Dynamic fracture surface energy values and branching instabilities during rapid crack propagation in rubber toughened PMMA

Valeurs de l'énergie de rupture dynamique associées aux instabilités de branchement en fissuration rapide dans un polymère "choc"

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Abstract
An experimental device based on a strip band geometry was designed to explore the brittle behaviour of polymers during rapid crack propagation. The macroscopic crack speed is found to be quasi-constant along any entire RT-PMMA specimen, even in the case of crack branching and until arrest, if any. At the macroscopic branching velocity, the experimental fracture surface energy value and the fracture surface roughness are found to have no single values in RT-PMMA. Indeed, the macroscopic fracture surface energy value increases with the amount of instabilities or aborted cracks branching.

Résumé
Un dispositif expérimental basé sur la géométrie de l'éprouvette en bande a été conçu pour étudier le comportement fragile en fissuration rapide des polymères. Pour un PMMA "choc", on observe que la vitesse de la fissure est quasi-constante tout au long de l'essai, qu'il y ait ou non branchement ou arrêt de fissure. A la vitesse de branchement macroscopique, la valeur mesurée de l'énergie de rupture ainsi que la rugosité de la surface de rupture ne sont pas uniques dans le PMMA "choc". En effet, l'énergie de rupture macroscopique croît avec la quantité d'instabilités ou "branchements avortés".
**INTRODUCTION**

Most polymeric materials exhibit a brittle fracture mode at high crack propagation speeds. The experimental system shown in Fig. 1(left) was designed to enable a steady state regime of brittle fracture and a simple mechanical analysis of the fracture energy, even for rapid crack propagations (RCP) [1, 2]. The geometry is based on the strip band specimen (SBS), which dimensions are typically $L \approx 200$ mm, $25 \text{ mm} < H < 45$ mm, $a_0 \approx 3$ H, $B = 2$ mm. The location of the crack tip during the propagation is determined by measuring the resistance of a metallic layer [3]. The loading device, shown in Fig. 1(right), ensures uniform and constant displacement of the strip band boundaries. The symmetry of the loading is verified by strain measurements on the sides of the specimen. Crack propagation is started by an impact of low energy on a razor blade placed in contact with one initial blunt notch and a crack propagates symmetrically at a macroscopic speed $a_m$. Only one of the twin specimens undergoes fracture. Owing to the weight of the grips and the short time to fracture, typically 200 $\mu$s, we assume that the boundary conditions are fixed during the crack propagation. Crack branching can be obtained by increasing the mechanical potential energy stored in the specimen. The crack branches are generally symmetrical for specimens made of polymethylmethacrylate (PMMA) and rubber toughened-polymethylmethacrylate (RT-PMMA). RT-PMMA is a blend of spherical particles of rubber and PMMA. The volume fraction of particles is approximately 40 % and the diameter of the particles is 200 nm.

![Fig. 1. (left) Schematic representations of the strip band geometry uniformly loaded and of the conducting layer used to record the crack tip position during propagation. (right) Experimental device to ensure a symmetrical loading.](image)

*DYNAMIC ENERGY RELEASE RATE COMPUTATION AND FRACTURE SURFACE ENERGIES*
As shown in Fig. 2, the macroscopic crack speed were constant, for a given specimen at a given temperature, whatever branching occured or not. Some experiments revealed effectively that crack propagation in RT-PMMA is unstable between approximately 1 and 0.6 \( c_r \), where \( c_r \) is the Raleigh wave speed. As the fracture surface energy decreases with increasing crack speed, the speed of propagation jumps from 1 to about 550 m/s when an increasing loading initiates the propagation of the crack.

The crack tip position during propagation and the stress state at initiation being known, the dynamic energy release rate \( G_{ID} \) may then be calculated by means of a transient dynamic finite element procedure, based on the software Castem2000©. Owing to the high level of strain rate, a linear elastic behaviour of the fracture mechanics is assumed [4]. It has been shown experimentally that the impact on the razor blade influences the crack propagation only on a few millimetres after initiation. Nevertheless, this crack initiation is simulated by imposing an initial crack tip opening, corresponding to the action of the razor blade at the crack lips. The Fig. 3 shows that the dynamic correction factor is generally of the order of 20 to 30 % for macroscopic crack speeds of about 0.6 \( c_r \). Indeed, this specimen geometry is known to induce a low dynamic correction factor [2, 5] and the remote stress field at the crack tip is actually not strongly influenced by the inertial effects in this range of crack speeds. In order to simplify the results, since \( G_{ID} \) displays relatively small oscillations during crack propagation, the mean value of the fracture energy, \( < G_{ID} > \), was calculated for each specimen before branching, if any. Practically, \( < G_{ID} > \) concerns the steady state regime and does not take into account the first 1.5 mm of propagation after initiation in RT-PMMA. In Fig. 4, it is clear that the experimental mean dynamic fracture surface energy values \( < G_{ID} > \) for a crack propagating at several hundred meters per second is substantially lower than the fracture energy \( G_{IC} \) at the onset of propagation, which is typically close to 10 kJ/m² at low medium stress intensity loading rates. All speeds, in the range 550-610 m/s for temperatures between 19 °C and 27 °C, are not
correlated to the values of \( <G_D> \). Crack branching occurs at a crack tip speed of nearly 0.6 times \( c_r \).

**Fig. 4.** Mean dynamic fracture energy values before branching (if any branching) vs macroscopic crack speed in a RT-PMMA. The numbers are temperature in °C.

Unlike in many other polymers like for instance PMMA, \( a_m \) does not change after branching during RCP in RT-PMMA. In RT-PMMA, the macroscopic fracture surface energy decreases with increasing crack speed and at nearly 0.6 \( c_r \), inertial effects modify the crack tip stress field to induce branching [6]. Subsequently, as the material behaviour tends to accelerate whereas the mechanical - inertial - effects tend to limit the rate of crack propagation, the crack speed stabilizes at approximately \( a_{mb} = 0.6 c_r \), which is the macroscopic crack branching speed for RT-PMMA.

**FRACTURE SURFACE ROUGHNESSES**

**Photo 1.** Optical micrographs showing the roughness change at branching - propagation from left to right, in a RT-PMMA specimen of thickness 2 mm – (left) 1mm before the crack branching, (right) 1mm after the crack branching.

The fracture surface displays a sharp change in roughness at branching, visible in the comparison of Photo 1(left) and (right). Optical microscopy reveals a relatively coarse surface.
texture a few millimeters before branching and a finer texture a few millimeters after branching. Fig. 5 (left) shows an atomic force microscopy image of a fracture surface a few millimeters before a macroscopic crack branching, corresponding to a high value of the fracture surface energy and Fig. 5 (right) just before a crack arrest, corresponding to the lowest level of the fracture surface energy. It is noticeable that also at the microscopic scale the surface are rougher prior to crack branching than before arrest at a similar crack speed. The fracture surface does not pass through the rubber particles of RT-PMMA.

Since the crack forms branches even though the experimental crack speed and the energy released inside the specimen remain quasi-constant, two or sometimes three branches must consume the same energy as a single crack propagating at the same speed. The varying parameter at crack branching is then the roughness of the fracture surface, which means the total surface created ($S_F$). Indeed, the usual planar crack surface (crack length increase times width, $B \Delta a$) should not be considered. The total surface created may be the approximately the same for two smooth cracks or one rough crack. In the case of a smooth crack, like in Fig. 5 (right), created at speed $a_{mb}$ the fracture energy is $G_{IDmin}$. If the energy release rate is greater than $G_{IDmin}$ the propagating crack may produce microscopic instabilities, since intertial effects are sufficient to render possible crack branching. These small instabilities are themselves smooth and short aborted branches. If the energy release rate exceeds twice $G_{IDmin}$, then macroscopic smooth branches can appear. SBS fracture tests were also performed using pure PMMA specimens. In these samples, optical microscopy revealed mirror like fracture surfaces, while as expected [7, 8] the measured fracture energy increased with crack speed. Fig. 6 shows that $G_{IDmin}$ is similar for RT-PMMA and pure PMMA at the corresponding crack speed. The fracture energy of RT-PMMA at a crack speed of $a_{mb}$ varies from 1 to 3 times $G_{IDmin}$.
–sometimes we have observed triple branching - and the latter corresponds to smooth crack surfaces. Hence the fracture energy might be expected to be directly correlated with the ratio $S_{fr} / (B \Delta a)$ and the fracture energy of pure PMMA.

$$\frac{S_{fr}}{B \Delta a}$$

Fig. 6. Fracture surface energy vs. normalized crack speed for PMMA and RT-PMMA.

**CONCLUSION**

RT-PMMA represents an interesting model material to investigate the micro mechanisms of dynamic fracture processes. It has been experimentally shown that and explained why, at the macroscopic branching velocity, the experimental fracture surface energy have no single value in materials, at least such that the fracture energy decreases in the RCP regime.

**REFERENCES**

6 Yoffé, E. H., Phil. Mag. 12, (1951), p 739-750.