

Figure 3 (a) Part of the perfect spinel unit cell projected onto the (111) plane [(0001) plane of sodium beta alumina] (after Hornstra [5]). The oxygen ions are not shown. The roman numerals give the relative heights of the tetrahedral interstices in multiples of $1/24 a_s \sqrt{3}$ (a_s is the lattice parameter of "spinel block" in sodium beta alumina). The Arabic numerals refer to the position of the octahedral interstices. The equivalence of R_1 and R_2 is shown. (b) Fault produced in the sodium conduction planes at the intersection of the fault boundaries. Note again the equivalence of R_1 and R_2 . O = position of oxygen ions in the conduction plane; x = possible position of sodium ions. The oxygen ions in the adjacent planes are on the grid intersections.

would be partially attributable to the presence of these faults.

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Fracture processes in polymethylmethacrylate

A recent paper [1] raises a number of points on the topic of fracture in PMMA that have been of considerable interest recently. Marshall *et al.*, (e.g. [2]) and the group at G.E. Schenectady, (e.g. [3–5]), have studied the phenomenon in considerable detail, and have emphasized the significance of the interrelation between crazing and fracture in glassy thermoplastics, (e.g. PMMA), albeit mostly at relatively slow crack speeds.

In particular, the effect of the molecular weight of PMMA on its fracture behaviour, as studied in [1], was considered in [6]. These data include

measurements of the increase of fracture surface energy (γ) with molecular weight in PMMA. A value of strain energy release rate derived from these measurements gives $G = 0.266 \text{ kJ m}^{-2}$ for a PMMA with a viscosity average molecular weight of approximately 190 000 at slow crack speeds. This is not inconsistent, at first sight, with the values of $G = 0.83 \pm 0.1$ and $1.43 \pm 0.79 \text{ kJ m}^{-2}$ quoted in [1] for a crack moving in a PMMA of molecular weight 163 000 at several hundred metres per second, since it has been demonstrated [7] that G increases significantly with crack speed in PMMA. However, if the data from [7] are compared with the previously published [8,9] data by the author

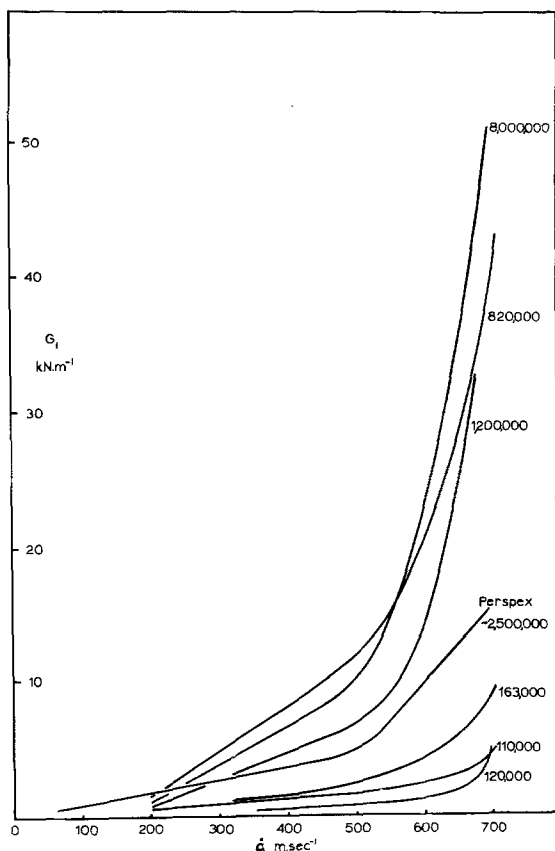


Figure 1 Photoelastic measurement of variation of G with fracture velocity for perspex, compared with data of Döll [8,9]. The Figures on the curves refer to molecular weights.

of [1], as in Fig. 1, significant inconsistencies are apparent. Two possible reasons for this inconsistency suggest themselves.

Firstly, the data in [1] quote weight average molecular weights, which may not be directly comparable with the data in [6] and those quoted for Perspex [10] in Fig. 1, which use viscosity average molecular weights. Further, the value of molecular weight of 490 000 quoted in [1] as a transition value for fracture behaviour is also derived from viscosity measurements [11] and may not be directly comparable with the data in [1].

The second reason is the point, correctly stated in [1], that a dynamic correction is necessary to results obtained from using the method of derivation of G used in [1]. It was shown in [7] that such corrected data are consistent with directly measured photoelastic data for G , and may thus be considered as a fair indication of the magnitude of G at the tip of a running crack. It was further shown that other data [12] using similar dy-

namically corrected measurements were consistent with the direct photoelastic measurements in [7]. Unfortunately, the molecular weight of the PMMA used in [12] is not quoted. Hence, it is suggested that the use of such a dynamic correction factor is necessary in [1] in addition to the use of a dynamic modulus, to adequately describe the conditions at a moving crack tip.

In [7], fracture surfaces were shown that were closely similar to those described in [1]. The regular markings were analysed on the assumption that they were a Wallner-type phenomenon, and this enabled a self-consistent description of the fracture process in PMMA to be made, relating crack speed, changes in modulus, changes in toughness, and transitions from smooth to rough fracture surfaces. The description of the formation of the parabolae on PMMA fracture surfaces in [5] as the intersection of a plane crack front with a radially expanding craze initiated ahead of the crack front was modified, with the parabolae occurring on different levels, as observed in [7], to produce subsequent surface roughness. It was further estimated that the energy absorbed in the crazing process was less than 18% of the total energy expended in forming fracture surfaces. The suggestion is made in [1], that for Wallner lines to be apparent in PMMA requires transverse waves of a magnitude that, due to damping, would seem unlikely. This, however, is not necessarily so, since the model described above provides a lower energy mechanism for the Wallner effect to make itself apparent, ahead of the crack tip. The subsequent passage of the crack front through these crazes produces the irregularly rough lines on the fracture surface, which are not seen in the more familiar type of Wallner-line modulated fracture surface, e.g. in glass, in which crazing does not occur. It is agreed that the origin of the source of the transverse waves producing this regular pattern is not immediately apparent, but then neither is it apparent in some of the strikingly regular Wallner-lines observed on glass fracture surfaces, e.g. [13]. It is suggested that further study of the formation of these markings would be useful in elucidation of the fracture process in PMMA, and the way in which they are influenced by molecular weight, as shown in [1] suggests that they are a fundamental material property, rather than an artefact of experimental techniques.

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