Crack Propagation in Glasses and Polymers

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It is generally accepted that a limiting crack velocity exists when brittle materials fracture. This was first observed for glasses by Schardin & Struth[1]. Subsequent work by Schardin and his co-workers[2] showed that the limiting velocity for a particular glass varied with the type of glass. Various theoretical treatments now exist which predict that the Rayleigh wave velocity is the limiting velocity for crack propagation in a brittle solid, but, in general, measured values fall below this limit (for reviews see, for example (3 - 4)). The variation of maximum velocity V with temperature T was investigated by Dimmick & McCormack[5] for glass, for the temperature range 70 to 450 K, and a slight variation was observed, given by $\frac{1}{V}\frac{dV}{dT} \sim -10^{-4}K^{-1}$. Recent work by Congelton & Petch[6] has indicated that the maximum velocity decreases with increasing fracture stress.

There is clearly uncertainty as to whether the maximum fracture velocity is constant for a given material and temperature. Results are presented below which show the behaviour of the maximum velocity in a soda-lime silicate glass for a wide range of fracture stress and for temperatures up to $\sim$ 800 K. In experiments described in the final section of the paper crack propagation in PMMA (polymethylmethacrylate) and PS (polystyrene) was studied and dynamic fracture surface energies and crack tip temperatures measured.

Maximum crack velocity and fracture stress

The glass specimens had dimensions 20 x 2.5 x 0.6 cm and were stressed to fracture in a Hounsfield tensometer. Fracture stress was varied by changing the dimensions of the added notch or pre-crack. The velocity measurements were made using the ultrasonic technique developed

by Kerkhof operating at a frequency of 5×10^6 Hz (for details of our apparatus see (4)). Fig. 1 shows a comparison between the present results and those of Congelton & Petch. There is obviously a large discrepancy between the two since no definite trend in the fracture velocity with stress was found in the present work. It is worth noting, however, various differences between the two investigations. Congleton & Petch measured the length to bifurcation on their specimens and found velocity using a 'universal' curve built up from measurements on other specimens. A problem with the Congelton & Petch work appears to be that if a crack grows from a blunt notch rather than a Griffith-type crack it will have excess energy and accelerate more quickly. Only by measuring the maximum velocity reached by each fracture (as in our experiments) is this difficulty overcome. Secondly, they measured velocities using Wallner lines and this is a more difficult technique to obtain accurate velocity measurements with than the Kerkhof method. Further, the way these authors measured velocities from 'secondary' Wallner lines (figure 2 in their paper) appears to be wrong since they ignore the fact that transverse waves must reflect with an angle equal to the incident angle. This error would vary from specimen to specimen but could be as large as 30%. Finally, whereas we simply varied initial crack or notch dimensions, Congelton & Petch changed fracture stress by variation of crack tip radius in the range 10^{-2} - 10^{-3} cm and by variation of the temperature and environment (see fig. 1).

Maximum crack velocity and temperature

The batch of glass was different to that used in the earlier experiments, which explains the small difference in maximum velocity at room temperature: specimen dimensions were 20 x 5 x 0.6 mm. The precnack or notch from which the fracture developed was altered as the temperature varied in order to maintain the same fracture stress of

$\sim 7 \times 10^6$ N/m^2. At the loading rate used ($\sim 3.5 \times 10^5$ N/m^2) there was an increase of fracture strength by a factor of about 10 as the temperature was lowered from 290 to 110 K.

The results are plotted in fig. 2. A straight line gradient indicates a variation given by $\frac{1}{V}\frac{dV}{dT} \sim -1.1 \times 10^{-4} K^{-1}$, though $\frac{dV}{dT}$ appears to increase in magnitude as the temperature is increased. The fracture velocity varies by only 100 m/s in the temperature range 0.1 to 0.9 T_s, where T_s is the softening point (920 K). This observation may be contrasted with the change in fracture strength of $\sim$ 100 times over the same temperature range. An interesting point was that the appearance of the fracture surfaces of specimens broken at high temperature was very similar to that observed at room temperature with regions of 'mist' and 'hackle' preceding bifurcation.

If it is assumed that the maximum fracture velocity is a constant fraction of the stress wave velocity, C, for a given material then any variation of C with temperature should affect the crack velocity. Calculations show that $\frac{1}{C}\frac{dC}{dT}$ for soda-lime glass is $\sim -10^{-4} K^{-1}$. It appears then that the measured variation in fracture velocity may be completely attributed to variations in stress wave velocity. Some workers have suggested that the limiting crack velocity in glass is less than the predicted Rayleigh wave velocity because of viscous effects at the crack tip. However, viscosity is a very sensitive function of temperature and bigger changes in the limiting velocity with temperature would be expected if viscous dissipation was significant during crack propagation in glasses.

Crack growth in polymers

The tip of a fast crack propagating through a glassy polymer is surrounded by a small zone of plastically deformed material. The energy absorbed in plastic flow dissipates as heat. Since the evolution is

virtually adiabatic, a substantial temperature rise is possible within the deformed zone. The quantity of heat evolved, Q_E, can be determined by measuring the temperature rise produced at a known distance from the crack as the heat diffuses into the specimen. Previously the rise was monitored with thermocouples [7]; in the present experiments a temperature sensitive (liquid crystal) film was tried, as this technique provided a record over the entire crack path. The plate specimens were notched and broken in tension; the reference parameter chosen was the crack velocity. The Q_E values obtained for PMMA are shown in fig. 3.

The rise observed in the above experiments was ∿ 1 K as the point of measurement was a relatively large distance (0.1 cm) from the crack plane. The increase in the plastic zone itself was determined with an infrared detector (Mullard RPY 36) which recorded the radiation emitted from the crack faces as they separated after fracture. With the detector's field of view limited by a slit perpendicular to the crack direction the signal shown in fig. 4 was obtained for PMMA. The initial fall as the crack approached the slit is due to the thermoelastic cooling associated with the large elastic strain ahead of the tip. The infrared detector results, combined with the values of Q_E and the velocity at which the crack faces separate, enable the temperature rise, at a certain time after fracture, to be calculated; the rise is a function of time as heat conduction, even on the time scale of this experiment, is significant. For PMMA, the rise after 10 μs was ∿450 K: there was no significant variation with crack speed; for PS the rise was 300-400 K after the same interval.

The presence of a large temperature increase is supported by the observation of thermal decomposition products during the fracture of PMMA [8]. It could explain the occurrence of highly ductile deformation (for instance the long fibrils seen on PS fracture surfaces [9]) at the

high strain rates involved in fast fracture, and the increase in fracture surface energy with crack velocity found with many polymers.

Acknowledgments: to S.R.C. and Min. of Def. (Aviation Supply) for support.

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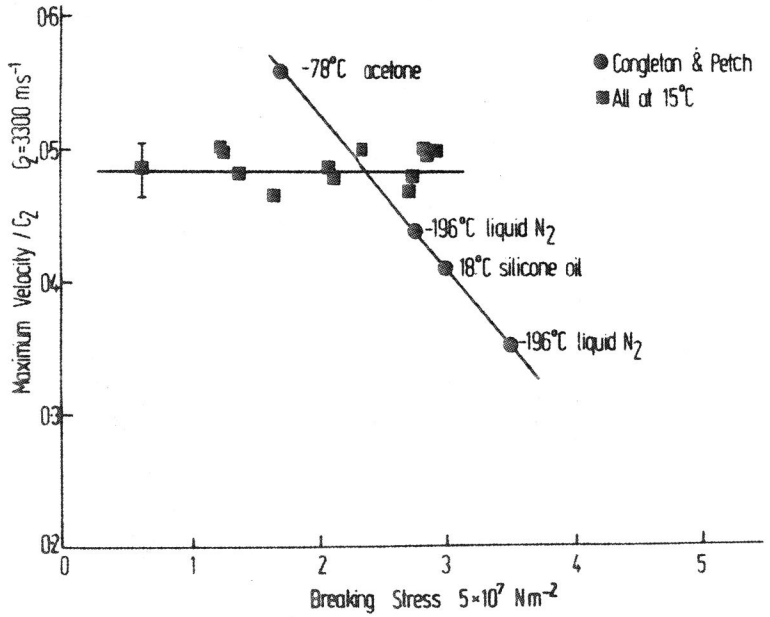

Fig. 1

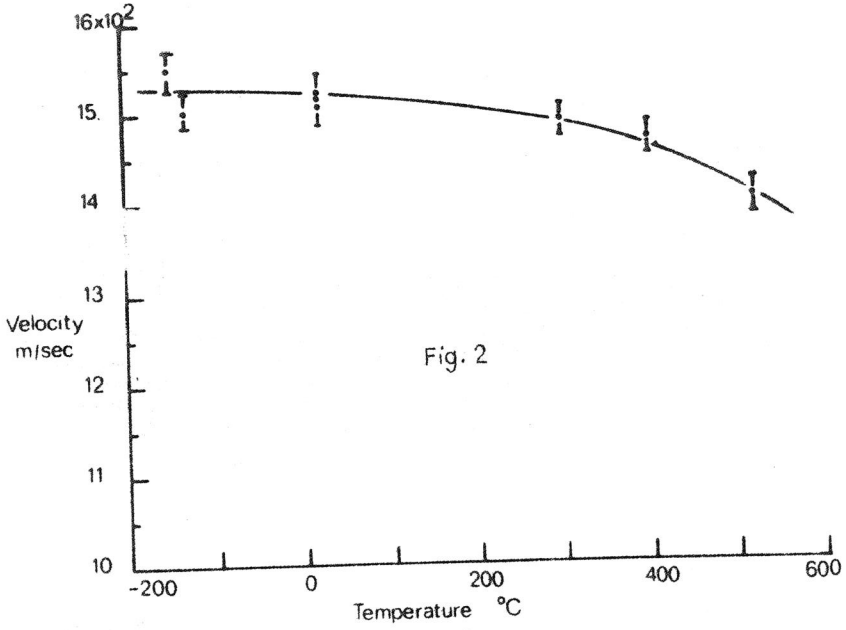

Fig. 2

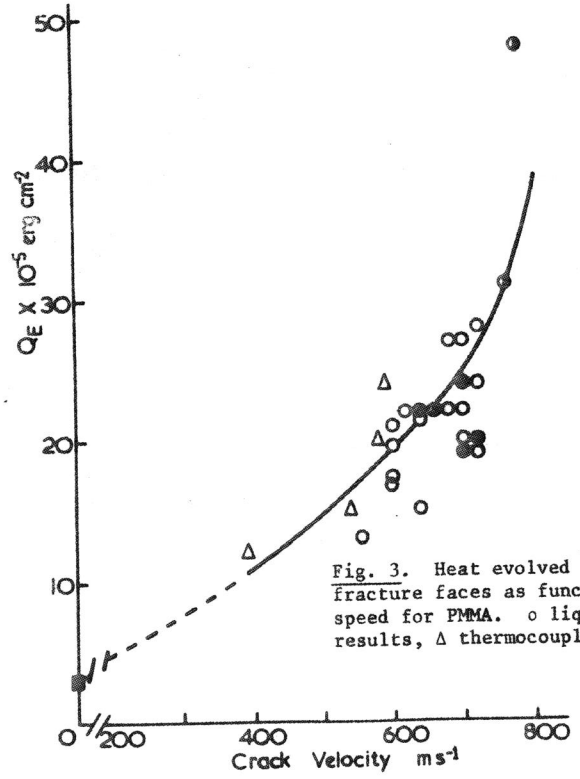

Fig. 3. Heat evolved on both fracture faces as function of crack speed for PMMA. o liquid crystal results, Δ thermocouple results.

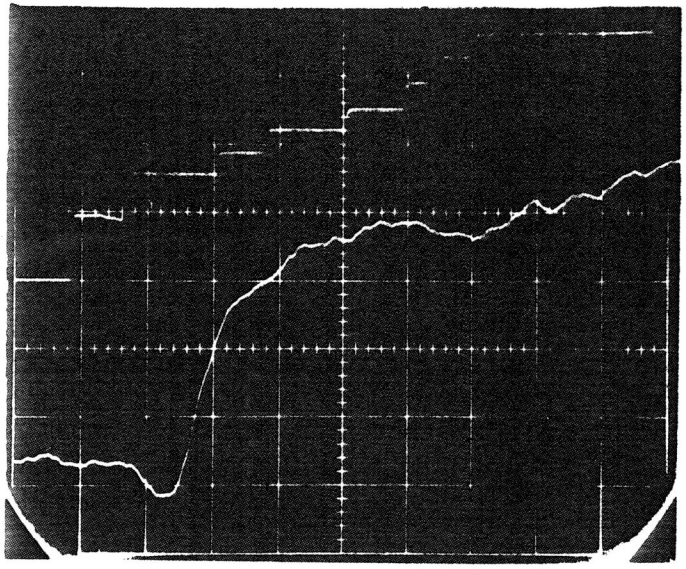

Fig. 4. Record from infrared detector experiment with PMMA. Upper beam: crack velocity measurement. Lower beam: radiation signal (+ve upwards). Horizontal scale 10 μs cm^{-1}.