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Tensile rupture of short fibre filled thermoplastic elastomer

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Tensile rupture in short silk fibre filled thermoplastic elastomer blends from low-density polyethylene and natural rubber (NR) containing pre-cuts of different lengths has been studied. σ_b, the tensile strength of the blends, was found to decrease with increase in the cut length in accordance with the Griffith's theory of fracture. However, unlike in the case of vulcanized NR, no critical cut length (at which the strength drops abruptly) was found in the case of both blends and unvulcanized NR. The energy to fracture per unit volume, W_b, also varied inversely with the length of the pre-cut. Values of inherent flaw size, l_0, of the composites, determined by extrapolation of W_b to the value obtained when no initial pre-cut was present, were found to increase with fibre loading. Blends with longitudinally oriented fibres showed higher l_0 values than those with transversely oriented fibres.

1. Introduction

Tensile fracture initiates at any site where the local stress exceeds a critical level and the small flaw present in the body begins to grow as a crack. Every solid body contains naturally occurring flaws due to the presence of inhomogeneities in composition or structure, which act as stress-raisers. In addition, some flaws are introduced into the sample through imperfect cutting of the sample edges, moulding imperfections, dirt particles and other heterogeneities incorporated during the process of preparation of the composites. The role played by these flaws in initiating and promoting failure of rubber composites has been studied by Thomas [1]. Although the exact nature of these failure initiation sites is not known clearly, a value of 40 ± 20 μm is considered to be their effective magnitude in rubbers and glasses.

Many researchers [2–6] have investigated the critical crack phenomena by inflicting pre-cuts of different lengths on the surface of the samples in tensile and fatigue failure experiments. References are also available on the effect of temperature on the critical cut length in various rubbers and in glass fibre reinforced polymeric composites [7, 8]. Setua and De [9] have observed a critical cut length in short fibre reinforced rubber composites.

In this paper we describe our studies on tensile rupture of short fibre filled thermoplastic elastomer blends. In our earlier studies [10, 11] we investigated the processing characteristics, physical properties and failure mechanism of short silk fibre filled composites of natural rubber (NR) and polyethylene (PE). In the present studies we have chosen a 70 : 30 blend of NR and low-density polyethylene (LDPE) filled with different loadings of short (6 mm) silk fibre. We have applied pre-cuts of varying lengths (0.25 to 3.0 mm) on tensile test specimens and determined their effect on the tensile strength. The theoretical inherent flaw size in the fibre filled blends has also been determined.

2. Experimental details

The compositions of the mixes are given in Table I. The details of the mixing and moulding procedures of these blends have been described elsewhere [10]. For studying the effect of pre-cut length on the tensile strength of the blends, dumb-bell shaped tensile test pieces were stamped out from the moulded sheets along the grain direction. Care was taken to ensure cutter sharpness. In some cases dumb-bells were punched out both along and across the grain directions. The direction of fibre orientation is shown in Fig. 1. A special jig was constructed to hold the specimens rigidly while the prescribed chisel cuts were being applied through the exact centre and directed perpendicularly towards the applied tensile force. The shape of the jig attached with a chisel of an arbitrary cut length and the position of a test specimen held rigidly by a die over the basement of the construction are given in Fig. 2. The diameter of all the chisels of different cut lengths was maintained to be the same.

All the properties were measured in the longitudinal direction of fibre orientation only, unless mentioned otherwise.

In the case of raw natural rubber samples, the rubber was masticated in a Brabender Plasti-corder (model PLE 330) for 5 min at 60° C and then moulded between aluminium foils at 150° C for 2 min. Dumb-bell shaped specimens were punched from the sheets and after the infliction of pre-cuts, tested in the Instron Universal Testing Machine (model 1195) within 30 min of moulding to keep shrinkage to a minimum. Natural rubber vulcanizates were prepared in the conventional manner on a two-roll mill at room temperature and cured to optimum time in a press at 150° C.

3. Results and discussion

Fig. 3 shows the variation of tensile strength (σ_b) with increasing pre-cut length (l) for various loadings of

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TABLE I Compositions of the mixes

Ingredients	Mix					
	0C	5C	10C	20C	30C	
NR*	70	70	70	70	70	100
LDPE†	30	30	30	30	30	–
Silk fibre‡	–	5	10	20	30	–
ZnO	–	–	–	–	–	5
Stearic acid	–	–	–	–	–	2
CBS	–	–	–	–	–	0.8
S	–	–	–	–	–	2.5

*Crumb rubber, ISNR 5 grade: obtained from the Rubber Research Institute of India, Kottayam. Density = 0.920 g cm^{-3}.

†Low density polyethylene (LDPE) – Indothene 16 MA 400: obtained from Indian Petrochemicals Corporation Ltd, Vadodara. Density = 0.916 g cm^{-3}; MFI = 40 g/10 min; Crystalline melting range = 110 to 112° C.

‡Waste silk fibre (mulberry type) was supplied by the Silk Khadi Mondol, Bishnupur. It was first separated from undesirable foreign matter and then chopped to 6 mm length. Average diameter = 0.045 mm. Silk fibre loading was based on the rubber phase only.

short silk fibre in natural rubber (NR)–low density polyethylene (LDPE) (70 : 30) thermoplastic elastomer blends. For comparison, a similar experiment has been carried out on raw NR and vulcanized NR (Fig. 4). As shown in Fig. 3, there is a decrease in tensile strength with increasing cut length in accordance with Griffiths's theory,

$$\sigma_b = \left(\frac{G_c E}{\pi l}\right)^{1/2} \qquad (1)$$

where G_c is the tearing energy, and E the Young's modulus of the material, or

$$\sigma_b \propto 1/l^{1/2} \qquad (2)$$

Equation 2 has been tested by various authors for rubbers [2–6] and plastics [7, 8]. As reported by several authors [9, 12, 13], we also observed that in the case of vulcanized NR, the tensile strength decreased steadily up to a fairly well-defined critical cut length at which there was an abrupt fall in tensile strength (Fig. 4). What is interesting is that no such critical cut length was observed in the case of NR–LDPE blends. Thomas and Whittle [14] have reported a critical temperature and a critical cut depth for vulcanized NR. Specimens undergo failure by two different mechanisms: below the critical cut length, it is essentially a crack growth process initiating from the edge of the specimen and moving towards the centre, while above it, the failure occurs primarily due to tearing commencing from the tip of the applied cut. Our results suggest that these distinct mechanisms do not operate in the case of NR–LDPE blends. This may be due to the fact that interfacial adhesion between the rubber and the plastic phases plays a dominant role in the deformation of these blends and in deciding the stress-transfer in the composites. If the interfacial adhesion is sufficiently low, small cavities may be opened up at the interface at a certain applied load. These cavities are equivalent to big flaws which tend to initiate and propagate failure. To check further, gum natural rubber was tested but no critical cut length was found (Fig. 4). From these observations it can be concluded that the tearing mechanism is the same throughout the range of cut lengths studied for both gum NR and NR–LDPE blends.

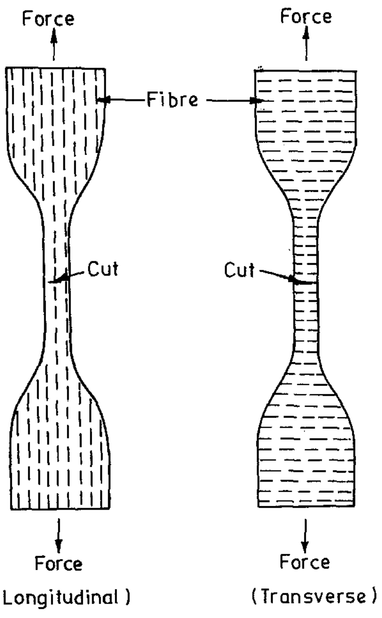

Figure 1 Tensile test specimen with direction of fibre orientation.

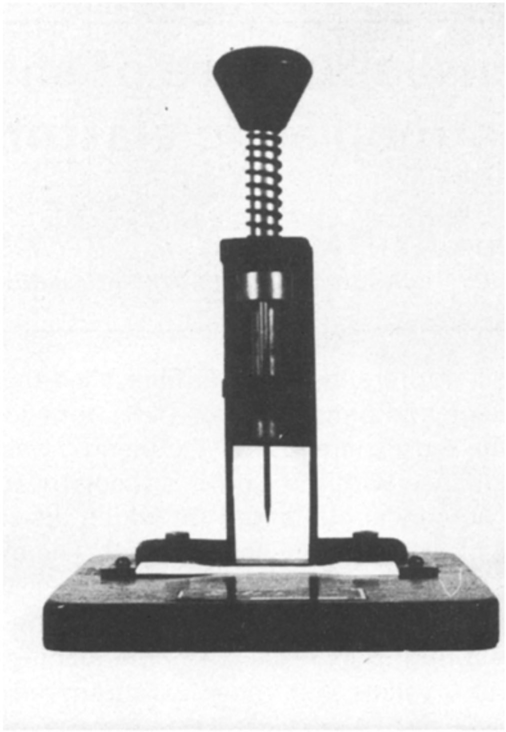

Figure 2 Shape of the jig.

Short fibre reinforced NR vulcanizates show a critical cut length [9]. However, when short fibres are introduced into the NR–LDPE blends no such critical cut length is found (Fig. 3). This implies, once again, that the mechanism of tearing remains the same over the whole cut length range studied. The exact mechanism by which the tearing occurs over the whole region is now under study.

The effect of fibre orientation on the tensile strength of NR–LDPE blends is shown in Fig. 5. The mix with transverse fibre orientation has lower strength properties than that with longitudinal fibre orientation. Similar observations have been made earlier [9, 15, 16] for

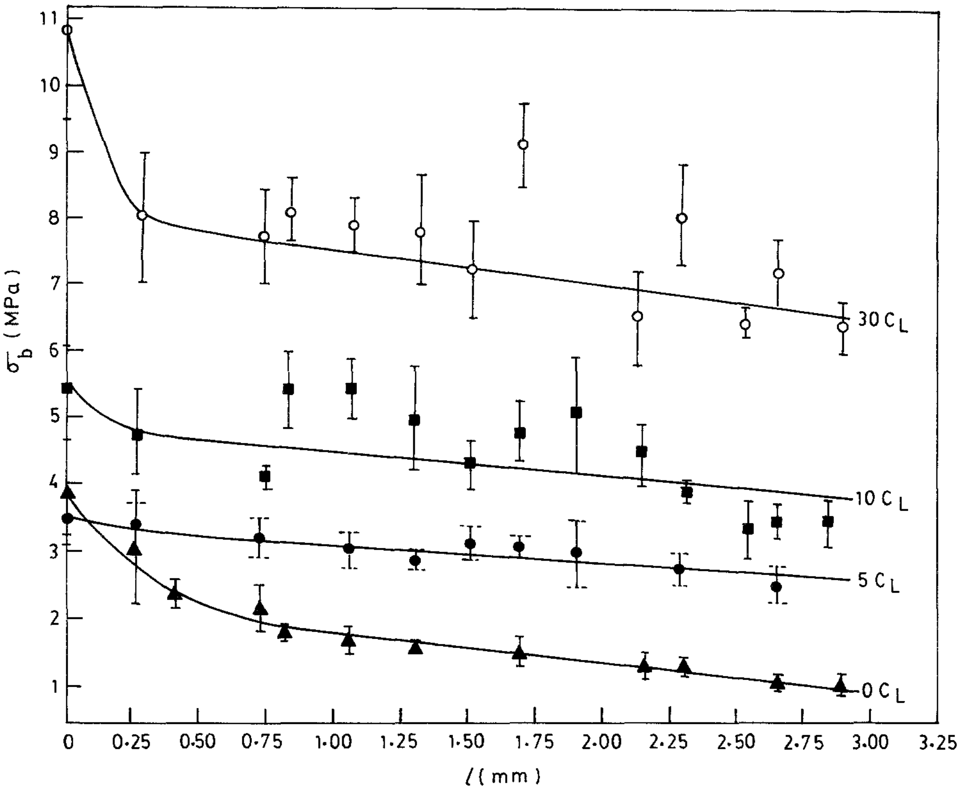

Figure 3 Plot of tensile strength (σ_b) against cut length (l) for fibre filled NR–LDPE blends.

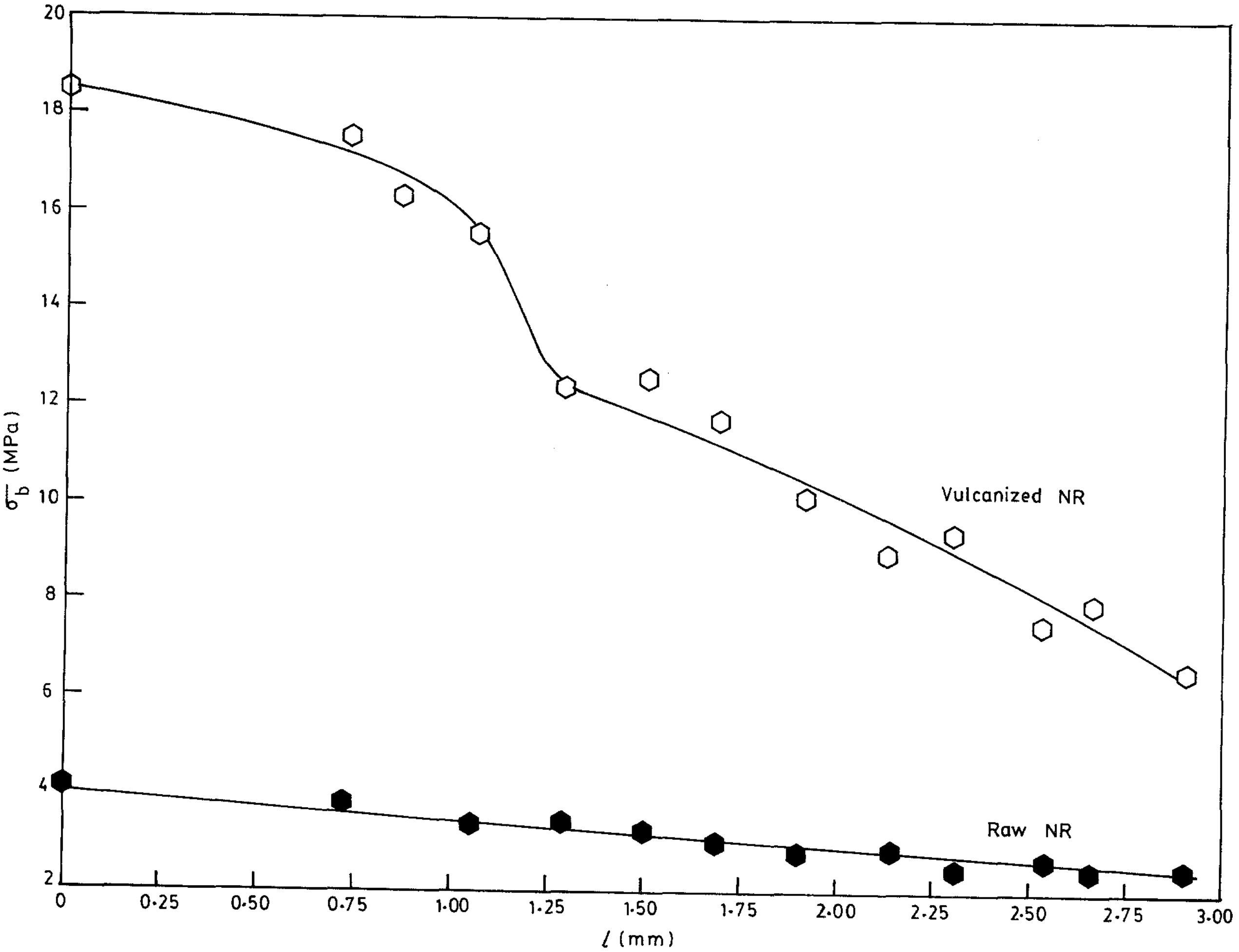

Figure 4 Plot of tensile strength (σ_b) against cut length (l) for gum raw NR and NR vulcanizate.

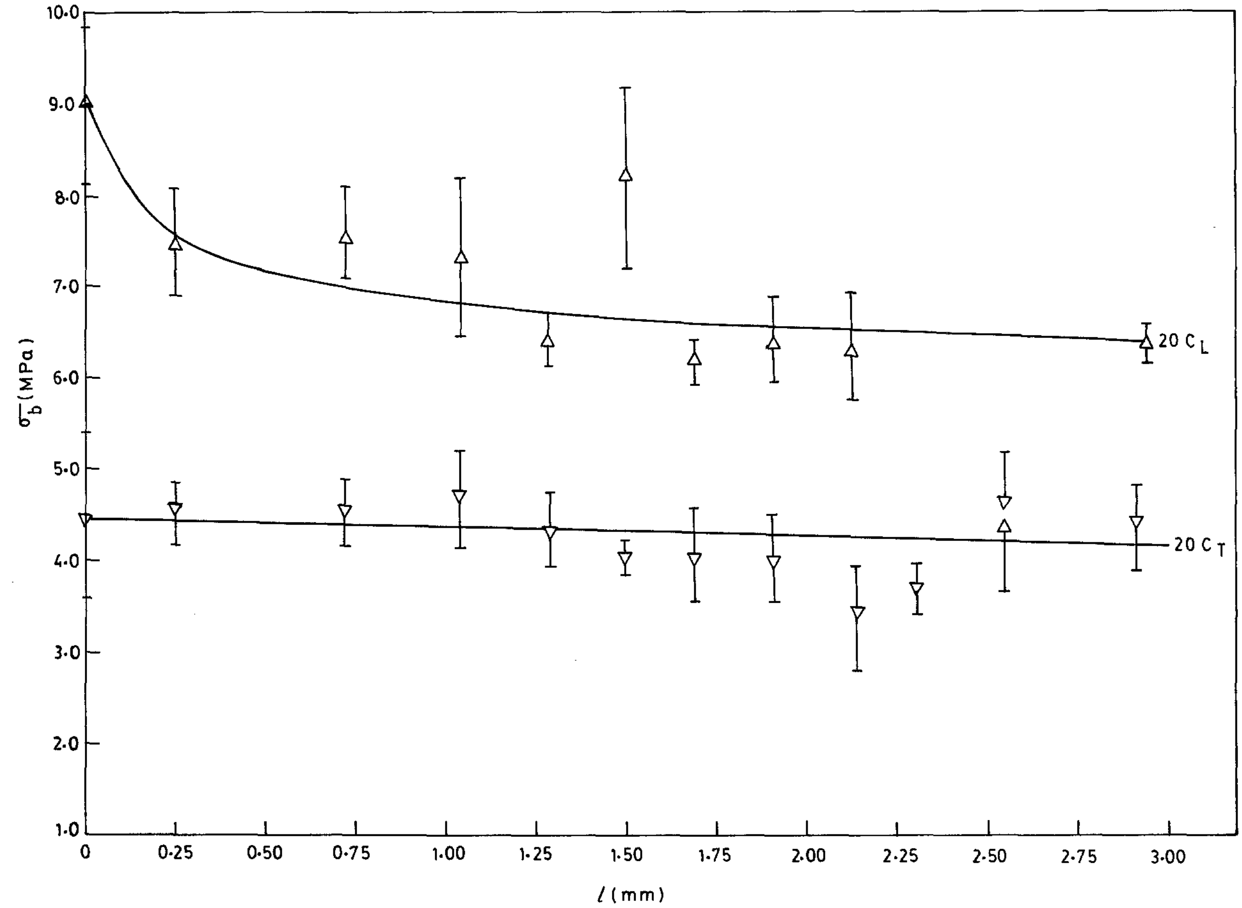

Figure 5 Plot of tensile strength (σ_b) against cut length (l) for NR–LDPE blends filled with 20 p.h.r. fibre. (L denotes longitudinal fibre orientation and T, transverse fibre orientation.)

short fibre reinforced rubber vulcanizates. From Fig. 5 it can be seen that the size of the artificial pre-cuts has little effect on the tensile properties of the blends in the case of mixes with transverse fibre orientation, while in the case of longitudinal fibre orientation the strength decreases parabolically with increasing cut length. This illustrates clearly the effect of fibre alignment in deviating the tear path.

Fig. 6 shows a plot of log σ_b against log l for the blends. The straight lines through the points were drawn with a slope of $-\frac{1}{2}$ according to the relationship [16]:

$$\sigma_b = kl^{-1/2} \quad (3)$$

where k is the proportionality constant. The value of k increases with increasing amount of fibre loading due to higher values of Young's modulus for fibre reinforced composites [10].

It can be seen from Fig. 6 that the fit of these lines is good only at low fibre loadings (up to 10 p.h.r.). At high fibre loadings, however, the correspondence between the theoretical and the experimental values is rather poor. This is due to the fact that in mixes with high fibre loadings there is a greater probability of failure initiating from several points at the same time rather than from a single point or site as is assumed by Equation 3. The role of a large number of fibres in deviating and obstructing the tear path may also add to the discrepancy.

At low cut lengths (less than 0.3 mm) the values of tensile strength for the fibre-filled matrix lie much below the lines through the points at higher cut lengths (Fig. 6). This is due to the fact that the apparent flaw in the matrix is more than 0.3 mm (as will be seen later) and hence artificial pre-cuts do not have any additional effect on the tensile strength of the samples.

The rupture energy per unit volume (W_b) has been plotted in Fig. 7 against the cut length (l) for various fibre-filled matrix. A series of lines with a slope of -1 were drawn over the cut length region of 0.25 to 3 mm. The lines were then extrapolated to the values of W_b when no intentional pre-cut was present. The cut length obtained is referred to as apparent flaw size (l_0) in the matrix. A similar procedure has previously been adopted for calculating the inherent flaw size in rubber vulcanizates [5, 18]. The values of apparent flaw size have been plotted against fibre loading in Fig. 8, which shows that l_0 increases rapidly with increasing fibre loading at low fibre loadings but the rate of increase becomes gradual and levels off after 20 p.h.r. fibre loading.

The constant increase in l_0 with increasing fibre loading implies that the sample without any pre-cut contained actual flaws of such a dimension, i.e. the dimension of the flaws increases with increase in fibre loading. A razor pre-cut of length l_0 is equivalent in stress-raising power at the crack tip to the stress-raising power of the real, unintentional flaws present in the samples [5]. We can conclude, therefore, that fibres oriented longitudinally in the test specimen obstruct the crack tip from propagating, as a

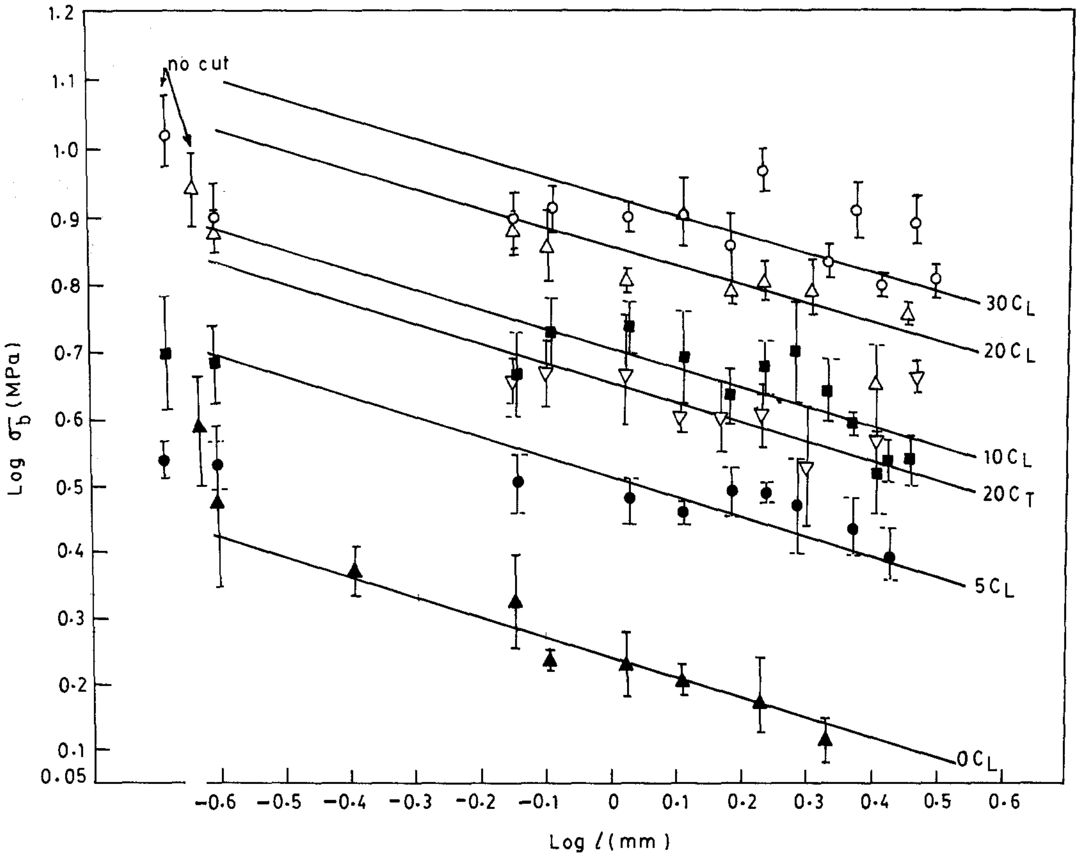

Figure 6 Plot of log tensile strength (σ_b) against log cut length (l) for fibre filled NR–LDPE blends. (L denotes longitudinal fibre orientation and T, transverse fibre orientation.)

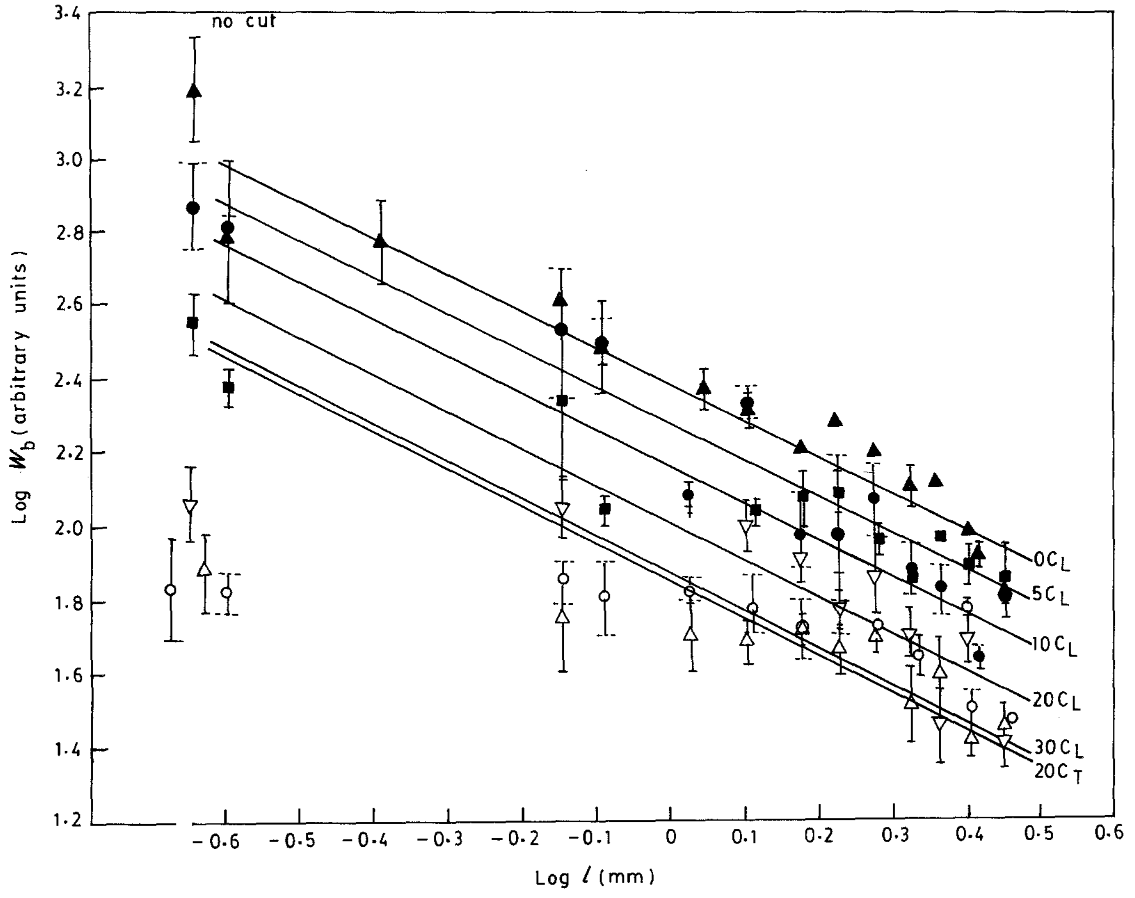

Figure 7 Plot of log rupture energy per unit volume (W_b) against log cut length (l). (L denotes longitudinal fibre orientation and T, transverse fibre orientation.)

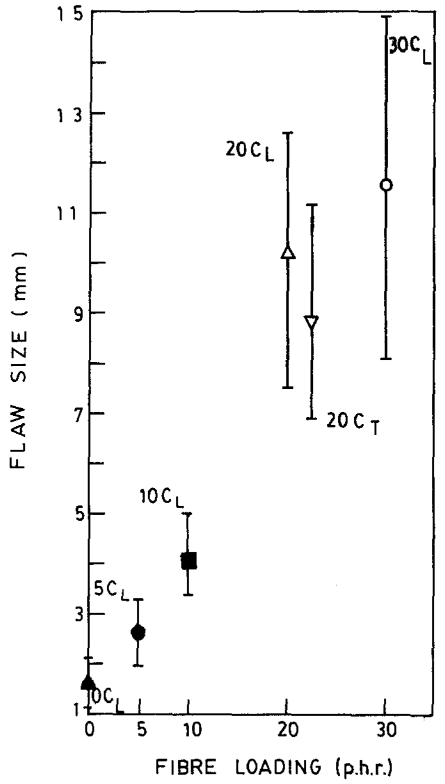

Figure 8 Theoretical inherent flaw size (l_0) plotted against fibre loading. (L denotes longitudinal fibre orientation and T, transverse fibre orientation.)

consequence of which it appears as if the inherent flaw has increased in magnitude.

We have studied the effect of fibre orientation on the inherent flaw size in mix 20C (Fig. 8). It was observed that the inherent flaw size was larger in the case of samples with longitudinal fibre orientation as compared to that in the case of samples with transverse fibre orientation. This result strengthens our earlier contention that the obstruction of the growth of the crack path is responsible for the apparent values of the inherent flaws. In the case of specimens with transverse fibre orientation, the fibres are aligned in the direction of the cut and crack growth and have much less contribution towards restricting the crack from propagating as compared to specimens with longitudinally oriented fibres which play a major role in obstructing the tear from growing rapidly [9, 10]. It has also been reported [19] that in short fibre filled thermoplastic mixes with transverse fibre orientation, the fibre ends tend to initiate crack growth because of the high stress concentration at these sites when the sample is stressed. The longitudinally oriented fibres present in the sample physically obstruct the advancing tear by deviating the tear path, whereby further resistance to tearing is experienced. The extent of physical obstruction caused by the fibres is proportional to the fibre concentration in the composites. An increase in the inherent flaw size is, therefore, observed with increase in fibre loading (Fig. 8).

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