s
Temperature	−80–20°C		−80–20°C
Testing Machine	Zwick 1386		Zwick 1386
	Dynamic Loading		
Specimens	Tension Vibration	Hopkinson Test	SENB specimens $B = 4$ mm, $W = 10$, $a = 4.5$ mm, $S = 40$ mm, $S/W = 4$
Loading Speed	Frequency $f = 1$ Hz	$v = 10$ m/s	Pendulum Hammer Speed: $v_H = 1$ m/s
Temperature	−150–20°C		−150–20°C
Testing Machine	Tension Vibration		Charpy Impact Tester with Maximum Impact Energy: 4 J

L_0 = length of reduced section; B = specimen thickness; W = specimen width; S = support span; H = specimen height; G = total specimen width.

Both the CT and SENB specimens that were extracted from the cold-rolled structure were prepared in such a way that the specimen orientation corresponded to the principal tensile stress in the transverse direction and crack propagation in the longitudinal direction (Fig. 2).

RESULTS AND DISCUSSION

Yield Strength and Young's Modulus

The temperature dependencies of the Young's modulus E and the yield stress σ_y are shown in Figure 3 at a crosshead speed v_T of 8.33×10^{-5} m/s. Both of them show a typical behavior as observed in different polymeric materials[12] (i.e., decreasing with increasing temperature).

For purposes of comparison, the energy damping spectra at 1 Hz for the unoriented and highly oriented PP structures are presented in Figure 3(a) as a function of temperature. No direct correlation between the tensile properties and the relaxation behavior could be observed. It was further worthwhile noting that the highly oriented PP showed anisotropic tensile properties and relaxation processes.[30]

Temperature Dependence of Toughness Under Quasistatic Loading

The values of fracture toughness K_{IC}^{LEFM} determined on CT specimens under static loading according to the LEFM concept, and K_{IC}^{E} according to the equivalent energy concept, are compared in Figure 4 as a function of temperature in the unoriented as well as in the highly oriented PP. Influences of the chain orientation degrees f_x and the loading directions on the fracture behavior were investigated in considerable detail in the literature[30,46] and are not treated here.

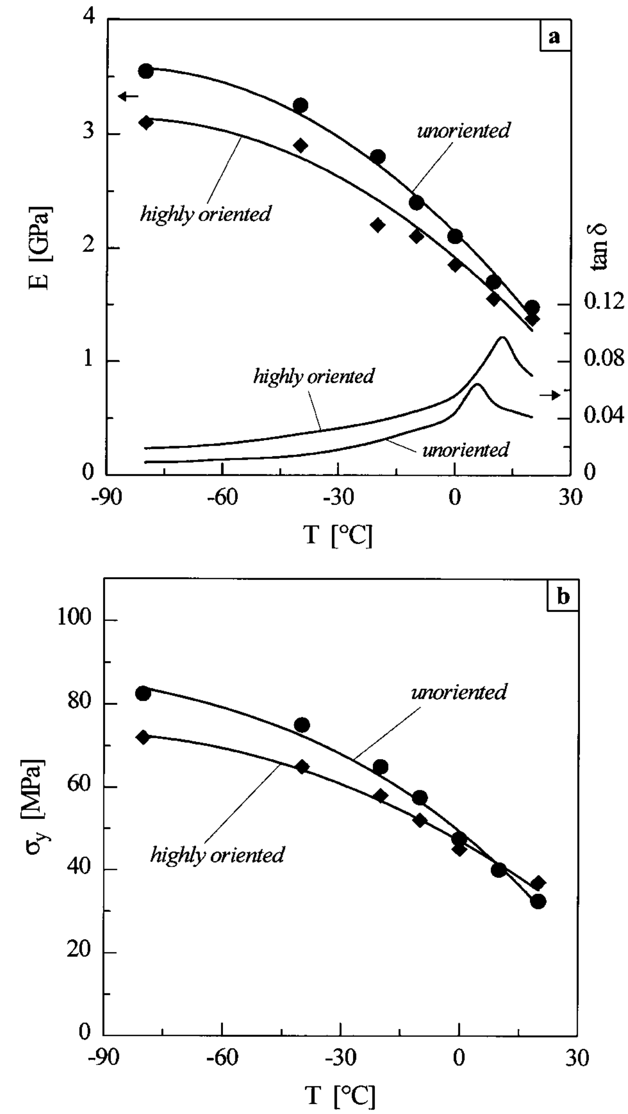


Figure 3 Temperature dependencies of (a) Young's modulus E and loss factor tan δ at a frequency f of 1 Hz and (b) yield strength σ_y ($v_T = 8.33 \times 10^{-5}$ ms^{-1}).

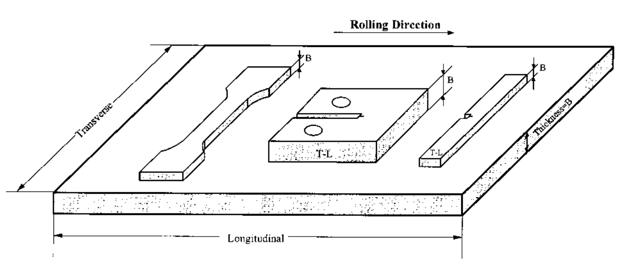


Figure 2 Orientation of fracture mechanics specimens extracted from a cold-rolled plate. The letters L and T denote the longitudinal and transverse direction relative to the rolling direction.

From Figure 4, the problem addressed in the introduction, that the toughness values determined as K_{IC}^{LEFM} decrease with increasing temperature, becomes obvious. This temperature dependence, in contradiction to the real material fracture behavior, is attributed to the fact that, at $T > -20$°C, the validity of the LEFM concept is lost and therefore can not be applied any longer because excessively great plastic deformations occur in the crack-tip zone.

Only with the application of the equivalent en-

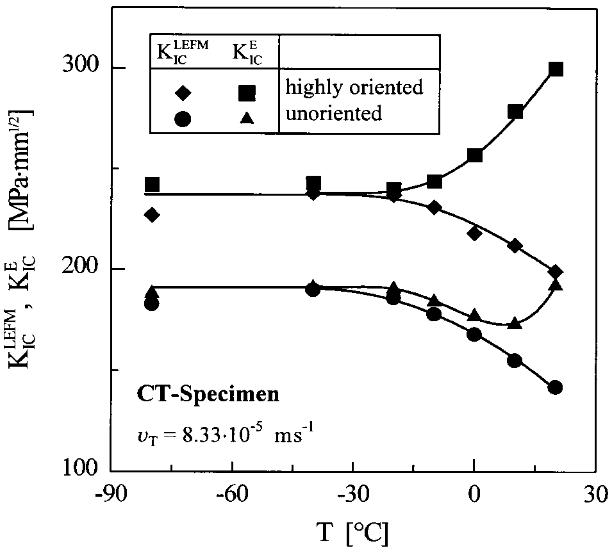


Figure 4 Influences of temperature on the critical stress intensity factors K_{IC} under quasistatic loading: K_{IC}^{LEFM} according to LEFM and K_{IC}^{E} according to the equivalent energy concept.

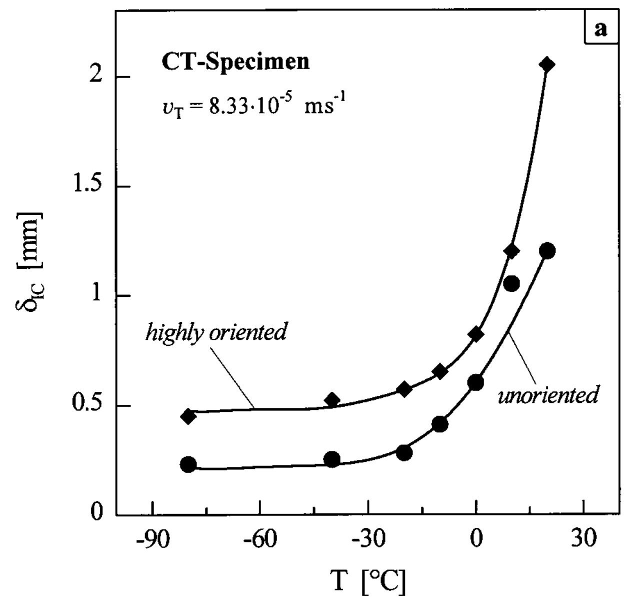
ergy concept and the determination of $K_{IC}^{E}(a_{\text{eff}})$ according to eqs. (1) and (3) can the real toughness values be obtained. In eq. (3) the effective crack growth was taken into account by the calculation of the fracture toughness. It can be further observed that, instead of the LEFM concept, the equivalent energy concept is very useful to assess the temperature dependence of fracture behavior in the tested polypropylene structures.

Figure 5 shows the temperature dependence of the critical COD estimated through eq. (9) and of the J values estimated through eq. (14). With the help of these EPFM parameters, the increasing deformation with increasing temperature is represented in a suitable way.

With the expanding use of polymeric materials, growing demands on decreasing ductile–brittle transition temperature and on increasing thermal stability are required. The ductile–brittle transition temperature is defined as the temperature at which the fracture toughness (e.g. J value) takes the mean toughness value of the lower and the upper shelf. The lower shelf covers the temperature range at which the fracture behavior is athermal. A decrease of the ductile–brittle transition temperature by about 10 K can be determined for the unoriented PP, when compared with the cold-rolled structure, on the basis of the EPFM toughness values in terms of either the J_{IC} or δ_{IC} values.

Temperature Dependence of Toughness Under Dynamic Loading

The temperature effects on the dynamic toughness were investigated on SENB specimens in a temperature range from −150 to 20°C. The difficulty in describing the fracture behavior is especially obvious under dynamic loading. The dynamic fracture toughness, when evaluated ac-



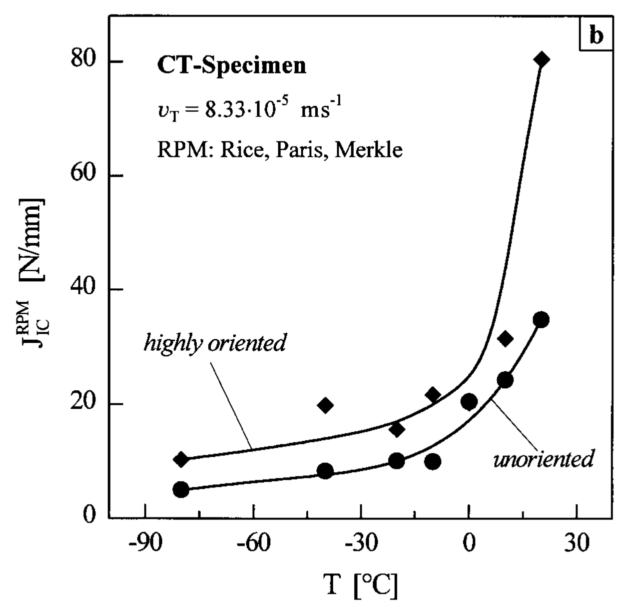


Figure 5 Effects of temperature on (a) the critical crack opening displacement δ_{IC} and (b) the critical J_{IC} values under quasistatic loading.

cording to the LEFM concept, decreases with increasing temperature in the whole temperature range, which does not represent the real toughness behavior of the tested material. With extension of the LEFM concept to the modified small-scale-yielding LEFM concept (i.e. under additional consideration of the effective crack length), a slight rise of fracture toughness can be detected for $T > 0$°C. However, the toughness value of 90 MPa $\sqrt{\text{mm}}$ at $T = -150$°C still remains unreached.

In contrast to the results under static loading, a qualitative description of the fracture toughness is still impossible (Fig. 6) even by using the equivalent energy concept. It is especially obvious that the K_{Id}^{E} values do not lie above that at $T = -150$°C. Apparently, the increasing deformability with increasing temperature under dynamic loading has not been sufficiently represented by the equivalent energy concept.

Only by using the EPFM parameters the (COD and J integral) can the temperature dependence of fracture toughness related to the real material behavior be qualitatively given (Fig. 7). The critical J_{Id} and δ_{Id} values increase with increasing temperature. However, the difficulty mentioned in the introduction still occurs for the unoriented polypropylene structure, although some improvement has been achieved.

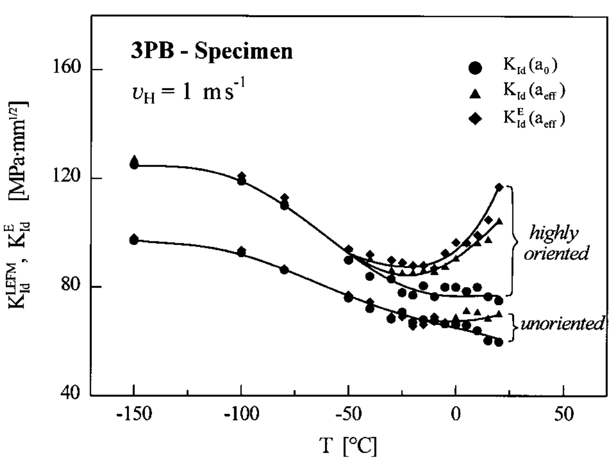


Figure 6 Temperature dependence of the critical stress intensity factors K_{Id} by using different fracture mechanics concepts under dynamic loading: $K_{Id}(a_{\text{eff}})$, $K_{Id}(a_0)$ according to LEFM with and without consideration of the stable crack growth and $K_{Id}^{E}(a_{\text{eff}})$ according to the equivalent energy concept with consideration of the stable crack growth.

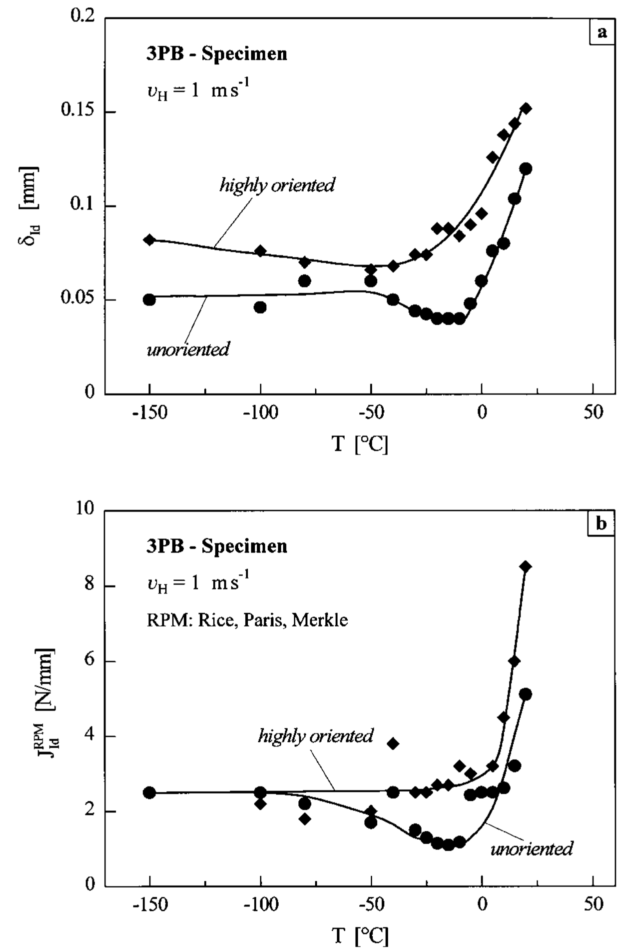


Figure 7 Effects of temperature on (a) the critical crack opening displacement δ_{Id} and (b) the critical J_{Id} values under dynamic loading.

Relationship Among K, J, and δ Values

Quantifying the toughness in terms of K enables the designer to apply linear elastic relationships among stress, flaw size, and toughness. Linear elastic approaches are much simpler and more versatile than a fracture design methodology based on the J integral or COD. As in structural steels, unstable fracture in the ductile-brittle transition region is often preceded by significant plastic flow, and linear elastic parameters are therefore not suitable to characterize toughness. It is preferable to convert critical J values to equivalent K values through the following relationship:

$$K_{IC}^{J};\, K_{Id}^{J} = \sqrt{(J_{IC};\, J_{Id})E^{*}} \tag{17}$$

where E^* is E for the plane stress state, E^* is $E/(1 - \nu^2)$ for the plane strain state, and ν is Poisson's ratio. The expressions J_{IC} and J_{Id} are the critical J values under static and under dynamic loading, respectively.

For linear elastic conditions, the relationship between J and COD is given by eq. (18) for small-scale yielding:

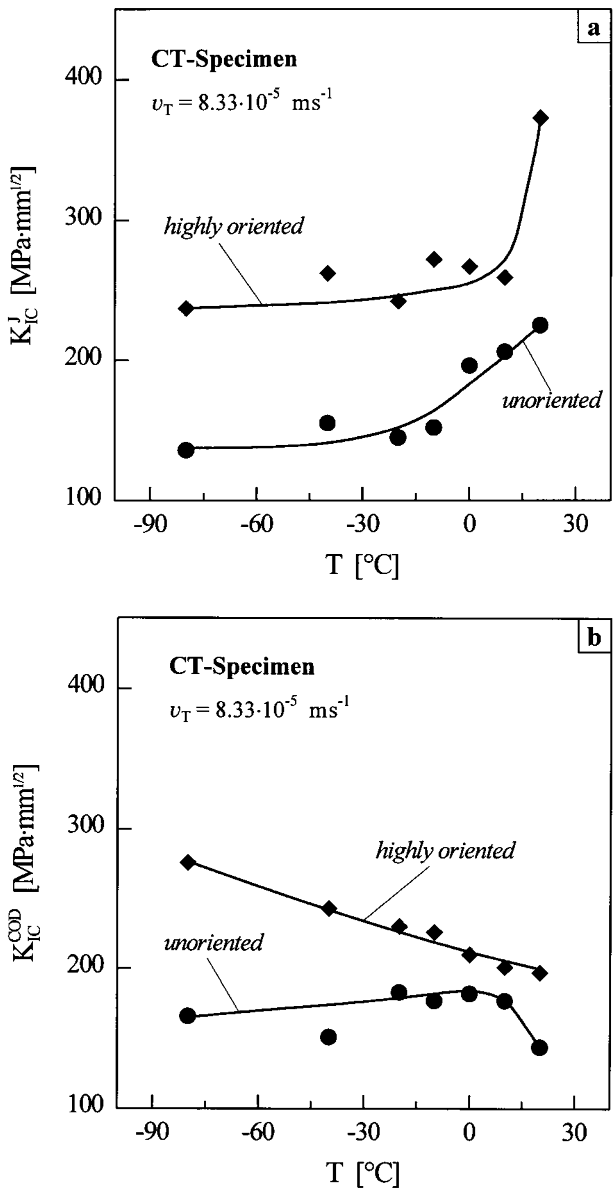


Figure 8 Temperature dependencies of the critical stress intensity factors K_{IC} under static loading: (a) K_{IC}^J deduced from the J_{IC} values; (b) K_{IC}^{COD} deduced from the δ_{IC} values.

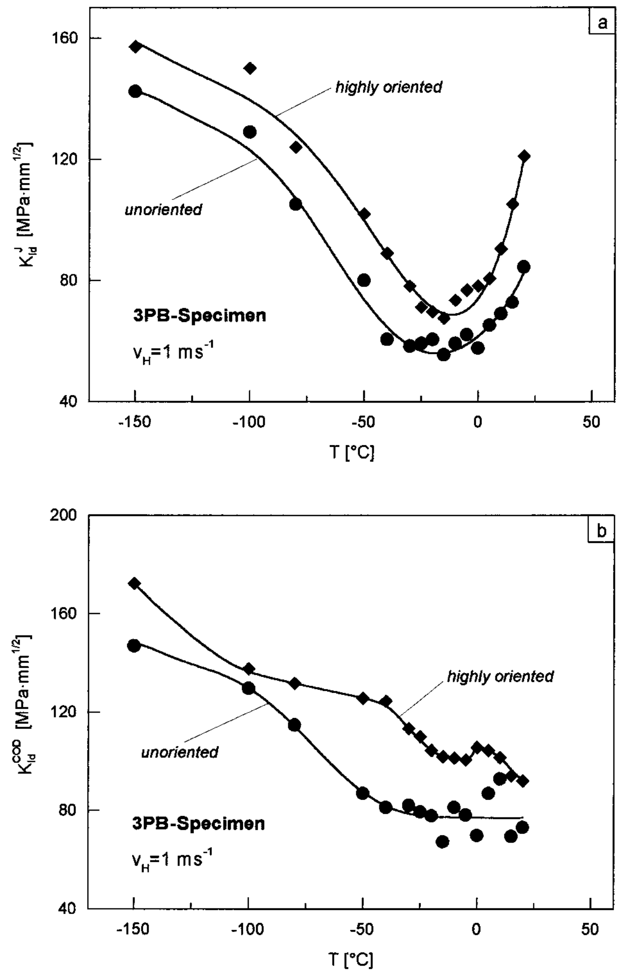


Figure 9 Temperature dependencies of the critical stress intensity factors K_{Id} under dynamic loading: (a) K_{Id}^J estimated from the J_{Id} values; (b) K_{Id}^{COD} estimated from the δ_{Id} values.

$$J = m\sigma_y\delta \qquad (18)$$

where m is the crack-tip constraint factor and depends on the stress state and the material properties. With eqs. (17) and (18), the equivalent K values can be estimated from the experimentally measured δ values:

$$K_{IC}^{\mathrm{COD}}; K_{Id}^{\mathrm{COD}} = \sqrt{m\sigma_y(\delta_{IC}; \delta_{Id})E^*} \qquad (19)$$

with $m = 0.7$ for the tested polypropylene.[30] The expressions δ_{IC} and δ_{Id} are the respective critical δ values under static and under dynamic loading.

Figure 8(a, b) presents the temperature dependencies of the K_{IC}^J and K_{IC}^{COD} values that were calculated through eqs. (17) and (19), respectively. Although the K_{IC}^{J} values increase with increasing temperature, but more slowly than the K_{IC}^{E} values, the K_{IC}^{COD} values decrease with increasing temperature. The reduction is especially significant in the highly oriented cold-rolled structure and is attributed to the temperature dependence of the Young's modulus and the yield stress, which affects the fracture toughness more dominantly than the enlargement of the plastic zone (a_{eff}) with increasing temperature.

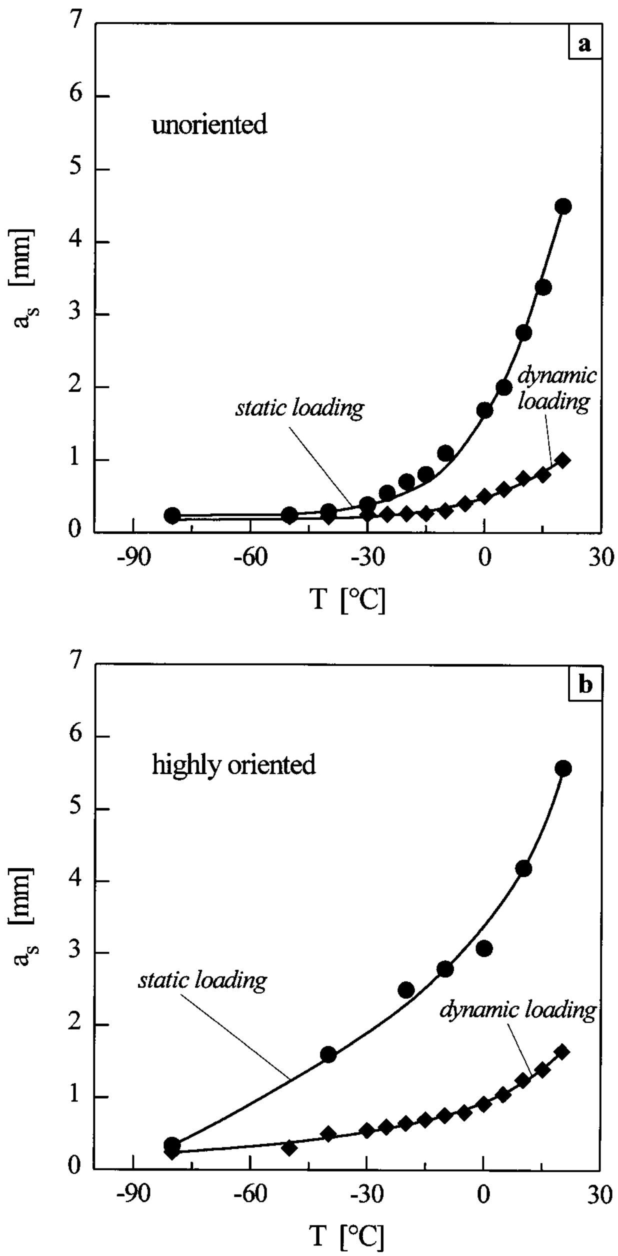


Figure 10 Comparisons between the maximum stable crack extensions under static loading and under dynamic loading in dependence on temperature in (a) the unoriented and (b) the highly oriented polypropylene.

In contrast to the results under static loading, under dynamic loading a negative temperature dependence of the fracture toughness in the whole temperature range investigated [Fig. 9(a,b)] occurs. Here the negative temperature effect on the Young's moduli, as observed from tensile vibration tests, may have played the decisive role.

Influence of Loading Rates on the Temperature Dependence of Toughness

The plastic deformations in the crack-tip zone can be microscopically quantified as the length of the stable crack growth on the fracture surface, which represents a measured quantity that can well describe the toughness behavior with increasing temperature. However, owing to their strong geometry dependencies, they can not be treated as fracture mechanics parameters.

Figure 10 shows that an increase of the loading rate from 8.33×10^{-5} to 1 m/s leads to a 3–5 factor reduction of the size of the plastic zone as a result of the increasing deformation restriction, both in the unoriented and in the highly oriented structure. All these findings represent the principal problems that were usually encountered when determining dynamic crack resistance curves.[48]

The influences of the loading rates on the J values in the unoriented and highly oriented polypropylene structures are presented in Figure 11. Increasing the loading rate from 8.33×10^{-5} to 1 m/s results in a 3–10 factor reduction of the J values in dependence on temperature. The rise of toughness with increasing temperature is especially significant under quasistatic loading, and the transition temperature locus can be related to the temperature locus of the β-relaxation.

Control of the Geometry Independence Conditions

The geometry independence conditions for the different fracture mechanics parameters under static loading as well as under dynamic loading were examined by using different size requirement criteria and are listed in Table II.

As stated above, the different fracture mechan-

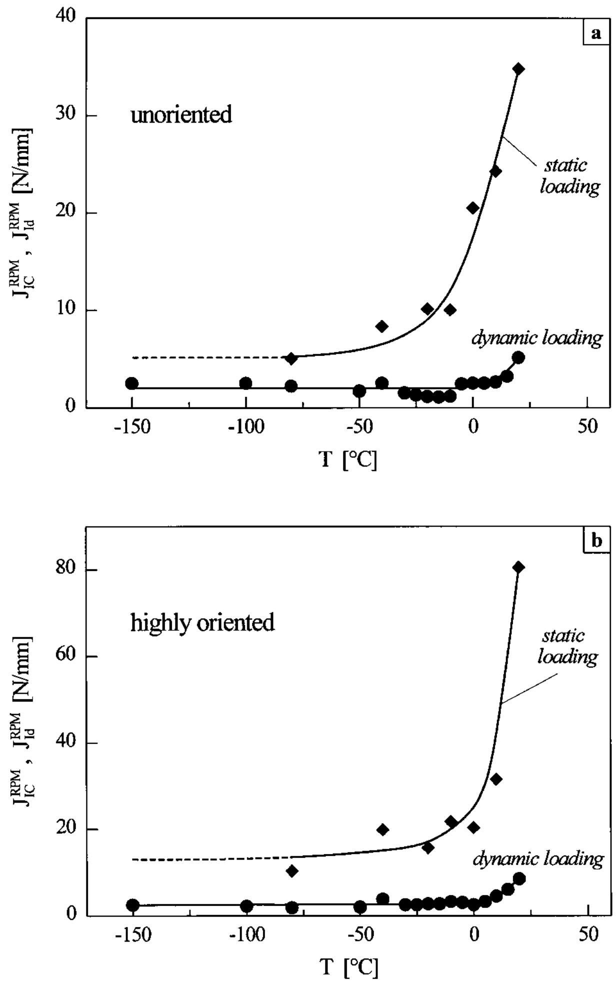


Figure 11 Effects of loading rates on the critical J integral in dependence on temperature in (a) the unoriented and (b) the highly oriented polypropylene.

ics parameters are only valid, when their corresponding size requirements are satisfied. If the specimen thickness requirements are checked to assess the temperature-dependent fracture toughness through eqs. (4), (5), (8), (12), and (15), it is evident that, with increasing fracture toughness and decreasing loading speeds, the range of validity is shifted to lower temperatures. Therefore, for tough polymeric materials at lower loading speeds and at higher temperatures greater values of specimen thickness are needed.

For the tested structures, by applying the J integral or COD concepts the geometry-independent fracture mechanics values could be determined in a wide temperature range on specimens with $B = 10$ mm under static loading and on specimens with $B = 4$ mm under dynamic loading. In contrast, the K concept, regardless of whether the effective crack growth and the pseudoelastic force in the equivalent energy concept have been taken into consideration, could only be applied with a great limitation under static as well as under dynamic loading.

By applying eqs. (13) and (16) with the material-specific constants ξ and ϵ, the temperature dependence of fracture toughness could be evaluated regardless of loading conditions and the nature of material failure. This enables one to make a material-relevant assessment of fracture behavior in the whole temperature range investigated.

CONCLUSIONS

1. Through examples of the temperature dependencies of fracture toughness values in unoriented and highly oriented polypropylenes under quasistatic loading and dynamic loading, the different statement capabilities of the various fracture mechanics parameters were presented.
2. The temperature dependencies of fracture toughness can only be assessed on the basis of geometry-independent fracture mechanics parameters in which the material-specific constants should be used in the size requirement criteria.
3. The expressions J_C and δ_C are the most appropriate fracture mechanics parameters to characterize the temperature-dependent fracture behavior over the wide temperature range investigated. These two parameters are most sensitive to the changes in orientation, texture, and temperature. By contrast, fracture toughness in terms of K, K^{COD}, and K^J can only be applied in a much more limited temperature range.
4. With the help of the geometry-independent fracture mechanics parameters J and δ, the material-relevant ductile–brittle transition temperatures can be determined. For the tested polypropylene, these temperatures can be related to their respective glass transition temperatures.

Table II Control of the Geometry Independence Conditions

		Static Loading		
	J_C	$\varepsilon = 25$		$\varepsilon = 224\ J^{-0.94}$
	Unoriented	$T \leq -8°C$		✓[a]
	Highly Oriented	$T \leq -15°C$		✓
	δ_C	$\xi = 20$	$\xi = 50$	$\xi = 3.6\ \delta^{-0.83}$
	Unoriented	$T \leq -2°C$	×[b]	✓
	Highly Oriented	$T \leq -40°C$	×	✓
	K	$\alpha = 2.5$	$\alpha = 2/\pi$	$\alpha = 3466\ K^{-1.73}$
K_C	Unoriented			
	LEFM	×	$T \leq 10°C$	✓
	Equivalent Energy	×	$T \leq -3°C$	✓
	Highly Oriented			
	LEFM	×	$T \leq -23°C$	✓
K_C^J	Equivalent Energy	×	$T \leq -23°C$	✓
	Unoriented			✓
	Highly Oriented			✓
K_C^{COD}	Unoriented			✓
	Highly Oriented			✓
		Dynamic Loading		
	J_d	$\varepsilon = 25$		$\varepsilon = 224\ J^{-0.94}$
	Unoriented	✓		✓
	Highly Oriented	✓		✓
	δ_d	$\xi = 50$		$\xi = 3.6\ \delta^{-0.83}$
	Unoriented	$T \leq 10°C$		✓
	Highly Oriented	$T \leq -22°C$		✓

[a] : valid
[b] : invalid

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