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Strain-rate sensitive tough fibre-reinforced composites

NAK-HO SUNG, TERENCE J. JONES,* NAM P. SUH
Materials Processing Laboratory, MIT-Industry Polymer Processing Program, Department of Mechanical Engineering, Massachusetts Institute of Technology, Cambridge, Mass., USA

A new type of composite was designed and tested which has greater fracture toughness under impact loading conditions than conventional fibre-reinforced composites. This composite is strain-rate sensitive and can be more than twice as tough as conventional composites having the same matrix and fibre. The key concept used was to coat the reinforcing fibres with a thin layer of viscous fluid in order to maximize the shear stress acting on the fibres during the fibre pull-out. At a given strain-rate the shear stress can be optimized by changing the fluid viscosity and thickness of the coating. The optimum results are obtained when the frictional force is equal to the fibre strength.

Composites were made with uniaxial and randomly oriented E-glass fibres in a polyester resin. Samples with uncoated fibres were used as reference. The viscous fluids used included Dow Corning 200 Fluid with viscosities of 10^5 cP and 10^6 cP, Zelec U.N., petrolatum and silicone vacuum grease.

Notched uniaxial samples with uncoated fibres (fibre volume fraction of 0.06) showed an energy absorption of 16.8 kJ m^{-2} (3.2 ft lb in.$^{-1}$) in the Izod test. The uniaxial samples coated with Dow Corning 200 Fluid showed an energy absorption from 6.7 kJ m^{-2} (1.28 ft lb in.$^{-1}$) to 41.4 kJ m^{-2} (7.87 ft lb in.$^{-1}$) depending on the thickness of the coating. The samples with random uncoating fibres (fibre volume fraction of 0.20) had an energy absorption of 14.2 kJ m^{-2} (2.71 ft lb in.$^{-1}$) while the samples with coated fibres ranged from 13.7 to 31.6 kJ m^{-2} (2.60 to 6.02 ft lb in.$^{-1}$).

1. Introduction

Increasing use of the fibre reinforced composites is largely attributable to such mechanical properties as high specific strength, specific modulus and also relatively high fracture toughness. These properties are derived from the unique structure of composites which allows the use of high strength and modulus fibres which are normally too brittle to be used otherwise.

High strength and stiffness properties of composites requires an efficient load transfer between fibres and matrix and thus, strong adhesion at the fibre/matrix interface is desired. The fracture toughness of composites, being defined as the energy required for fracture across a unit cross-section, depends largely on the volume of material which absorbs energy by undergoing plastic deformation and fracture during the fracture process. Strong interfacial bonding in brittle-matrix/brittle-fibre composites generally leads to a catastrophic failure with cracks propagating right through the matrix and the fibres. The energy dissipation is limited to the work done in creating fracture surface. Weak interfacial bonding leads to an increase in the size of damage zone, and thus increases the energy absorption through various mechanisms to be reviewed in this section. Because of these contradicting requirements, improvement of fracture toughness is normally accomplished by sacrificing the flexural and tensile properties

*Present address: Instrumentation Laboratory, Inc. Lexington, Mass., USA.

through the control of the fibre/matrix interface.

The fracture toughness of a composite based on brittle-fibre/brittle-matrix is known to be much greater than that of the fibre or the matrix alone [1]. Several mechanisms have so far been put forward to explain the origins of toughness in fibrous composites. One of the theories by Cottrell and Kelley [2, 3] is the consideration of frictional work during fibre pull-out. In composites where fibres may fail randomly at a weak point, a section of the fibre is pulled-out from the matrix as the crack surface separates. The work done against frictional force during pull-out is considered a major contribution to fracture energy. This concept is valid when the frictional force is maintained at a level of original interfacial shear strength as in some of the metal fibre/metal matrix composites.

Another concept is that by Outwater and Murphy [4] who attribute the fracture energy to the energy required to debond the fibre from the matrix. In a composite, such as glass fibre/epoxy, fibre-resin debonding and interfibre splitting may occur at the propagating crack tip and work done in creating these new surfaces is considered as the primary contributor to the fracture energy. This theory assumes that the failure strain to the fibre is greater than that of the matrix, so that, the crack front in the matrix will go around the fibre. The excess strain energy of the fibre from the additional stretch beyond the strain of matrix failure is the source of the debonding work. Cook and Gordon [5] view the debonding differently such that the interfacial failure occurs in the region ahead of the crack tip due to the tensile component of the stress field in the direction perpendicular to the fibre. In either case, it is predicted that the larger the debonding area, the higher the energy dissipated and thus the larger the fracture toughness.

Piggott [6] proposed that the energy is dissipated through plastic deformation of the matrix when the strain energy of the fibre is transferred to the matrix after a fibre breaks and relaxes. This theory assumes that fibre-resin bonding remains intact throughout and that the strain energy gained by the matrix is insignificant.

Whether or not any particular mechanism of energy absorption is predominant and, therefore, can predict fracture toughness of a certain composite depends primarily on the properties of the composite in relation to various assumptions made in each theory. As pointed out by Marston *et al.* [7], none of these theories alone can successfully predict the fracture toughness of a certain composite. They proposed an approach of combining all the above mechanisms with Outwater/Murphy's debonding contribution being substituted with a more generalized term of surface energy. The surface energies of the fractured fibre and the matrix are included as an important modification as well as that of the surface created by debonding.

All the above theories are primarily for uniaxially oriented fibre-reinforced composites. Theories predicting the fracture toughness of randomly oriented fibrous composites are rather scarce. Similarity in the fracture toughness of oriented and random composites has been reported recently by Hing and Groves [8], and models for predicting crack initiation energies and fracture energies of notched and unnotched impact specimens of random fibre composites have been presented by Williams *et al.* [9].

Fracture toughness of yet another type of composites, namely, cross-plied unidirectional ply laminates and woven fabric laminates has been investigated by McGarry and Mandell [10]. The formation of splits between longitudinal fibres, delamination of the longitudinal and transverse plies, and some degree of fibre pull-out are considered as the major energy dissipation mechanisms. The fracture toughness was estimated as the loss in elastic energy of the ligament material along the split when the crack propagates.

From the review of various theories, it is clear that the fracture toughness of fibrous composites is largely determined by the properties of the fibre-matrix interface. Various attempts have been made towards the modification of interfacial properties, Notable examples may be the concept of the "intermittent bonding" by Atkins [11, 12] and that of the "duplex fibres" by Morley [13, 14].

In Atkins' approach, strong and weak interfacial bondings were arranged in an alternating manner along the length of the fibre by introducing intermittent coating. The strong regions provide the load transfer requirements whereas the weak regions serve to blunt the propagating cracks by the Cook/Gordon debonding mechanism [5] and thereby increase the area of debonding and the length of fibre pull-out [12]. While some improved fracture toughness has been reported in boron/epoxy composites, question still remains as

to why the fracture toughness values differ depending on the type of coating used, should the proposed debonding mechanism of Cook/Gordon be operative.

The Morley's duplex fibre consists of a hollow outer element strongly bonded to the matrix and an inner core element weakly bonded to the outer shell. The strong bond of matrix/outer shell interface is to provide good flexural and tensile properties where as the weak bond of inner core/outer shell is to resist the transverse crack propagation and to cause the pull-out of the inner core [14]. The concept is attractive. However, the relative effectiveness of employing the "duplex fibre" is not yet clear and the preparation of such a "duplex fibre" appears to be rather difficult for this concept to be practical.

Recent studies on fibre-reinforced phenolics [15] indicated that the frictional work done during the fibre pull-out appears to be the major source of the fracture energy. In this paper, a new concept of improving the fracture toughness of fibre-reinforced composites is presented. The concept, developed at M.I.T., utilizes fibres coated with a thin layer of viscous fluid [16]. During the fracture of composites, fibres at the fracture plane are subjected to pull-out through the viscous shear force acting on the fibre. The energy is dissipated by viscous shear work during fibre pull-out. The shear stress acting on the fibre is strain-rate sensitive and can be varied through the control of the viscosity and the thickness of the coating to maximize the energy of the fibre pull-out and thus the fracture toughness. This contrasts distinctively with conventional composites in that the force acting on the fibre during the pull-out in debonded area is Coulomb friction force which is strain-rate independent, and also that there is no way of controlling this force for the purpose of maximizing the energy absorption. Theoretical analysis of the concept and the experimental results are presented.

2. Theoretical consideration

Let us consider a case where a fibre, coated with viscous fluid and embedded in a matrix, is being pulled out from the matrix as shown schematically in Fig. 1. The force resisting the fibre pull-out, F_r, is

$$F_r = 2\pi r l \tau \tag{1}$$

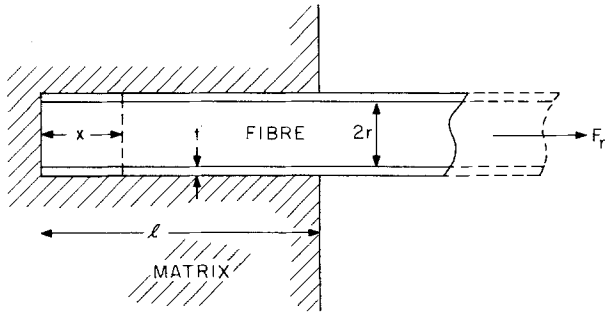
where r is the radius of the fibre, l the length of the fibre being pulled out and τ the viscous shear stress acting on the fibre surface.

Whether the fibre will be pulled-out or will break depends on the relative magnitude of F_r and the ultimate breaking strength of the fibre. To achieve the maximum energy dissipation through the viscous shear work, the fibre must be pulled out against the maximum possible resisting force without breaking. This condition can be stated as

$$r\sigma_f \geqslant 2l\tau \tag{2}$$

where σ_f is the tensile strength of the fibre.

The shear stress is a function of the viscosity of the coating fluid and the velocity gradient across the thickness of the coating. As for most viscous fluids, the relationship among the shear rate, viscosity and shear stress can be expressed



F_r = Fibre pull-out force
ℓ = Length of the fibre embedded
r = Fibre radius
t = Thickness of the coating material
x = Fibre pull-out distance

Figure 1 A schematic model of a coated fibre embedded in a matrix.

as follows

$$\tau = \mu_{\text{eff}} \left(\frac{dV}{dR} \right) \tag{3}$$

where V is the velocity of fluid parallel to the fibre axis relative to the matrix, R is a co-ordinate axis perpendicular to the fibre axis and μ_{eff} is the effective viscosity.

From Equations 2 and 3, the conditions for the maximum energy dissipation can be generalized for most fluids as

$$\sigma_f r \geqslant 2l\mu_{\text{eff}} \left(\frac{dV}{dR} \right). \tag{4}$$

If Equation 4 holds, the work of fibre pull-out, w, can be written as a function of the pull-out distance, x, as

$$w = 2\pi r\tau \left(lx - \frac{x^2}{2} \right) \tag{5}$$

or

$$w = 2\pi r\mu_{\text{eff}} \left(\frac{dV}{dR} \right) \left(lx - \frac{x^2}{2} \right). \tag{6}$$

Thus, for the complete pull-out of the fibre, $x = l$, and the total amount of work becomes

$$w = \pi r\mu_{\text{eff}} \left(\frac{dV}{dR} \right) l^2. \tag{7}$$

For a non-Newtonian fluid, μ_{eff} is a function of the shear strain-rate and a power law may well express the relationship as

$$\tau = m \left| \frac{dV}{dR} \right|^{n-1} \left(\frac{dV}{dR} \right) \tag{8}$$

where the exponent n is a material property, known as the flow index, m is a constant for a given fluid. Thus,

$$\mu_{\text{eff}} = m \left| \frac{dV}{dR} \right|^{n-1}. \tag{9}$$

The effective viscosity μ_{eff} is related to the apparent viscosity μ_{app}, which is measured empirically by a capillary viscometer, as

$$\mu_{\text{eff}} = \frac{4n}{3n+1} \mu_{\text{app}}, \tag{10}$$

and thus,

$$\tau = \frac{4n}{3n+1} \mu_{\text{app}} \left(\frac{dV}{dR} \right). \tag{11}$$

In order to obtain the qualitative interdependence among the variables, one may approximate the fluid as being Newtonian and the shear field being uniform. This is a reasonable approximation for pseudo-plastic fluid at very low and high strain-rates. Then, $\mu_{\text{eff}} = \eta$ and $dV/dR = V_0/t$ where η is the Newtonian viscosity, t is the thickness of the coating, and V_0 is the velocity of fibre pull-out. Equations 4 and 6 respectively become

$$\sigma_f r \geqslant 2l\eta \left(\frac{V_0}{t} \right) \tag{12}$$

and

$$w = 2\pi r\eta \left(\frac{V_0}{t} \right) \left(lx - \frac{x^2}{2} \right). \tag{13}$$

Equation 12 shows that for a known, or anticipated velocity of impact, an appropriate coating thickness and viscosity can be chosen to achieve the maximum energy absorption condition. For a given construction of composite, Equation 13 shows how the total pull-out energy depends on the velocity of pull-out (or the rate of impact).

For a composite with discontinuous fibres of length L, the possible length of the fibre pull-out during fracture will vary from zero to $L/2$, depending on the position of crack tip with respect to the fibre length. The average pull-out length will be $L/4$. The optimum fibre aspect ratio for maximum energy dissipation may be found by substituting $l = L/4$ into Equation 13 and rearranging terms. The result is

$$\frac{L}{D} \leqslant \frac{\sigma_f t}{\eta V_0} \tag{14}$$

where D is the fibre diameter.

To demonstrate the concept described above, experiments were carried out to show the dependence of the fracture toughness of composite on varying thickness and the viscosity of the coating. As can be seen in Equation 12, shear force during fibre pull-out could be very small if viscosity of the coating fluid is too low or the thickness of the coating too large. The fracture toughness of composites with coated fibres could, therefore, be lower than that with uncoated fibres. At a given velocity of impact, an increase in viscosity or a decrease in thickness of the coating generally represents an increase in shear force and thus, of fracture toughness.

3. Experimental

3.1. Materials

Two types of composites were studied in this work. They were: (i) uniaxially oriented long fibre

composite, and (ii) randomly oriented short fibre composites.

The uniaxial composites were made of unsaturated polyester resin (Laminac 4155 by American Cyanamid) and 3-ply E-glass yarn (by Owens Corning). Methyl ethyl ketone peroxide was used as a curing agent. The fibre volume in this composite was about 6%.

The randomly oriented composites consisted of unsaturated polyester resin (Paraplex 340, Rohm & Hass, Inc), 0.635 cm length chopped E-glass fibres (type 308A, Johns-Manville) and calcium carbonate fillers (by Pfizer, Inc). The curing agent was *t*-butyl perbenzoate. Composites contained about 40% by weight of fillers and the fibre volume was about 20%.

3.2. Coating materials

It is imperative that the coating material forms and remains in a discrete layer between the matrix and fibre in a final composite. Suitable materials should, therefore, be incompatible and non-reactive with resin matrix as well as with other ingredients in the composite system.

A number of viscous fluids were evaluated for possible use as a coating material. Qualitative experiments were made to test the miscibility and the reactivity of various fluids with polyester resin system and the results are listed in Table I. Among the materials listed, frequent use of silicon vacuum grease and Dow Corning 200 Fluid was made because of their thermal stability and the availability of a wide range of viscosities (0.6 to 2 x 10^6 cP). Fig. 2 shows the typical temperature dependence of the viscosity of various grades of the Dow Corning 200 Fluid.

3.3. Composite fabrication

3.3.1. Uniaxial composites

Composites with uniaxially oriented fibres were fabricated in the following manner. E-glass yarn was coated by passing it through a vessel containing the viscous fluid and then through a series of dies with decreasing hole diameters. The coating thickness was controlled by the inner diameter of the last die. Since the yarn was three-plied, the cross-section was not perfectly circular. Therefore, the amount of the coating was determined by the weight gain. It was noticed that the geometry of the coating die (in particular the angle of the taper) had a strong influence on whether or not the coating materials would penetrate into the interfibre spaces and possibly result in a non-uniform coating.

TABLE I Miscibility and reactivity of the coating fluids with polyester resin

Coating fluid	Miscibility	Reactivity
Polypropylene glycol-100	none	high
Mineral oil	none	none
Zelec (ne)	slight	none
Petrolatum	none	none
Silicone grease	none	none
STP motor oil	none	slight
Dow Corning 200 Fluid	none	none

The E-glass yarn, either coated or uncoated, was wrapped around a metal frame as shown in Fig. 3. The spacing between adjacent fibres was 0.254 cm. Four of the frames wrapped with yarns were then stacked on top of each other and clamped together by a set of bolts. The spacing between the layers of yarn was also 0.254 cm. Premixed resin and catalyst were then poured over these frames in an open pan and allowed to cure at room temperature. After curing, composite was then postcured at 77° C for an additional hour. The fibre volume of composites thus prepared was about 6%.

3.3.2. Randomly oriented composites

A solvent coating technique was used to coat the short fibres. The fibres were immersed in a solution of the coating materials in benzene. The excess solution was then decanted and the wet fibres were dried on a wire mesh under forced air convection. The amount of coating was controlled by the concentration of the coating solution.

Composites with randomly oriented short fibres were prepared using a steel trap mould shown in Fig. 4. The resin, filler and catalyst were pre-mixed using a mechanical stirrer at 66° C to reduce the viscosity of the mixture. The chopped fibres were then mixed in and the entire mixture was poured into a pre-heated mould, and placed in a vacuum oven for 15 min to degas the mixture. After releasing the vacuum, the mould temperature was maintained at 104° C for 35 min to complete the cure reaction.

3.4. Fracture toughness testing

The fracture toughness of the composites was evaluated on an Izod impact tester. Standard notched specimens of 6.35 cm x 1.27 cm x 1.27 cm were cut from the composite panels. The

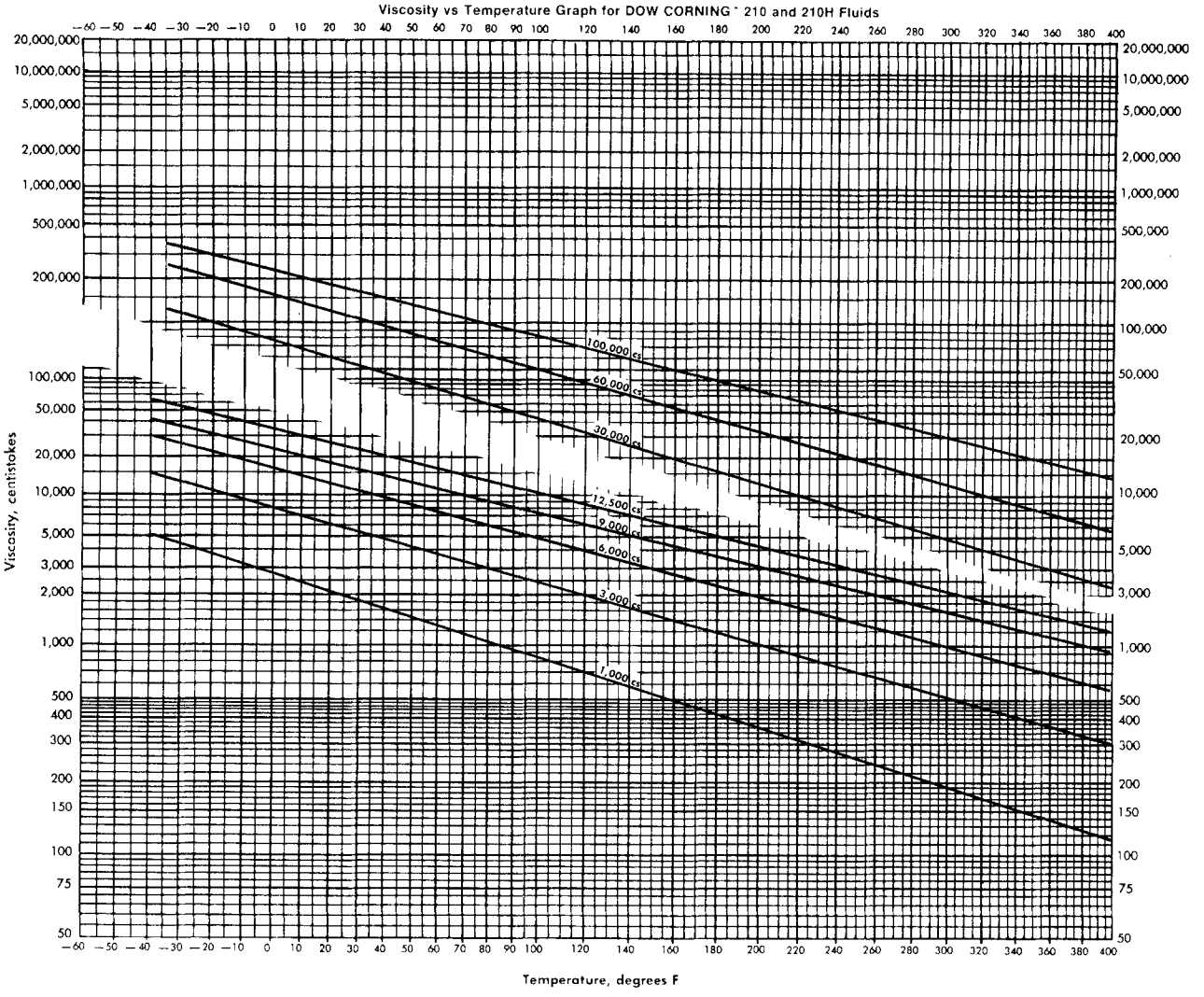


Figure 2 Temperature–viscosity relationship of Dow Corning 200 Fluid (from [18]).

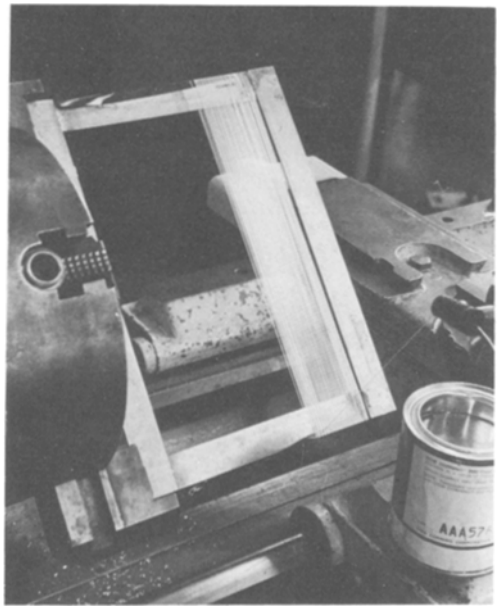


Figure 3 Photograph of the yarn wrapping arrangement.

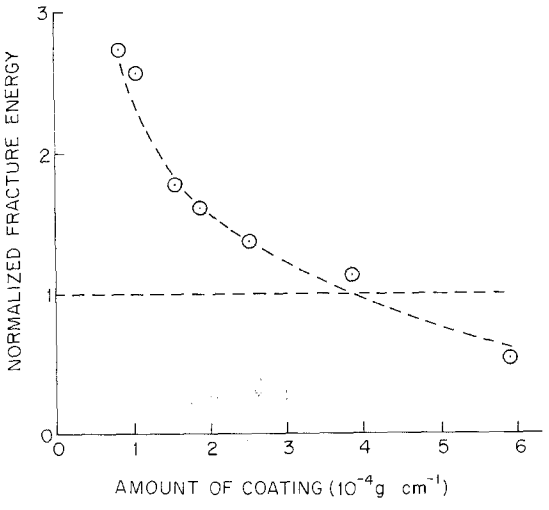


Figure 5 Fracture toughness of composite with uniaxially oriented fibres coated with silicone vacuum grease (Notched Izod impact energy is normalized to that of the composite with uncoated fibres).

full scale of 10.85 J (8 ft lb) was used and the velocity of the impact was 3.35 m sec^{-1}.

4. Results

4.1. Uniaxial composites

The fracture energies of the composites containing fibres coated with varying amounts of silicone vacuum grease were normalized to that of the uncoated fibre composite and are plotted as a function of the amount of coating in Fig. 5. The fracture energy of the control composite was 16.7 J m^{-2} (3.2 ft lb/in. of notch).

As expected from the analysis, Fig. 5 shows an approximate inverse relation between the fracture toughness and the thickness of the coating. When the coating thickness is too large, (i.e. about 5.9 × 10^{-4} g/cm of yarn) the total energy absorbed is less than that of the composite with uncoated fibres. The fracture energy increases gradually as the coating thickness decreases. At a coating level of ~0.79 × 10^{-4} g/cm of yarn, the fracture energy is more than twice that of the uncoated fibre composite.

Fig. 6 shows the different modes of failure in the specimens. The uncoated fibre composite (Fig. 6a) undergoes typical brittle fracture, where the fibres and the matrix are completely broken with very small amounts of fibre pull-out. In contrast, the coated fibre composites (Fig. 6b) exhibit failure of the matrix, but with fibres pulled-out of the matrix without being fractured. When the coating is too thick, fibre pull-out is even more extensive due to a low viscous shear force (Fig. 6c) and the associated energy absorption is accordingly very low.

Fig. 7 shows similar results obtained with the composite using fibres coated with Dow Corning 200 Fluids of two different viscosities (10^5 and 10^6 cP). The dependence of the fracture energy on the thickness of the coating is similar to the previous result. As expected, the higher viscosity fluid gives a higher value of fracture energy for a given coating thickness. However, the increase in

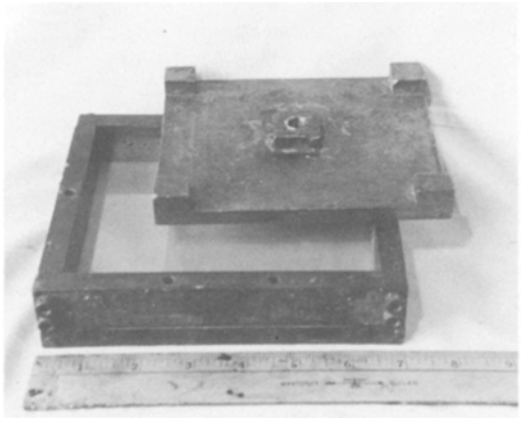


Figure 4 Steel trap mould used for moulding randomly oriented short fibre composites.

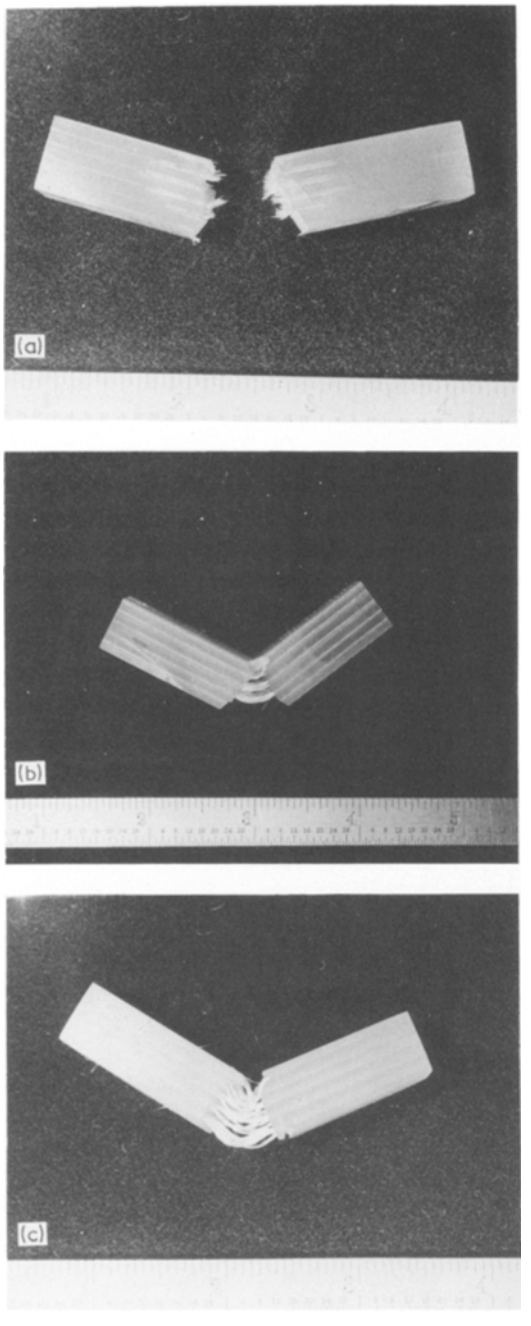


Figure 6 Photographs of fractured specimens; (a) specimen with uncoated fibres, (b) specimen with fibres coated in appropriate thickness, and (c) specimen with fibres coated excessively.

energy was not found to be linear with increase of the viscosity.

4.2. Randomly oriented composites

The fracture toughness of the composites with

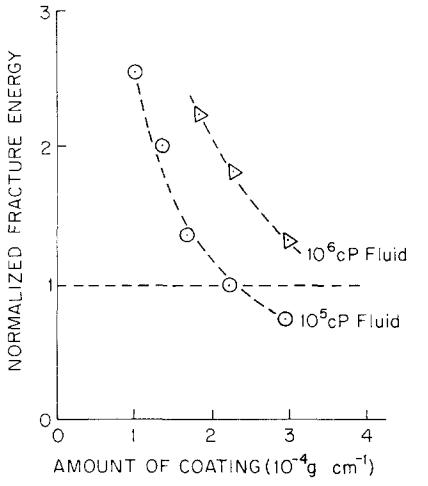


Figure 7 Fracture toughness of composite with uniaxially oriented fibres coated with Dow Corning 200 Fluid (Notched Izod impact energy is normalized to that of composite with uncoated fibres).

randomly oriented short fibres coated with Dow Corning Fluid 200 (10^5 cP), is presented in Fig. 8. The values are again normalized to that of the composite with uncoated fibres which showed an energy absorption of 14.2 kJ m^{-2} (2.71 ft lb/in. of notch). The dependence of the fracture energy on the amount of coating is similar to that in uniaxial composites. At a coating level of ca. 0.32×10^{-4} g/cm of fibres, the fracture energy was more than twice the value of the control composites. A substantial difference in the degree of the fibre pull-out was observed between the composites with coated and uncoated fibres as shown in Fig. 9.

5. Discussion

The results presented above support the validity of the concept that the fracture toughness of a fibre-reinforced composite can be improved by applying a viscous coating at the fibre/matrix interface. The dependence of the fracture energy on the thickness of the coating clearly demonstrates that the toughening mechanism is the energy absorbed by the viscous shear work of the coating fluid.

However, it appears that in designing the composites containing coated fibres, it will be necessary to consider several factors which were left out of the approximate analysis shown in Section 2. To illustrate these, consider the specific example of the data shown in Fig. 7 where the highest fracture of 41.1 kJ m^{-2} (7.87 ft lb/in. of notch) is obtained with a coating of 0.98×10^{-4} g/cm of yarn, with Dow Corning 200 Fluid

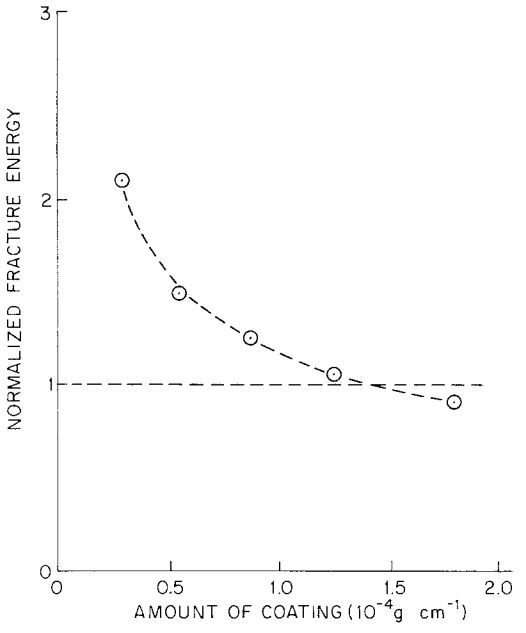


Figure 8 Fracture toughness of composite with randomly oriented fibres coated with Dow Corning 200 Fluid (Notched Izod impact energy is normalized to that of composite with uncoated fibres).

(10^5 cP). First, the actual shear force acting on the fibre during fracture can be estimated from the data. If we assume the energy of the matrix fracture is negligible, the total fracture energy of the specimen, W, can be related to the average shear stress acting on the yarns, $\bar{\tau}$, by Equation 5;

$$W = N2\pi r\bar{\tau}\left(lx - \frac{\bar{x}^2}{2}\right) \qquad (15)$$

where N is the number of yarns subjected to pull-out, and $\bar{x}$ is the average length of the pull-out. By substituting the values $W = 41.1\ \text{kJ m}^{-2}$ (7.87 ft lb/in. of notch), $N = 17.05\ \text{cm}^{-2}$ (44/in. of notch), $l = 3.18$ cm (1.25 in.), $\bar{x} = 0.60$ cm (0.235 in.), $r = 0.02$ cm (0.00787 in.), one obtains that $\tau = 11.23 \times 10^5\ \text{N m}^{-2}$ (163 psi).

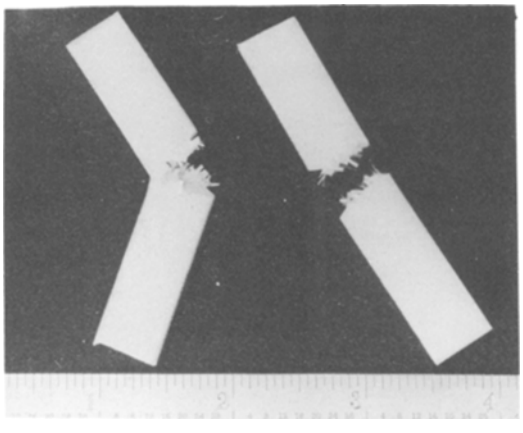


Figure 9 Photograph of fractured specimens with randomly oriented fibres; with coated fibres (left) and with uncoated fibres (right).

The predicted value of $\bar{\tau}$ can also be estimated from the velocity of fibre pull-out (or the impact velocity), and the thickness and viscosity of the coating. If we approximate the cross-section of the yarn to be a perfect circle, the coating thickness can be obtained from the coating weight. For the specific example concerned, the thickness estimated is 7.62×10^{-4} cm. The mean velocity of the fibre pull-out ($\bar{V}$) estimated from the velocity of the impact and the geometry of the fibre direction with respect to the impact direction is about 60 cm sec^{-1}. The average shear rate ($\bar{V}/t$) is then $7.87 \times 10^4\ \text{sec}^{-1}$. According to the experimental flow curve of the Dow Corning 200 Fluid shown in Fig. 10, the flow index, n, and the apparent viscosity, μ_{app}, at a shear rate of $7.87 \times 10^4\ \text{sec}^{-1}$ are 0.32 and 2.0×10^3 cP respectively. Substituting these values into Equation 11, we find that $\tau = 1.02 \times 10^5\ \text{N m}^{-2}$ (14.8 psi). This value is an order of magnitude lower than the value of $11.23 \times 10^5\ \text{N m}^{-2}$ from the fracture energy data.

Some factors which may be responsible for this discrepancy in the predicted and actual shear stress are: (i) inaccuracy of estimated coating thickness, (ii) validity of using power law, and (iii) transient response of coating fluid to shear field.

The thickness of the coating could be over estimated due to two factors. First, the cross-section of the three-ply yarn is not circular as shown in Fig. 11. Therefore, the actual interface area is larger than that estimated using the idealized circular cross-section. Secondly, the coating fluid could have penetrated into the yarn to a certain extent thus reducing the actual amount of the coating fluid at the yarn/matrix interface. If the actual coating thickness is lower by a factor of three, the shear strain-rate becomes $2.35 \times 10^5\ \text{sec}^{-1}$. From the extrapolation of μ_{app} in Fig. 10, the corresponding value of μ_{app} can be found to be 1×10^3 cP. Substituting into Equation 11, the average shear stress is then $1.20 \times 10^5\ \text{N m}^{-2}$ (17.4 psi). Therefore, the errors in estimating the coating thickness can only account for a small part of the discrepancy between the actual and the predicted shear stress.

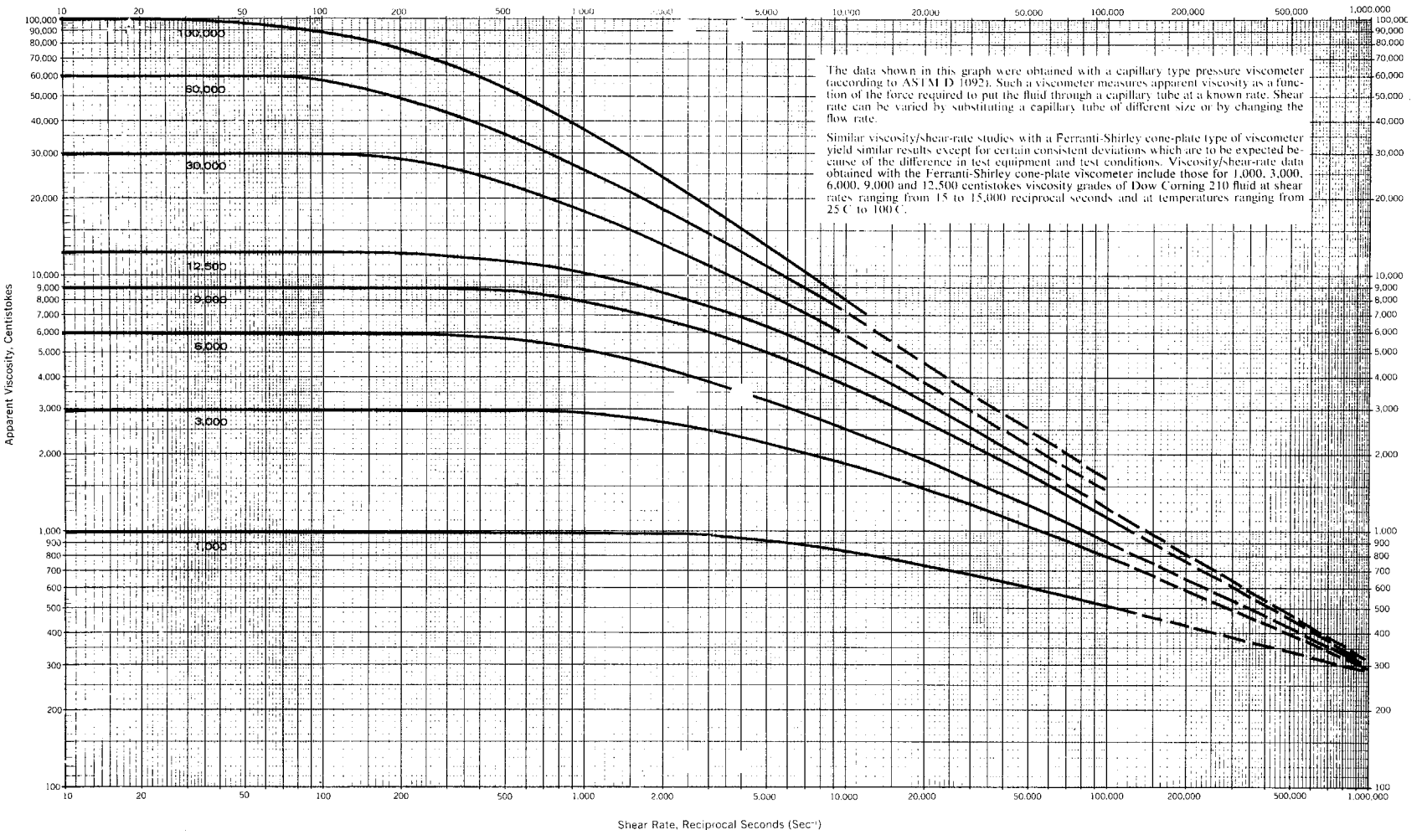


Figure 10 Apparent viscosity–shear strain rate relationship of the Dow Corning 200 Fluid (from [18]).

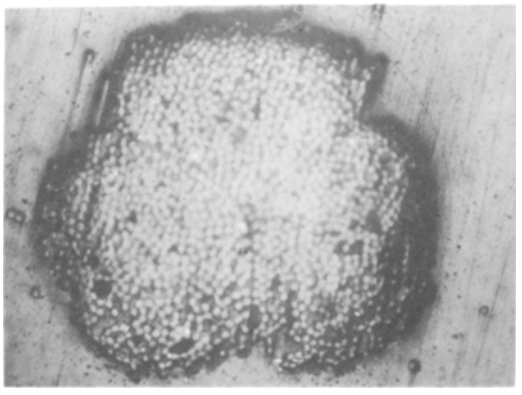


Figure 11 Photomicrographs of the cross-section of three-ply E-glass yarn.

The validity of using the power law representation for the viscosity depends on the level of the shear strain-rate. According to the experimental data in Fig. 10, the power law appears to be a reasonable approximation at the shear rate up to 10^5 sec^{-1}. At higher shear rates, however, the fluid may be in the Newtonian region (as are most visco-elastic fluids at high shear rates). If so, the Newtonian viscosity would be 1400 cP or less. Using the shear rate value of 2.35×10^5 sec^{-1} and the viscosity of 1400 cP, the maximum value of the shear stress is estimated to be 3.28×10^5 N m^{-2} (47.6 psi). This value is still much less than that obtained from the fracture energy. Clearly, the possible errors considered so far cannot completely account for the discrepancy between the actual and predicted values of $\bar{\tau}$.

In estimating the shear stress using the power law or the Newtonian approximation, it was implicitly assumed that the fluid flow in the coating had reached a steady state condition. For a viscoelastic fluid, however, there is a characteristic relaxation time during which the fluid adjusts to a step change in shear stress. During this transient period, the shear stress is generally much higher than during steady state. Fig. 12 shows the typical transient effect of shear stress with respect to the characteristic time scale of the fluid. To assess the significance of the transient effect, the relaxation time of the Dow Corning fluid is compared with the time scale of the fracture process of the composites.

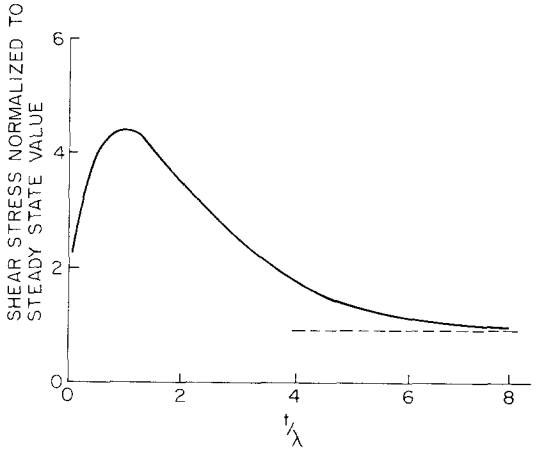


Figure 12 Transient response of viscoelastic fluid to sudden imposition of shear field (from [17]).

The relaxation time, λ, of the fluid can be estimated from Bueche's model ([17] p. 147).

$$\lambda = \frac{12\eta M}{2\rho RT} \tag{16}$$

where η = viscosity (P), M = molecular weight, ρ = density (g m^{-3}), R = gas constant (8.32×10^7), T = temperature (K). The molecular weight of the Dow Corning 200 Fluid (10^5 cP) used in this experiment is 75 000 and, therefore,

$$\lambda = 3.7 \times 10^{-3} \text{ sec.}$$

The duration of the fibre pull-out during fracture is approximately equal to the time required by the hammer to pass twice the sample thickness

$$t = \frac{2 \times 0.5 \text{ in.}}{131 \text{ in. sec}^{-1}} = 7.6 \times 10^{-3} \text{ sec}$$

and, therefore, $t/\lambda \cong 2$.

It is apparent from Fig. 12 that for the time scale of $t/\lambda = 2$, the fluid is in the transient regime and the actual shear stress acting on the yarn during fracture is much higher than the estimated steady state value. Transient effects can, therefore, account for a large part of the discrepancy between the actual and the estimated shear stresses.

The relaxation time of the Dow Corning 200 Fluid with a viscosity of 10^6 cP is likewise estimated to be about 0.15 sec. This is much longer than the duration of the fibre pull-out and t/λ is approximately 0.05. According to Fig. 12, the shear stress developed in a time scale of $t/\lambda = 0.05$ is much less than the maximum transient stress and possibly lower than the steady state value. Therefore, fibres coated with the 10^6 cP fluid never experience a high transient shear stress during the fracture. This may explain the results

shown in Fig. 7, where the ten-fold increase in viscosity results in only about a two-fold increase in fracture toughness.

Finally, the dissipation of a large amount of energy in the coating fluid will result in a significant temperature rise which, in turn, will affect the viscosity. An upper bound estimation of the temperature rise can be made by assuming that the energy dissipation occurs adiabatically. The total fracture energy less the energy of matrix fracture was 40.86 kJ m^{-2} (7.82 ft lb/$in.^{-1}$). This is equivalent to a temperature rise of about 361° C in the coating fluid.

While the heat loss to the matrix can be neglected due to its very low thermal conductivity. that lost to the E-glass fibres appears to be significant. Rather than conducting an exact heat transfer analysis, it is assumed that the amount of heat lost reduces the maximum fluid temperature to half of the upper bound value, namely 182° C. According to the experimental data for the Dow Corning Fluid [18], the viscosity will be reduced from the room temperature value of 10^6 cP to 1.5×10^4 cP. This reduction is an order of magnitude smaller than the change (from 10^6 cP to 1.5×10^4 cP) brought about by the shear rate effect. Therefore, the temperature effect on the shear stress is considered to be secondary in importance when compared to the strain-rate effect.

Since the energy absorption mechanism is only operative when fibres are subject to pull-out, this type of composite is useful when the matrix failure is tolerable. Therefore, this concept can best be used in applications where the amount of energy absorption is the primary design parameter. Examples for such applications may be safety related products such as highway guard rails, safety helmets and circuit breaker boxes.

Undoubtedly, the viscous coating on the fibres will reduce the static properties of the composite, such as its tensile and flexural strengths. Therefore, in engineering applications the optimization of static and impact properties may be achieved by employing both coated and uncoated fibres.

6. Conclusions

(1) The fracture toughness of fibrous composites can be improved by a factor of more than two over that of conventional composites by coating the fibres with a thin layer of viscous fluid.

(2) Energy is absorbed through the work done against the shear stress acting on the fibres during fibre pull-out.

(3) At a given strain-rate (or a velocity of impact) the shear stress can be optimized by changing the viscosity and the thickness of the coating fluid to give the maximum energy absorption.

(4) This composite is a rate sensitive material and can best be used in safety related applications where failure of the parts can be tolerated and energy absorption is the primary concern.

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