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# Impact properties of polyurethane and glass fibres reinforced composites

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Impact resistance of the polyurethane matrix and composites with glass fibres of various volume fractions was investigated. The theory of linear elastic fracture mechanics (LEFM) was successfully applied to obtain a quantitative assessment of a parameter of toughness, the critical strain energy release rate ($G_c$), which was determined from the energy ($W^*$) required to fracture sharply-notched specimens by taking into account specimen dimensions and notch depth. It is found that the $G^*$ is not a linear function of reinforcement concentration. The impact resistance with low volume fraction $\phi_f = 0.045$ of glass fibres decreases as compared to that of the matrix. However, with further incorporation of glass fibres, the impact resistance gradually increases, reaching its maximum for the volume fraction $\phi_f = 0.158$. An explanation of this non-linear behaviour is provided in this paper.

## 1. Introduction

For design engineers one of the most important properties of a material is toughness against impact. In order to study toughness under conditions of rapid deformation, various tests have been developed [1–4] where the specimens are subjected to rapid blows. The main drawbacks of these tests are that they provide information of overall impact energy and are strongly geometry-dependent. The measured impact energy contains various inseparable contributions, due to elastic deformation, visco-elastic process, plastic flow and deformation of the material at the tip of the propagating crack. Impact strength is often expressed as the specific fracture energy, which is in turn expressed by dividing the measured fracture energy by the cross-sectional area of the fracture ligament. This method is not entirely satisfactory, since fracture energy decreases with increase in $a/D$, the ratio of the notch length to the width of the sample [5, 6].

To overcome these difficulties and to obtain true material properties from the impact tests Williams and co-workers [7–11] have published a considerable amount of literature in which the fracture process in polymers is explained through a fracture mechanics approach, which takes into account notch length and specimen geometry. The theory of linear elastic fracture mechanics (LEFM), as applied to impact testing, presupposes that the material is homogeneous and fails in a brittle manner.

According to Marshall *et al.* [7] the theory involves the concept of a critical strain energy release ($G_c$) which is derived from the energy ($W^*$) required to fracture sharply-notched specimens by taking into account specimen dimensions and notch depth. The two terms are related [7] in Equation (1)

$$W^* = G_c B.D.\phi \quad (1)$$

where $B$ is the specimen thickness and $D$ the specimen width. $\phi$ is a geometrical correction factor and is expressed as:

$$\phi = \int \frac{Y^2(a/D)d(a/D) + S/18D}{Y^2 a/D} \quad (2)$$

A plot of $W^*$ against $B.D.\phi$ should give a straight line with slope $G_c$. $G_c$ is a constant for any given material and in this paper is used to compare the impact strength of different composites.

## 2. Impact testing

A fully instrumented machine, developed at the department of Polymer Science and Technology, UMIST, Manchester, UK, which is capable of being used for a wide range of polymeric and composite materials, as shown in Fig. 1, was used for impact testing.

The apparatus is a drop-weight device in which the striker may fall under gravity or be propelled at higher velocity by an impulse from a spring-loading system, at velocities ~4 to 13 m/sec. Information on the complete fracture process occurring during the impact event is gathered in the form of force/time data from a quartz force transducer situated on the striker attached to the cross-head. The output from the transducer is fed through a charge amplifier into a transient recorder which stores the information over the short duration required for the impact, usually < 5 m sec. This stored force-time information can be displayed continuously as a trace on an oscilloscope, and a permanent record of the impact may be produced by transferring the trace to chart recorder and/or by processing the information in a computer, which gives crude graphical representation of the data and has been programmed to compute essential fracture parameters.


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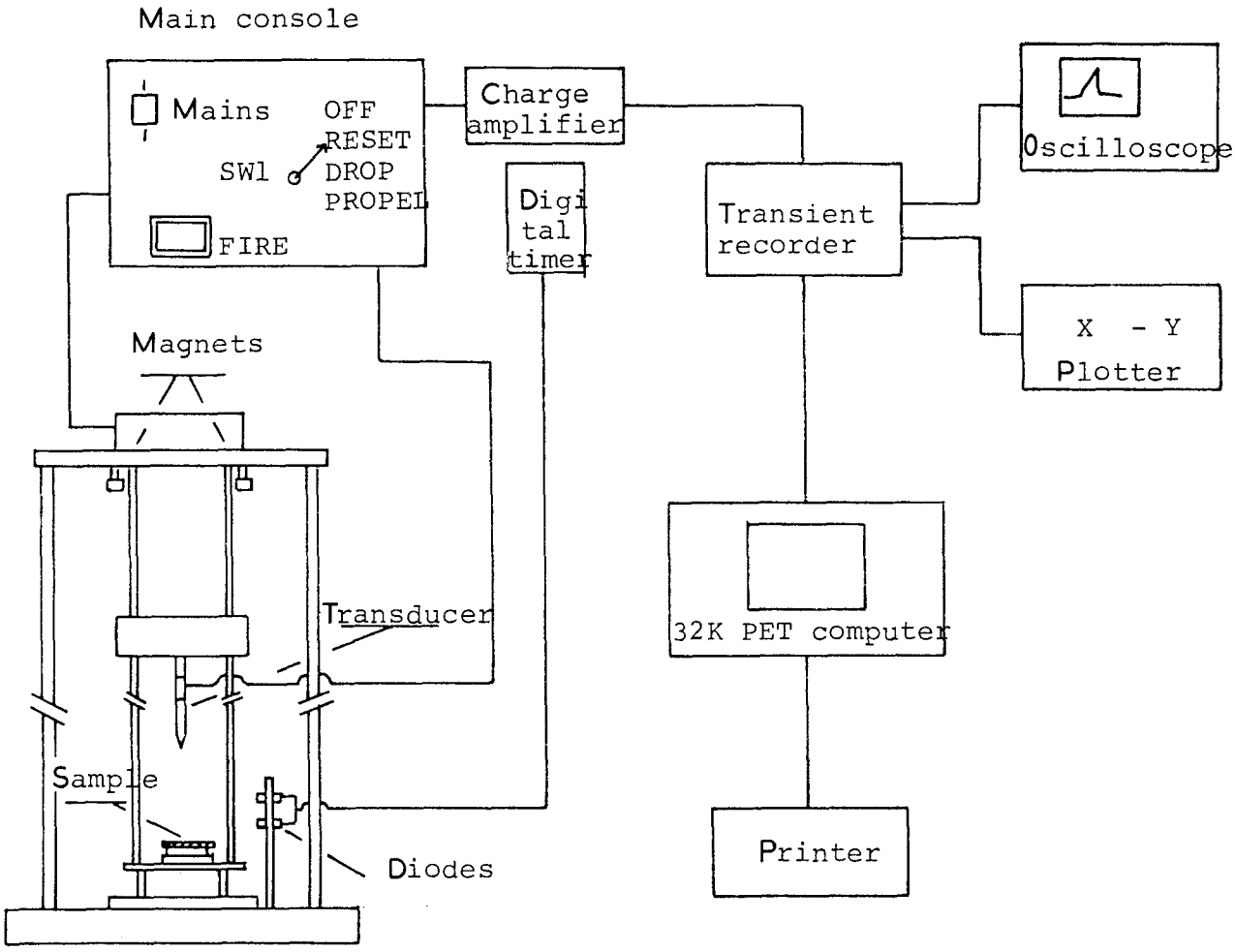


*Figure 1* Instrumented impact Tester-PET computer and electronic equipment for control of mechanical apparatus and data handling.

## 3. Experimental methods

The polyurethane matrix was prepared by reacting together an industrial tetrol UG1 3310 (supplied by Produits Chemiques Ugine Kuhlmann) and Suprasac VM-10 (supplied by ICI) at a constant temperature of 40° C. The mixture was thoroughly mixed and poured into rectangular metallic moulds subsequently cured at 150° C under vacuum. The required amounts of hammer-milled E-Glass fibres were introduced to tetrol contained in a three-neck flask in order to prepare composites having different volume fractions [12]. Rectangular beam specimens of polyurethane matrix with various volume fractions, ($\phi_f$ = 0.0, 0.045, 0.088, 0.126 and 0.158) were cut from moulded plates and notches of various lengths up to 40% of the width $D$, were machined. The highly viscous nature of the reactants and non-availability of RIM machine, limited the glass fibre reinforcement to a small $\phi_f$ of 0.158. $\phi_f$ can be increased by raising the working temperature and by the incorporation of glass fibres in both reactants separately which can be easily achieved on RIM where the whole mixing, reacting and moulding operations are completed in a few sec.

Before testing, a razor was introduced a small way

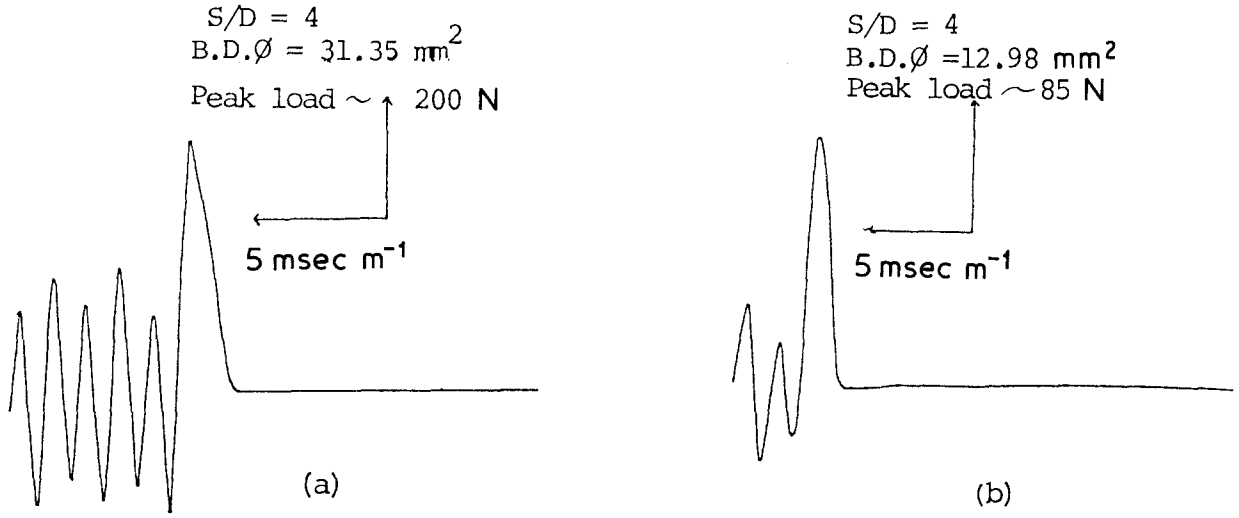


*Figure 2* Impact load-time curve for unfilled polyurethane with values of SD and $B.D.\phi$ as shown.

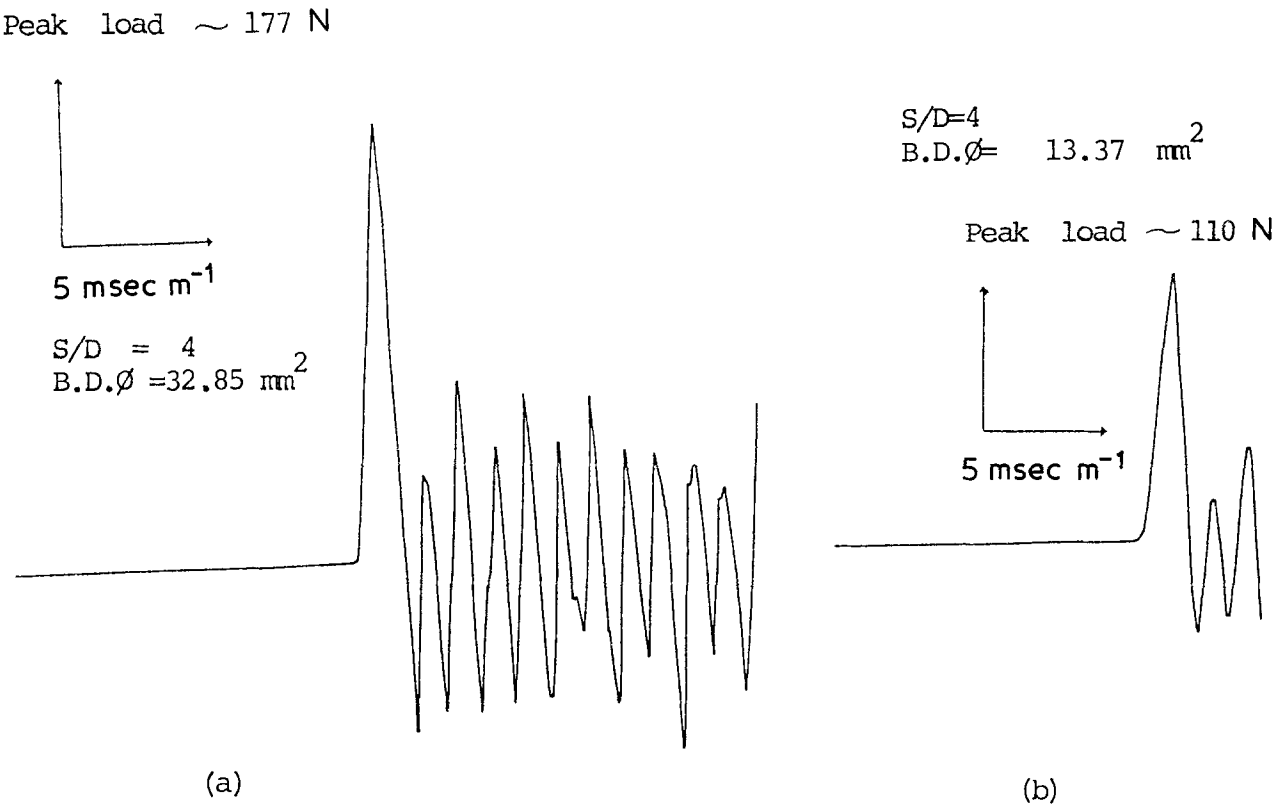


*Figure 3* Impact load-time curve for composite of $\phi_f = 0.045$ with value of SD and $B.D.\phi$ as shown.

into the tip of the notch of each sample; specimen depth ratio was kept at 4. A thin layer of synthetic putty (Blutak) was applied on the central portion of the specimen to avoid the superimposition of jagged peaks onto the main fracture curve which can occur because of striker oscillation and specimen vibrations. A thin layer of Blutak helps in reducing, to some extent, the effects of oscillation of striker and vibration of specimen during the event of impact fracture. The curves obtained from the specimens with a thin layer of Bluetak were found to be fairly reproducible. Some striker ringing still occurred after the fracture of the specimen, particularly with higher modulus materials, but this region of force deflection curves is unimportant, being easily detectable and rejected in the computer analysis. It has been observed that application of a thick layer of Blutak gives a smooth load time curve and almost all oscillation peaks can be eliminated, but the curve obtained would yield superficial values [14]. The usefulness of Blutak for obtaining reproducible and smooth fracture curves has been studied and documented by Stanford, Stepto and Taylor [13].

More than five specimens were tested at room

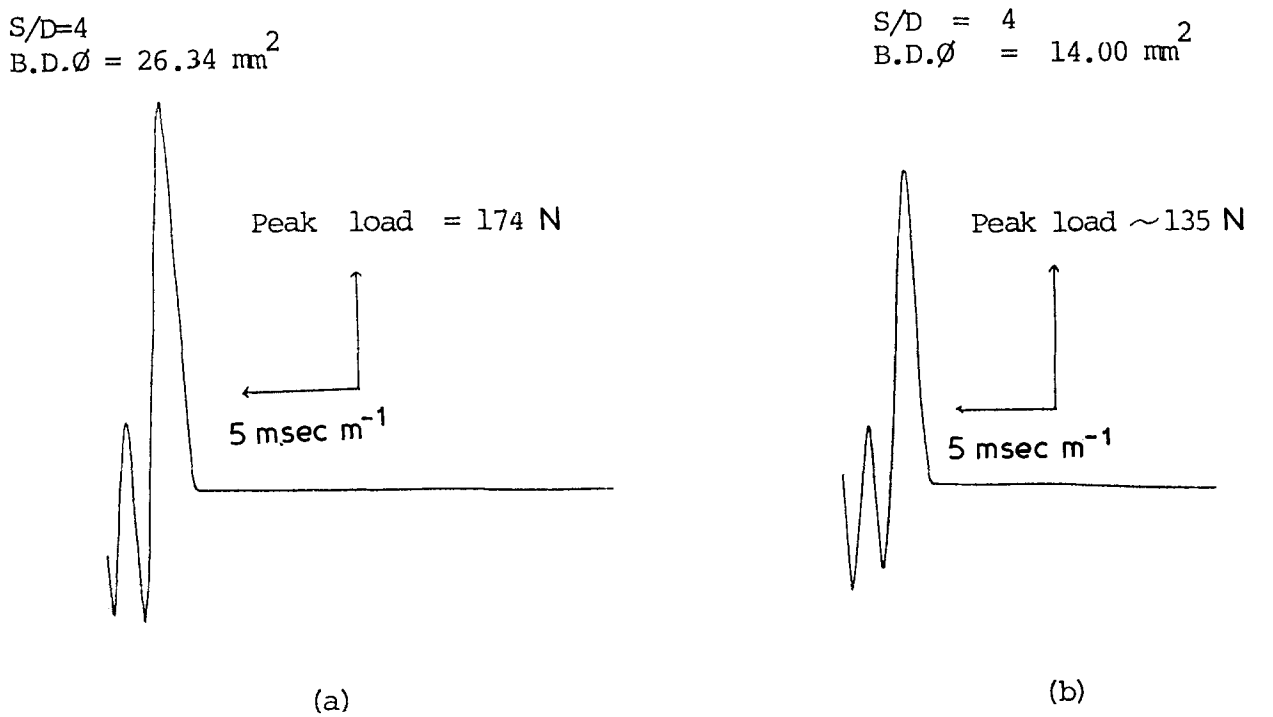


*Figure 4* Impact load-time curves of composite of $\phi_f = 0.088$ with values of SD and $B.D.\phi$ as shown.

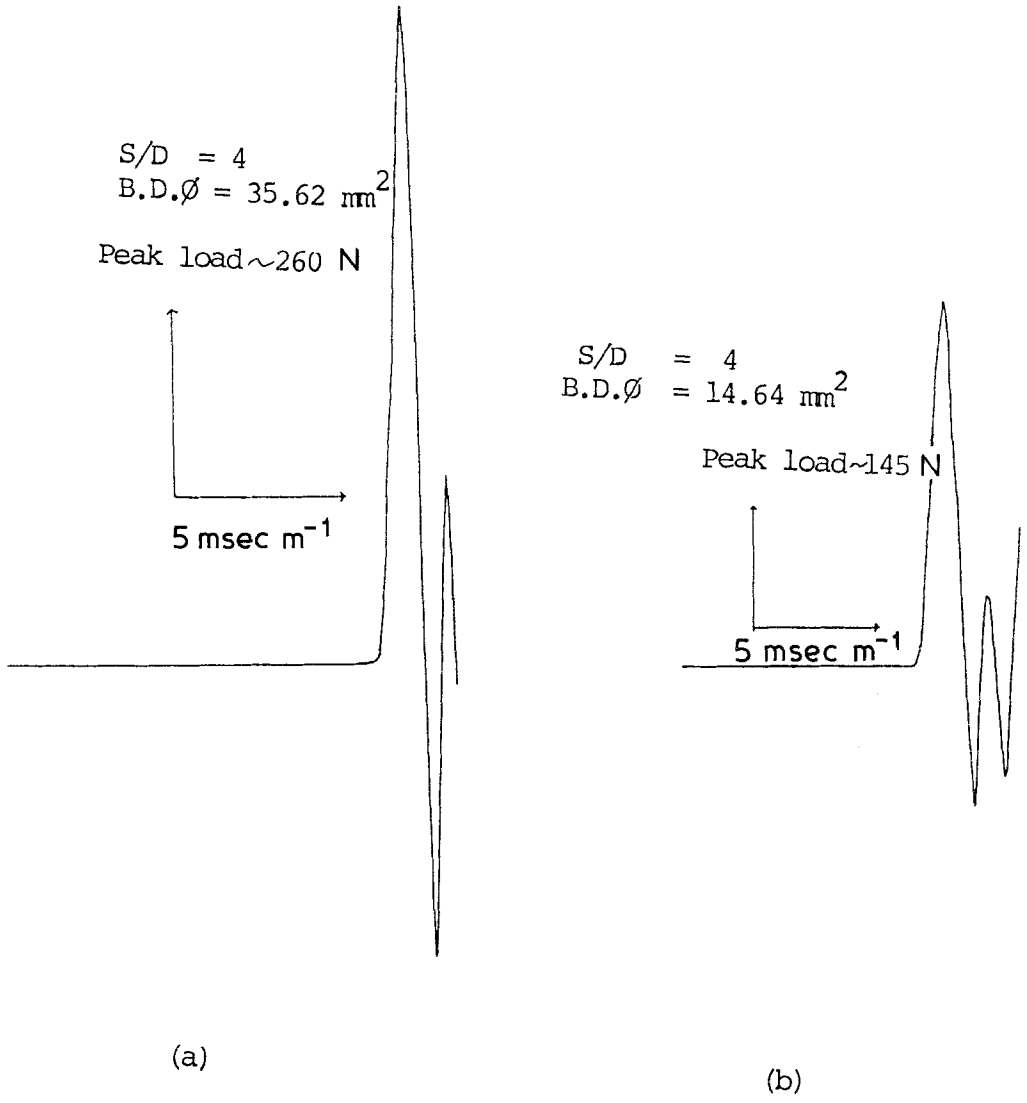


*Figure 5* Impact load-time curve for composite of $\phi_f = 0.126$ with values of SD and $B.D.\phi$ as shown.

temperature, 20° C, for each sample type and comprehensive force-time data covering the entire impact fracture process were obtained for each test by means of a transient recorder interfaced through to a computer. The force-time curve for each test is displayed simultaneously on an oscilloscope and a permanent record is made, through a transient recorder, on the chart recorder.

## 4. Results and discussion

The materials which were examined in this paper underwent brittle fracture, therefore it is possible to evaluate the critical strain energy release rate, $G_c$, as a material parameter which is independent of sample dimension and notch depth. $G_c$ is evaluated as a plot of the total fracture energy. $W^*$, the area under the force deflection curve, against the sample geometry factor $B.D.\phi$. The

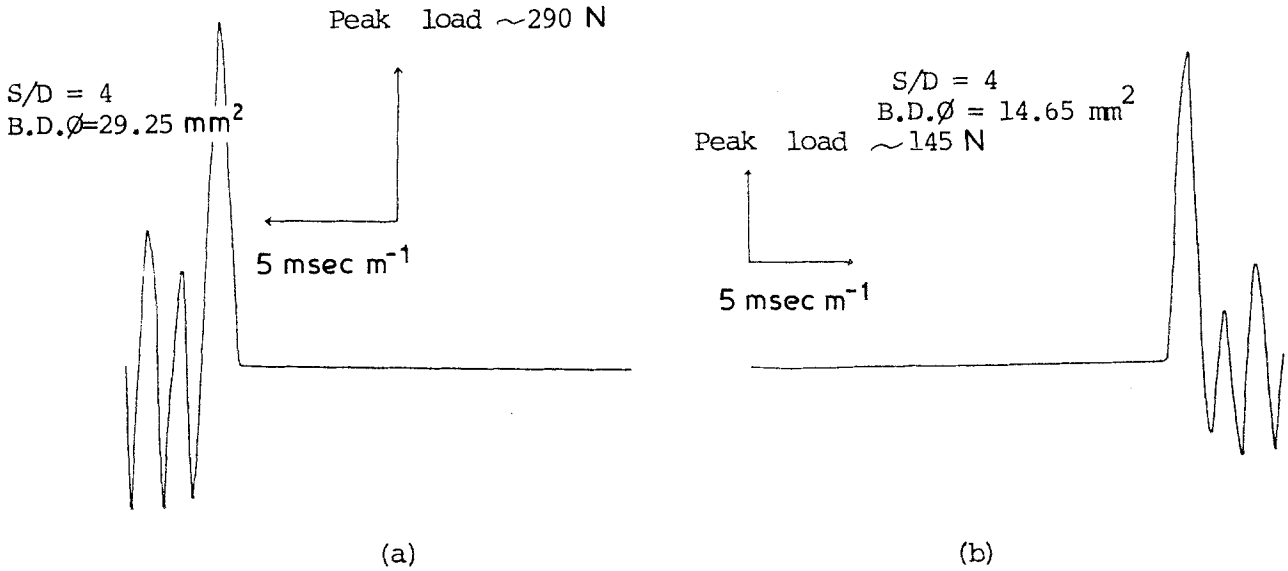


*Figure 6* Impact load-time curve for composite of $\phi_f = 0.158$ with values of SD and $B.D.\phi$ as shown.

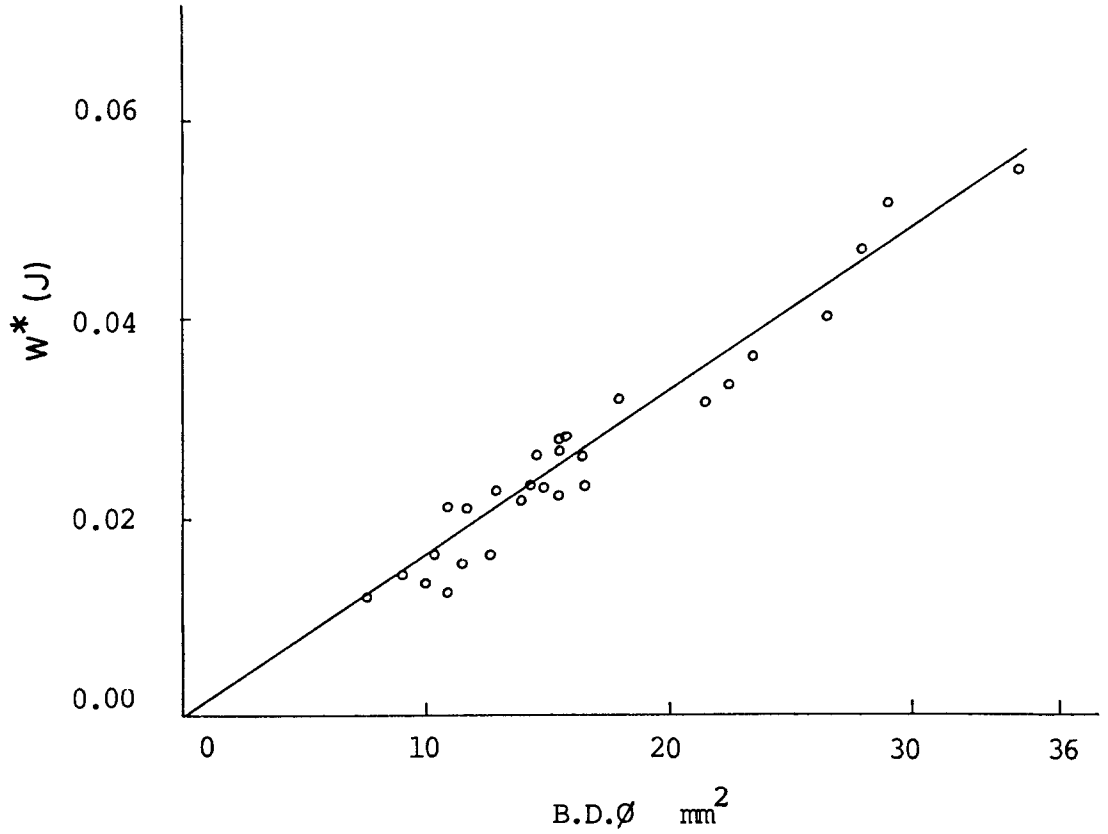


*Figure 7* Plot of $W^*$ against $B.D.\phi$ for specimen of polyurethane.

parameters $B$ and $D$ are the breadth and depth of the specimen and $\phi$ depends on $B$.D: sample length and notch depth. Thus, $W^*$ is evaluated for a series of samples notched to different depths, giving a series of values of $B.D.\phi$.

Figs 2 to 6 show the representative load-time curves for unfilled and reinforced polyurethane. It can be seen that the time to fracture varies with notch depth. Specimens with small notch depth require more time to fracture. Correspondingly, the area under the load-time curve, a measure of total energy required to fracture a specimen and the sample geometry factor $B.D.\phi$, also varies with the notch depth.

Figs 7 to 11 show plots of total energy, $W^*$, against

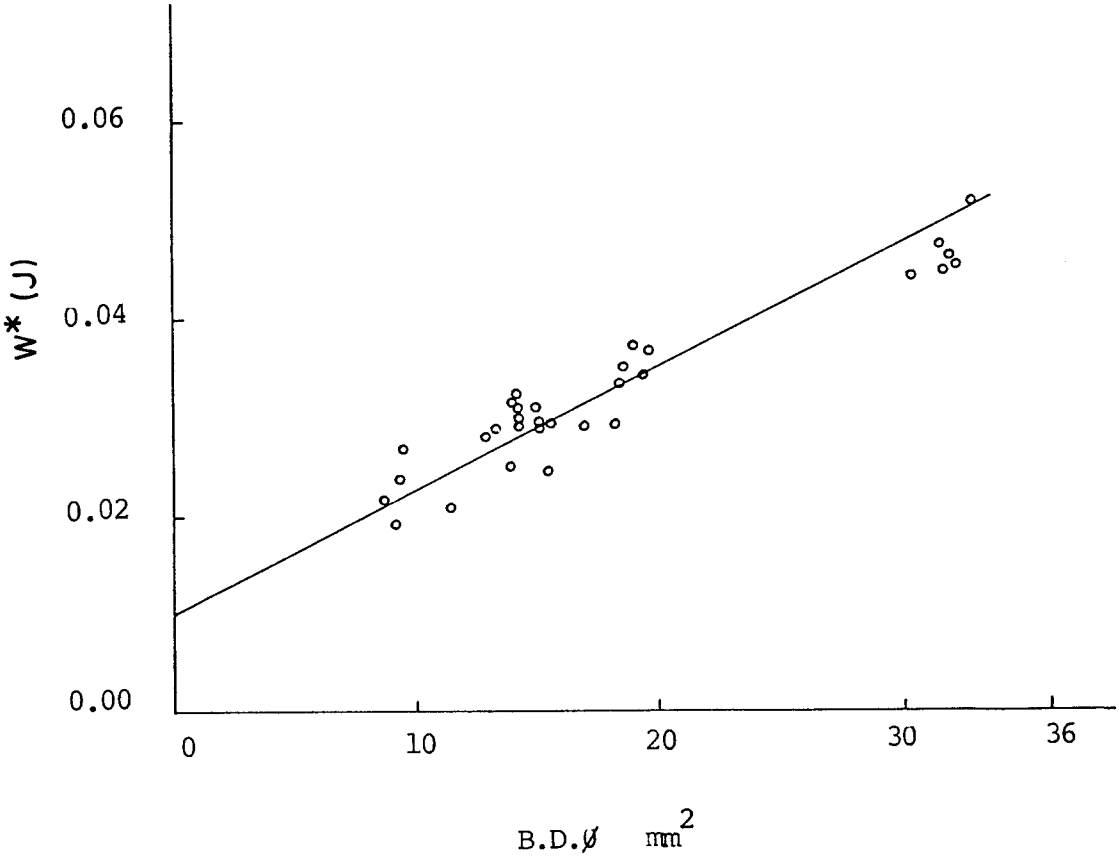


*Figure 8* Plot of $W^*$ against $B.D.\phi$ for reinforced specimens of $\phi_f = 0.045$.

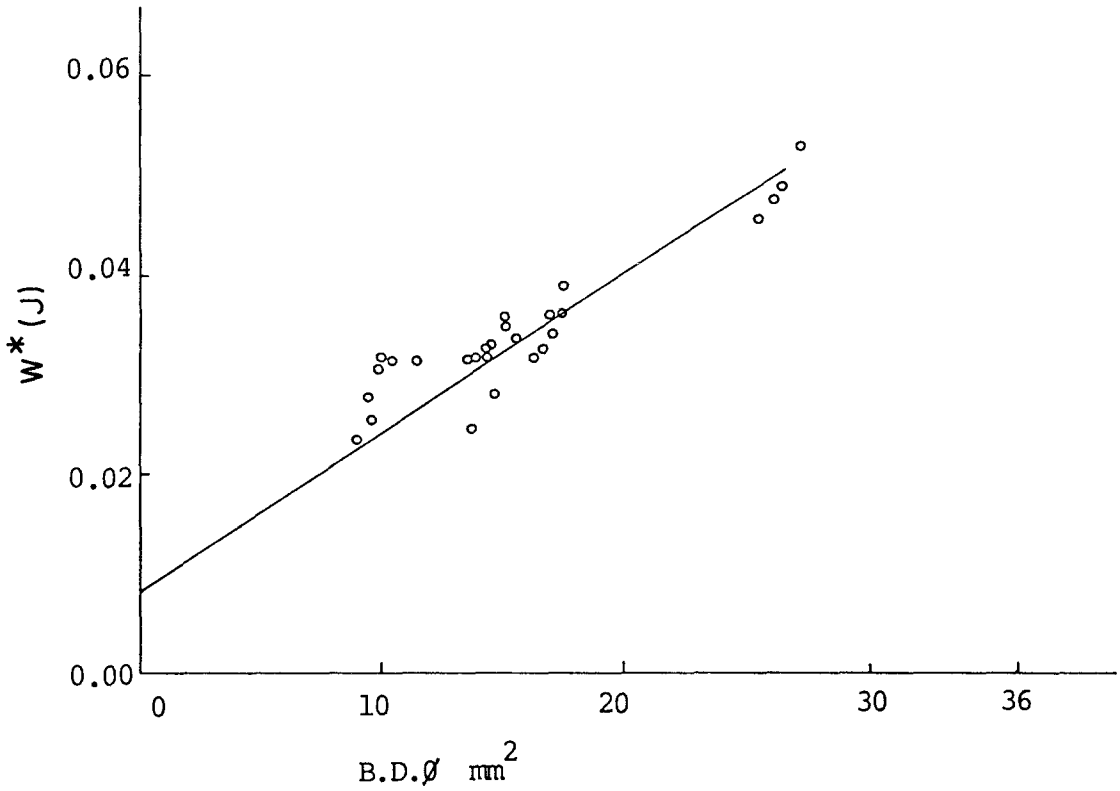


*Figure 9* Plot of $W^*$ against $B.D.\phi$ for reinforced specimens of $\phi_f = 0.088$.

$B.D.\phi$, for a polyurethane matrix and composites of all volume fraction series. According to theory, the plot of $W^*$, against $B.D.\phi$, should give a straight line passing through the origin. This has only been achieved in case of unfilled polyurethane matrix whilst the plots of $W^*$ against $B.D.\phi$. for all composites give a straight line with a positive intercept. This deviation from theory can be attributed to the loss in kinetic energy after the specimen is fractured. Every straight line shown in Figs 7 to 11 has a correlation factor

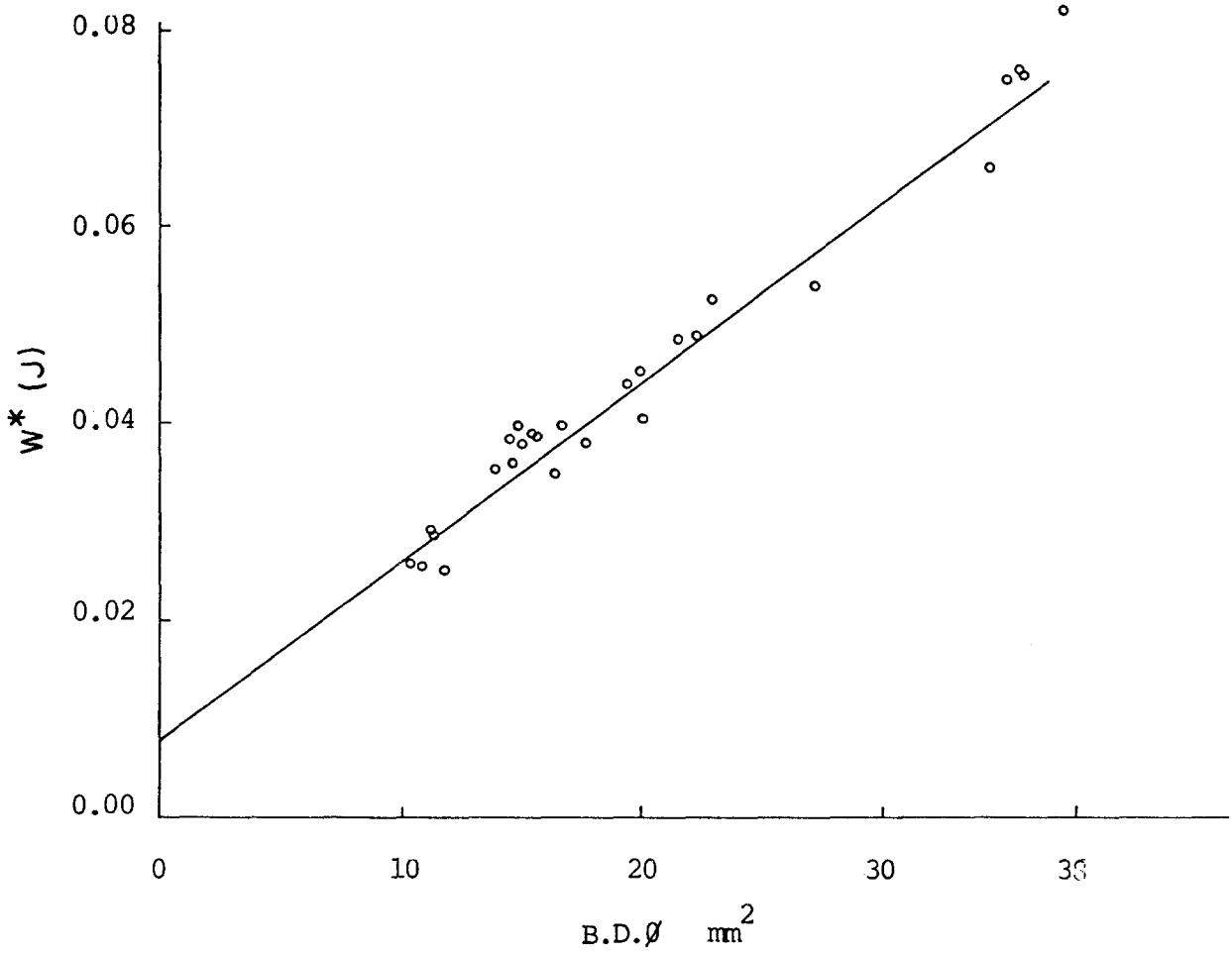


*Figure 10* Plot of $W^*$ against $B.D.\phi$ for reinforced specimens of $\phi_f = 0.126$.

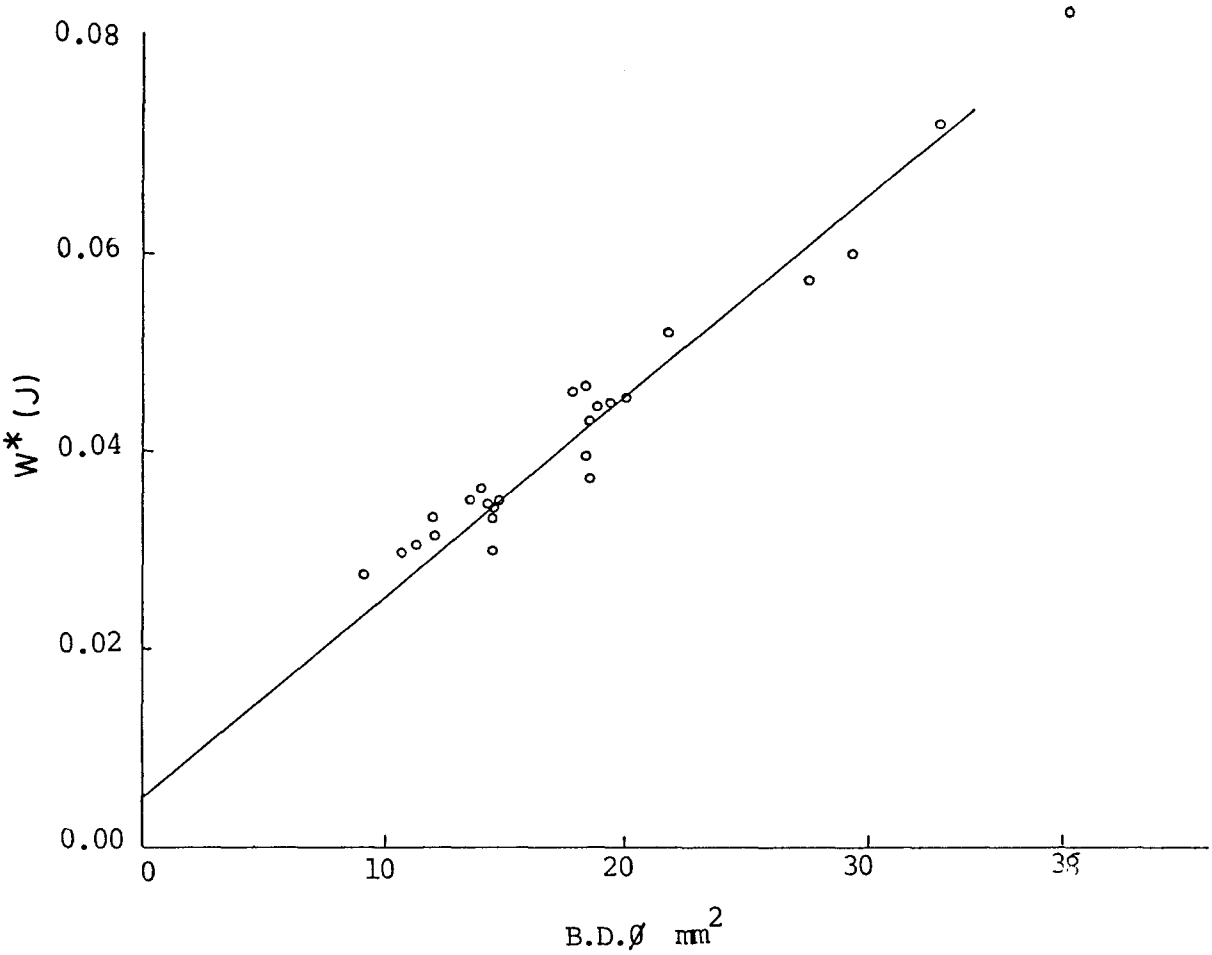


*Figure 11* Plot of $W^*$ against $B.D.\phi$ for reinforced specimens of $\phi_f = 0.158$.

$> 0.90$. The values of $G_c$ obtained for each volume fraction and ratio of composite $G_c$ to matrix $G_c$ are listed in Table I. Fig. 12 shows a plot of $G_c$ against volume fraction for the composites and polyurethane matrix. A plot of ratio of composite $G_c$ to matrix $G_c$ is also shown in Fig. 13. It is clear from Figs 12 and 13 that the value of $G_c$ decreases for the lower volume fraction ($\phi_f = 0.045$). The observation that a small volume fraction of fibres reduces the toughness of a polymer is expected and is well known. With increasing value of $\phi_f$ there is a gradual increase in the value of $G_c$ and it almost equals the value of $G_c$ for unfilled polyurethane matrix (for $\phi_f = 0.088$), the increase in the value of $G_c$ continues to increase with further increase in the filler concentration ($\phi_f = 0.126$ and 0.158).

The impact strength of particulate filled composites usually decreases with increase in filler concentration of rigid fillers. On the other hand, incorporation of fibres as reinforcing materials generally leads to an increase in impact strength, up to a reasonable level of loading and then must fall again with a further increase in concentration of fibres. In the case of composites of discontinuous fibres which have a wide range of fibre lengths, as is the case with hammer-milled fibre glass used in this study, the analysis of impact behaviour of composites becomes complicated.

TABLE I Values of strain energy release rate, $G$, of unfilled polyurethane and composites.

| $\phi_f$ | SD | $G_c$ KJ m$^{-2}$ | $G_c^c G_c^m$ |
|---|---|---|---|
| 0.00 | 4 | 1.642 | 1 |
| 0.045 | 4 | 1.25 | 0.761 |
| 0.088 | 4 | 1.615 | 0.984 |
| 0.126 | 4 | 1.823 | 1.110 |
| 0.158 | 4 | 2.031 | 1.237 |

$G_c^c$ = Strain energy release rate of composite.
$G_c^m$ = Strain energy release rate of matrix.

The toughening effect of fibres can be attributed to the energy required to fracture the matrix, fibre removal, resistance to crack propagation, interfacial bond, fibre length, concentration of fibres and their geometrical organisation. Decrease in impact strength of polyurethane composites of lower volume fraction ($\phi_f < 0.1$) and subsequent increase in the impact strength of higher-volume fraction composites suggests that two different types of mechanism are playing their part in the process of fracture. For lower-volume fraction ($\phi_f < 0.1$), fracture takes place mainly by fibres being removed from the matrix. In this case, removal of fibres is facilitated by the growth of cracks at the fibre end, particularly the fibres which are much shorter than the critical fibre length and any misaligned fibres are pulled through the matrix before the phenomenon is completed. It is also possible that a few fibres may fracture in the process. At a higher-volume fraction both, the cohesive strength of the matrix and the energy required to pull out contribute to the impact strength of the composite in accordance with the rule of mixtures.

This behaviour can be explained by considering crack propagation. At lower-volume fractions, the

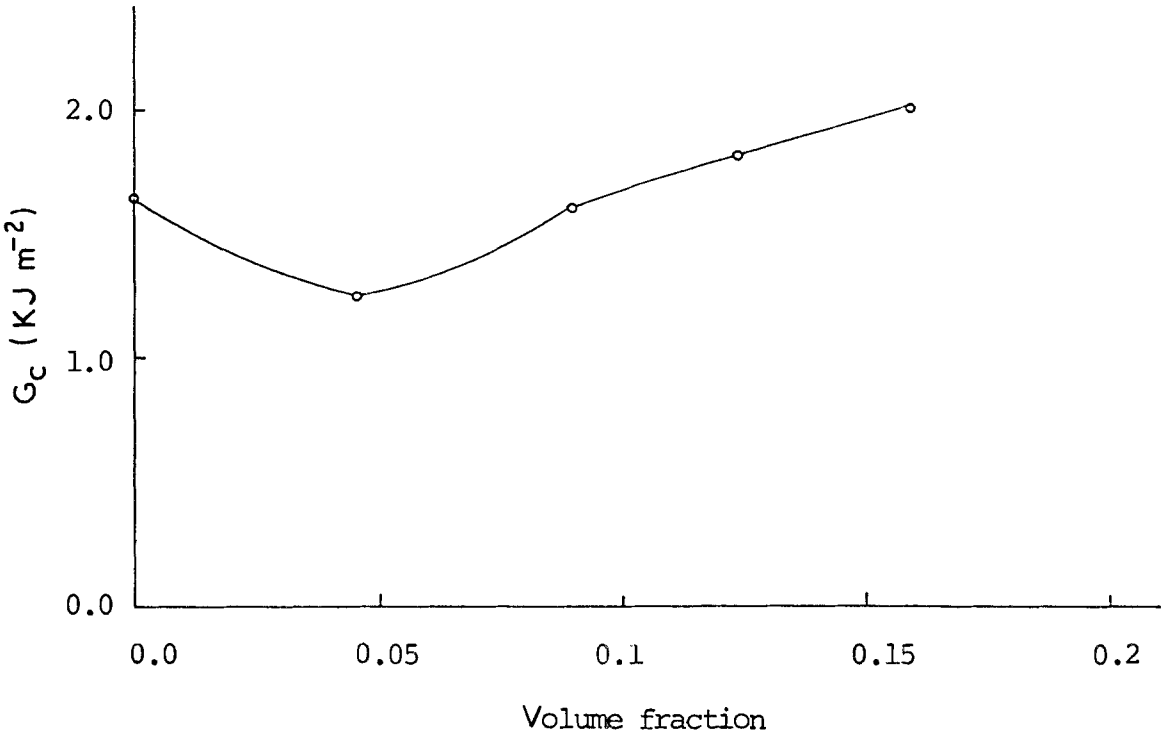


*Figure 12* Plot of $G_c$ against volume fraction for polyurethane composites.

number of fibres in the composite is not sufficient enough to stop, blunt or deviate the propagation of the crack. The distribution of stresses at the crack-tip region is not altered and the crack continues to grow smoothly. As a result, the crack can advance easily through the matrix, leaving fibres bridging the crack. Consequently, the energy to fracture a specimen is the work required to pull these fibres out of the matrix.

As the volume fraction of fibres is increased to a critical level where the number of fibres is sufficient to blunt, deflect, or even stop the propagation of the crack, the distribution of stresses at the crack-tip alters, and the crack cannot possibly propagate through the matrix without altering its path. This change, however, depends upon the particular arrangement of fibres and also on the interfacial bond between the fibres and the matrix. In the case of the composites ($\phi_f > 0.1$), the energy required to fracture a specimen will be a combination of energy required to pull the fibres out from the matrix. Moreover, at higher-volume fractions, there is a likelihood of interaction of fibres. When one fibre is surrounded by another, the toughening effect may come from the energy dissipated in frictional sliding of one fibre inside the other during deformation [15]. A similar trend has been observed by Phang [14] while studying the fracture

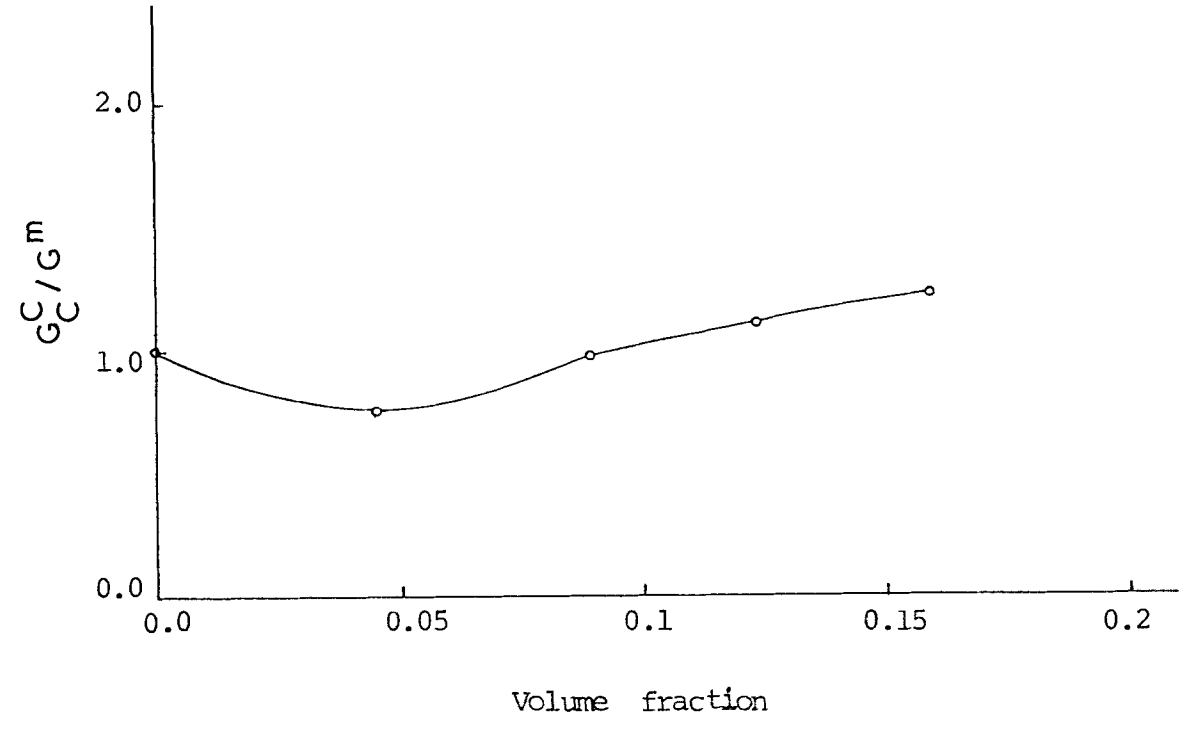


*Figure 13* Plot of $G_c^c/G_c^m$ against volume fraction for polyurethane composites.

properties of short glass-fibre reinforced nylon 6.6. Evidences of two different modes of fracture mechanism can be observed from the fracture surfaces of composites of different volume fractions. Lower volume fractions ($\phi_f > 0.1$) will show the evidence of the contribution of increased number of fibres.

## 5. Conclusions

The quantitative assessment of the parameter of toughness, the critical strain energy release rate, for unfilled and randomly oriented hammer-milled glass-fibre reinforced polyurethane can be successfully made by using an instrumented falling-weight impact machine and applying the linear elastic fracture mechanics (LEFM) approach.

It is concluded that for lower-volume fractions ($\phi_f = 0.045$) the impact resistance of the polyurethane matrix has decreased as compared to that of the matrix. However, further incorporation of hammer-milled glass fibres into the polyurethane matrix has lead to a gradual increase in the impact resistance. For $\phi_f = 0.088$ the value of $G_c$ is almost equal to that of the matrix. This trend has been observed to continue up to the maximum volume fraction ($\phi_f = 0.158$) tested in this study.


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