

INTERFACIAL DEBONDING OF NATURAL-FIBRE REINFORCED COMPOSITES

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ABSTRACT

The mechanisms and processes involved in interfacial debonding and fracture of natural fibre reinforced composites (NFRC) have been investigated. The Pullout theory, which depends upon factors such as lateral deformation, Poisson's effect and non-uniformity of fibres, was used to determine the fracture-toughness. It was observed that the total fracture-energy is a contribution from the surface energy, redistribution energy and pull-out energy, while the surface-energy term originates from the creation of three new surface-energy terms (fibre, matrix and fibre-matrix interface). The fracture-mode is complex, as it involves fibre-splitting, decohesion, crack-formation and propagation.

Using Eshelby's equivalent-inclusion method, the transition-stresses between the three stages of stress-strain curves were predicted. The first stage of the curve is linear and indicates that bonding of matrix-interface is essentially perfect and both phases deform elastically. The second and third stages are non-linear. During the second stage, microcracks are initiated from fibre-ends and extend to the matrix. In the third stage, the microcracks become abundant, interlink or coalesce into macrocracks, which propagate to fracture. The transition from second to third stage occurs when the crack extends further in the matrix and propagates as a Griffith-type crack.

The propagation of stress-corrosion cracks in aligned NFRC were also investigated by using the concept of strain-energy release-rate, as primary parameter, to derive values of stress-intensity factor for plane strain.

INTRODUCTION

The manufacture and use of improved natural-fibre-reinforced composites have not been commercialized on a large scale in Ghana, and there is an urgent need to improve upon the physical, chemical and

engineering properties of these natural composites, so as to facilitate easier fabrication and prolong their service-life, since there are unique advantages to be gained by developing new and sophisticated uses of the natural-fibre composites. The fibres are readily available and renewable, and their extraction and processing into composite materials require simple techniques and minimum input of energy. Due to their low density, these fibres can be used to produce composites with high specific strength. In addition, the natural fibres are non-toxic and are also environmentally degradable.

Unfortunately, the epidermis of the fibre has a thin layer of silica at its outer surface, which inhibits good bond-formation with polyester resins. Caustic soda or ammonia can remove the silica-rich layer, to improve surface adhesion. Alternatively, the fibre could be mechanically crushed, by rolling, to improve the fibre-to-resin bond, reduce porosity and air-content of the parenchyma and lumen, shear the hard epidermis, thus allowing easy penetration of the resins and other binders to the softer tissues beneath.

The binders or matrix materials used can be categorized as polymer-based, laterite-based and concrete-based (with additive of ash to decrease the weight). The common polymer binders are elastomers, wood glues, latex, natural and commercial resins. The most common natural fibre reinforced composites (NFRC) produced in Ghana are polymeric and concrete based.

By incorporating strong, stiff and brittle natural fibres into a softer or more ductile matrix, the NFRC material produced would have improved strength, fatigue-resistance, stiffness and high strength-to-weight ratio [1-7]. The matrix material transmits the force to the fibres and provides ductility and toughness, while the fibres carry most of the applied force. The fibre-reinforcing materials can be arranged in a variety of orientations, such as, uni-directional, randomly oriented, orthogonal and multiple-ply. In this paper, the complex nature of misoriented short-fibre composite interfacial-debonding and fracture is

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discussed, since the understanding of such a structure-sensitive property is essential for material design and applications.

POLYMERIC NATURAL-FIBRE REINFORCED COMPOSITES (NFRC)

The polymeric NFRC produced in Ghana are based on thermosetting phenolic, epoxy and polyester resins as binders, to produce straw-polyester composites, bagasse-phenol formaldehyde composites, coir-polyester composites, plantain/banana-polyester composites, pineapple- and sisal- epoxy composites and jute-epoxy/polyester/phenol formaldehyde composites. The fibres are produced by manual picking and are processed with the matrix by either pressmoulding or hand lay-up. Table 1 shows data on physical properties of various fibre-polyester resin composites, while Table 2 shows data on environmental degradation of fibre-reinforced composites [8]. Tables 1 and 2 deserve careful study, as they show that mechanical properties of NFRC are lower than glass-fibre reinforced composites (GFR) under normal and adverse conditions. The reason for the discrepancy can be attributed to the poor bonding of natural fibres to the matrix and this phenomenon is the main subject of this paper.

NATURAL FIBRE CONCRETE

Natural fibre-concrete (NFC) is basically made of sand, cement, fibres and water, and is used mainly in the production of corrugated sheets and pantiles for roofing in low-cost construction. The plant fibres used are stems (jute, kenaf, hemp), leaves (sisal, pineapple, raffia, palm), fruit hair (coir) and wood fibres (reeds, bagasse, bamboo).

The fibre content (1-2% by weight) is required primarily to hold together the wet mix during manufacture, to inhibit drying shrinkage-cracking and to provide early strength until the roof is installed [9]. In normal Portland-cement matrices, the fibres decay within months or a few years on account of alkali attack. This alkaline attack can be minimised by using high-alumina cement or replacing 50 % of the cement with pozzolana.

FIBRE-MATRIX INTERFACIAL DEBONDING

Interface bonding and debonding of NFRC is a

structure-sensitive property, which must be evaluated to determine the compatibility of fibres with matrices. Natural fibres are themselves fibre-reinforced materials and the microfibril angle and cellulose content determine their mechanical behaviour [10].

The effectiveness of the fibre-matrix bond is dependent on the chemical compatibility and the presence of mechanical "keying" between the fibre and matrix. An irregular surface of the fibre is more likely to enhance the efficiency of fibre-matrix interfacial bond than "smooth ones, due to greater surface-area present at the interface. However, natural fibres contain waxes and fatty acid by-products that are unlikely to form any chemical bonds between the fibre-surface and resin, thereby causing a weaker interfacial bond [1].

Natural fibres fail because, as the applied stress increases, the weak primary cell-wall collapses and decohesion of cells begins, following decohesion of cellulosic and non-cellulosic molecules (mainly through cracks and imperfections). The applied stress also causes the uncoiling as well as extension of the crystalline fibrils in the secondary cells-walls.

The fracture of NFRC, therefore, depends on the fibre-failure strain and matrix-failure strain resulting from viscoelastic behaviour. The failure of the fibres results from uncoiling of microfibril angle, accompanied by decohesion and tearing of cell walls. The fracture-mechanics of the Pullout theory and Eshelby's equivalent-inclusion theory were applied in order to study the processes involved in fibre-matrix interfacial debonding of NFRC. The work of fibre-pullout energy, which depends upon factors such as lateral deformation, Poisson's effect and non-uniformity of fibres, was used to determine the fracture toughness of the composites. The Eshelby's equivalent-inclusion method was used to predict the transition-stresses, at which interfacial debonding takes place.

Figure 1 shows typical stress strain curves of a natural fibre and NFRC. The low strength characteristics of the NFRC, compared with the single fibres can be attributed to lack of compatibility and debonding. By the rule of mixtures (ROM), the strength of NFRC is given as

Table 1. Properties of various fibre-polyester resin composites [8]

Property	Polyester Resin (0%, V_F)	Glass Reinforced Polyester Resin (unknown % fibre)	Banana Reinforced Polyester Resin (11Wt % Fibre)	Coir Reinforced Polyester Resin (9 Wt % Fibre)
Density x 10^3 (kgm^{-3})	1.30	1.50 - 1.90	1.22	1.16
Tensile strength (MPa)	41.38	241.4 - 689.6	27.96	18.61
Flexural strength (MPa)	89.69	344.8 - 862.1	64.00	38.15
Elastic modulus (GPa)	2.06	6.90 - 41.38	3.34	4.05
Impact resistance (kJm^{-2})	7.75	31.16 - 84.76	329.2	39.10
Water absorption (%*)	0.21-0.40	0.2 - 1.0	1.36	1.36
Volume resistivity ($\Omega\text{-m}$)+	1000	--	400	--

* 24 hour soaking at room temperature + At 100 V(dc)

Table 2. Environmental effect on physical properties of fibre-reinforced composites, before and after weathering for seven years [8].

Physical Property	Unweathered Jute-Fibre Reinforced	Weathered Jute-Fibre Reinforced
Bulk density x 10^3 (kgm^{-3})	1.150	1.025
Fibre content (%)	12 - 15	--
Water absorption % (25°C)		
(a) 24 hours	2.34	3.23
(b) 3 days	2.88	4.16
(c) 7 days	3.87	5.07
Water absorption % (100°C, 1 hour)	3.08	3.90
Flexural strength (MPa)		
(a) Dry	23.00	11.60
(b) 24 hour soaking	32.10	28.20
(c) 3 days soaking	42.60	19.60
(d) 7 days soaking	34.00	19.10
Tensile strength (MPa)	24.20	14.8

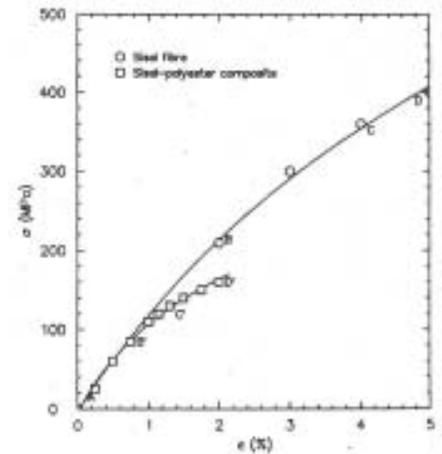


Fig. 1: Typical stress-strain curves of a sisal fibre and NFRC, showing transition stage at B,C,B' and C' respectively.

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$$\sigma_c = \sigma_f V_f + \sigma_m (1 - V_f) \quad (1)$$

where σ_c and σ_f are the composite and ultimate fibre-strength, respectively; and σ_m is the stress taken up by the matrix at the failure-strain of the fibres (ϵ_f) and is given as $E_m \epsilon_f$. Similarly, for the tensile modulus,

$$E_c = V_f E_f + (1 - V_f) E_m \quad (2)$$

where E_c , E_f and E_m are the initial tensile moduli of the composite, fibre and matrix, respectively.

PULL-OUT THEORY OF FIBRE-MATRIX INTERFACIAL DEBONDING

The Pullout theory assumes that the fibres break due to presence of flaws that are randomly distributed, and in the absence of this randomness the fibres will break in the crack-plane and no pull-out will occur [7]. The Pullout theory can be used to determine the fracture-toughness, in terms of the volume fraction, V_f , fracture critical length, l_c , diameter, d_f , tensile modulus, E_f , failure strength, σ_f , of the fibres, and the interfacial frictional shear strength (τ_f). The fracture mode of NFRC is complex, since it involves fibre-splitting, decohesion, crack-formation and propagation and other energy-absorbing mechanisms.

The fracture critical-length is calculated by assuming that the average pullout length. T_p , is half the maximum attainable pullout length and, therefore, the interfacial frictional shear strength can be obtained from the Kelly-Cottrell equation [11].

$$\tau_f = \frac{\sigma_f d_f}{2l_c} \quad (3)$$

The fracture toughness can be predicted from contributions of the total work of fracture, W_T , by summing the contributions of all sources of fracture-mechanisms involved [12]; namely, the surface-energy term, W_S , the redistribution-energy term, W_R , and the pullout-energy term, W_P :

$$W_T = W_S + W_R + W_P \quad (4)$$

where [13, 14]

$$W_R = \frac{V_f \sigma_f + d_f}{6E_f \tau_f}, \quad W_P = \frac{V_f \sigma_f d_f}{24\tau_f} \quad (5)$$

The major contributions to the overall work of fracture of the composite are W_S and W_P . The surface-energy term originates from the creation of three new surfaces, namely; the fibre fracture energy, W_P ; the matrix fracture energy, W_M ; and the fibre-matrix interface energy W_I . The energy absorbed to create these new surfaces is given by [12]

$$W_S = W_M (1 - V_f) + W_F V_f + W_{IF} \quad (6)$$

where W_M is matrix-energy term ($\sim 1.365 \text{ kJm}^{-2}$), $W_F = \frac{1}{2} \sigma_f \epsilon_f l_d$ is the fibre fracture-energy term or energy absorbed in creating the fibre-fracture surface and l_d is the average debonded length (which is assumed to be of the order of l_c whereby $l_d \approx l_c$). The interface fracture-energy term, W_{IF} for mode I debond, where the fibre-fracture strain is less than matrix strain ($\epsilon_f < \epsilon_m$), is given by

$$W_{IF} = V_f \frac{l_c}{d_f} W_I \quad (7)$$

where W_I is the interface-energy term or interfacial fracture toughness. It is difficult to determine W_I but it can be approximated to be $\approx W_M$ since the matrix adheres to fracture-surface.

Usually, the debond length l_d is in the range of l_c such that $l_d \approx l_c \approx 4T_p$. If $l_d > l_c$ the predicted energy of the debonding, W_D (which is implicit in the surface-energy term), and the energy to fracture the fibre (W_F) would be affected. The fracture-toughness or energy increases linearly with fibre-content up to $V_f = 0.24$ [7]. As volume-fraction of fibre becomes higher than 0.24, fibre-fibre interactions would increase and cause changes in τ_f and l_c , which would lead to low values of debonding energy and fracture-toughness.

The expression deduced for W_T then becomes

$$W_T = W_M (1 - V_f) + W_F V_f + \frac{V_f l_c}{d} W_M + W_R + W_P \quad (8)$$

For resins, $W_M = 1.365 \text{ kJ/m}^2$, and eqn. (8) becomes

$$W_T = 1.365 \left[1 + V_f \left(\frac{l_c}{d_f} - 1 \right) \right] \frac{1}{2} \sigma_f \epsilon_f l_c V_f + \frac{V_f \sigma_f^3 d_f}{6E_f \tau_f} + \frac{V_f \sigma_f^2 d_f}{24\tau_f} \quad (9)$$

The experimental data obtained for sisal fibre were $l_f = 0.40$ mm, $d_f = 0.02$ mm, $V_f = 0.20$, $\epsilon_f = 5\%$, $\sigma_f = 400$ MPa, $\tau_f = 0.70$ MPa and $l_c = 0.20$ mm, and these were used to calculate the contributions of each energy-absorbing mechanism, as shown in Table 3. For the NFRC, $\epsilon_f = 2\%$, $\sigma_f = 160$ MPa.

The total fracture toughness or energy of 44.044 kJm⁻² obtained by the Pullout theory is very high, as compared to typical experimental value of the work of fracture, W_{FE} , of 21.000 kJm⁻² [7] determined by the Izod Impact Test for the NFRC. The difference could be attributed to the mode of specimen-preparation and testing conditions.

ESHELBY'S EQUIVALENT INCLUSION METHOD (EEIM)

Typical stress-strain curves of a fibre and a NFRC, characterized by an initial linear region and followed by a curvature, indicating the increased rate of strains produced with increases in stress are shown in Fig. 1, which confirm the viscoelastic behaviour of the materials as a two-element Maxwell model.

The first stage of $\sigma - \epsilon$ curve is linear and indicates that bonding of matrix-interface is essentially perfect and that both phases deform elastically. The second and third stages are non-linear; the non-linearity is due to the initiation, development and extension of microcracks. During the second stage, microcracks are initiated from fibre ends and extend to the matrix. In the third stage, the microcracks become abundant, interlink or coalesce into macrocracks, which grow to a large size and propagate, leading to failure of the composite. The transitions at B' and C' in the stress-strain curves or the transition-stress between stage I (AB'), stage II (B'C') and stage III (C'D') of the $\sigma - \epsilon$ deformation curve of short fibre reinforced composites can be correlated with the interfacial debonding.

As an analytical method, the EEIM can be used to predict the transition-stress, σ_1 , between the first and second stages, where it is assumed that, while a majority of the fibres are randomly oriented as shown in Fig.2, a penny-shaped crack can be initiated from the end of a short-fibre which is aligned to the loading direction as illustrated in Fig.3. In addition, the critical stress, σ_2 , necessary for the crack to propagate into the fibre as penetration type, or into the fibre-matrix

interface as debonding type, could also be predicted.

TRANSITION STRESS, σ_1

The critical applied stress at which a penny-shaped crack arrested by adjacent fibres penetrates the fibres and the cumulative crack-density function were determined. Figure 2 shows the configuration of a randomly oriented short-fibre reinforced composite, showing both the global and local coordinates denoted by X_i and X'_i respectively. The applied stress, σ_0 , along X_3 axis, can be decomposed into three components, $\sigma_{33} = \sigma_0^2 \cos^2\theta$, $\sigma_{22} = \sigma_0^2 \sin^2\theta$, and $\sigma_{23} = \sigma_0 \sin\theta \cos\theta$, where the prime denotes the local coordinates attached to the fibre, and θ is the orientation angle of a short fibre with respect to the X_3 axis (loading direction). The short fibres shown in Fig.2 are assumed to be elongated ellipsoids of the same size and oriented randomly such that the composite material possesses transverse isotropy. The probability-density function, or the orientation factor, $g(\theta)$, is taken as [15]:

$$g(\theta) = 1/\alpha \quad (0 \leq \theta \leq \alpha); \quad g(\theta) = 0 \quad (\alpha \leq \theta \leq \pi/2) \quad (10)$$

A plot of the orientation-factor versus orientation-angle is shown in Fig.4.

The transition stress, σ_1 , between the first and second stages was calculated for the condition at which the penny-shaped crack was initiated from the end of a short fibre which is aligned to the loading direction as shown in Fig.3(a). Toya and Mitra [16] have computed σ_1 , for a completely aligned short-fibre composite. If the total free-energy of the composite, before and after the fibre-end crack is initiated, are U_1 and U_2 respectively, then to form a small penny-shaped crack at the fibre-end, the following inequality must be satisfied, $U_1 \geq U_2$. The inequality can be expressed as [13]:

$$\frac{8\sigma_0^2 (1 - \nu_m^2)(1 + \xi)c_1^3}{3E_m} \geq \pi c_1^2 \gamma_m \quad (11)$$

where c_1 is the radius of the penny-shaped crack, and E_m , ν_m , and γ_m are the Young's modulus, Poisson's ratio and surface energy of the matrix, respectively. The fibre-interaction parameter, ξ , is a function of elastic constants of the matrix and fibre, the fibre-aspect ratio and volume-fraction. It is also assumed

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Table 3. Contribution of energy-absorbing mechanisms to debonding and fracture

NFRC	W_s (kJm ⁻²)			W_P (kJm ⁻²)	W_R (kJm ⁻²)	W_T (kJm ⁻²)	W_{FE} (kJm ⁻²)
	W_M (kJm ⁻²)	W_F (kJm ⁻²)	W_{IF} (kJm ⁻²)				
0.20V _f	1.365	2.000	2.730	38.100	1.722	44.044	20.0-35.0

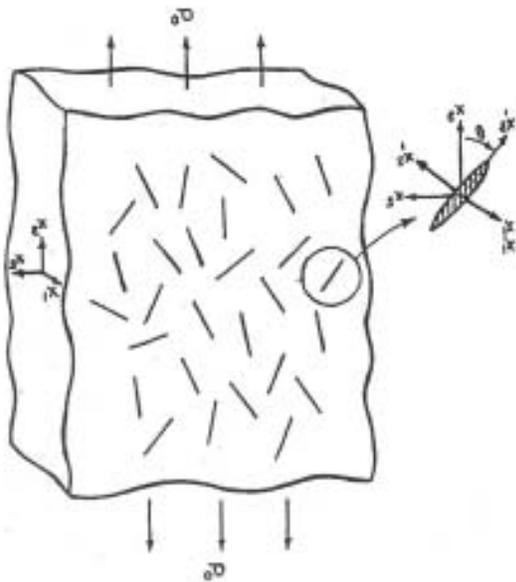


Fig.2: A schematic diagram of a randomly oriented short-fibre reinforced composite

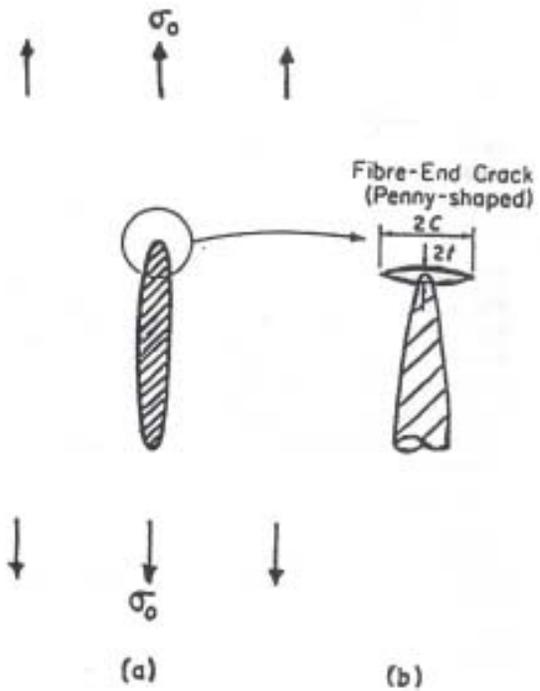


Fig.3: A crack-model for calculating the first transition stress

that the surface-energy term of the matrix-fibre interface, γ_i , is much smaller than that of γ_m . For completely aligned short fibre system,

$$\sigma_0 = \left[\frac{3\pi \gamma_m E_m^{1/2}}{8(1 - \nu_m^2)(1 + \xi_1)c} \right] \quad (12)$$

To account for the random orientation of the short fibres, and by neglecting the shear and transverse-stress components, the effective applied stress for a short fibre, at an angle θ with respect to the X_3 axis, is $\sigma_0^2 \cos^2\theta$. By replacing σ_0^2 with $\sigma_0^2 \cos^2\theta$ and, considering the contribution of all misaligned fibres, the transition stress becomes:

$$\sigma_1 = h_1 \sqrt{\frac{3\pi \gamma_m E_m}{8(1 - \nu_m^2)(1 + \xi_1)c_1}} \quad (13)$$

where h_1 is the orientation factor defined by [15]:

$$h_1 = \int_0^\alpha \cos^4\theta \frac{1}{\alpha} d\theta \quad (14)$$

TRANSITION STRESS, σ_2

The end of the stage II corresponds to the position when the crack, of radius, c_1 and c_2 , is about to propagate as indicated by the solid line in Fig.5. Therefore, the transition from second stage to third stage occurs when the crack extends further into the matrix, and the critical stress σ_2 for the penny-shaped crack of radius c_1 and c_2 to propagate, as a Griffith-type crack, is [15]:

$$\sigma_2 = \sqrt{\frac{\pi \gamma_m E_m}{2(1 - \nu_m^2)(c_1 + c_2)}} \quad (15)$$

The crack-propagation may be into the fibre (penetration type) or into the fibre-matrix interface (debonding type), and in either case may occur at a stress-level below σ_2 , and hence it is feasible for the crack to extend into the matrix at the beginning of the σ - ϵ third stage of the curve.

COMPUTATION AND DATA-ANALYSIS

Consider a 20% short-fibre reinforced resin, with the following materials properties: *Matrix*: $E_m = 2 \times 10^9$ Pa,

$\nu_m = 0.42$, $\gamma_m = 1.356$ kJm⁻²; *Fibre*: $E_f = 2 \times 10^{10}$ Pa, $\nu_f = 0.20$, $l_f = 0.40$ mm, $d_f = 0.02$ mm; and *Composite*: Interfacial energy, $\gamma_i = 2.730$ kJm⁻², $c_1 = d_f$, and $c_2 = 1.5d_f$, $g_i = E_R/E_A = 0.73$, $\alpha \approx 30^\circ$, and $h_i = 0.8407$ (where E_R and E_A are, respectively, the values of Young's modulus for composite with fibres arranged in random (R) and aligned to the loading (A)).

The predicted value of σ_1 is 133 MPa, as compared to experimental value of about 100 MPa, while the predicted value of σ_2 is 191 MPa, as compared to experimental value of about 160 MPa. As expected, the theoretical values are respectively 33% and 20% higher than the experiemntal values; and these results follow the trend observed for the discrepancy in values of the total fracture-energy obtained by the Pullout theory (44.044 kJm⁻²) as compared with the experimental value of work of fracture (21.000 kJm⁻²). The two models indicate that stronger and tougher NFRC can be produced if properties of the materials and processing-technologies could be improved, since, the discrepancy in the results could be attributed to poor bonding of matrix to fibre in the fabricated composites. Therefore, to produce NFRC of high performance, the composite must be designed with mininum imperfection, low stress concentrators, and very strong mechanical "keying" between fibre and matrix.

ENVIRONMENTAL EFFECT AND STRESS-CORROSION

Under the combined influence of stress and environment, NFRC may fail at much lower stress than in the absence of the environment. Spontaneous cracking of fibres in acids, in the absence of an extrenally applied stress, based on ion-exchange and leaching of material at crack-tip have been reported [17]. Crack-growth data can be used to predict lifetimes of components and the existence of inherent flaws in the material. The analyses of crack-growth in composite materials are based on strain-energy release-rate, G , as a primary parameter, from which values of stress-intensity factor for mode I opening, K_I , can be derived using the plane strain equation:

$$K_I^2 = \frac{EG}{(1 - \nu^2)} \quad (16)$$

where E and G are the Young's modulus and Poisson's ratio parallel to the fibres, respectively.

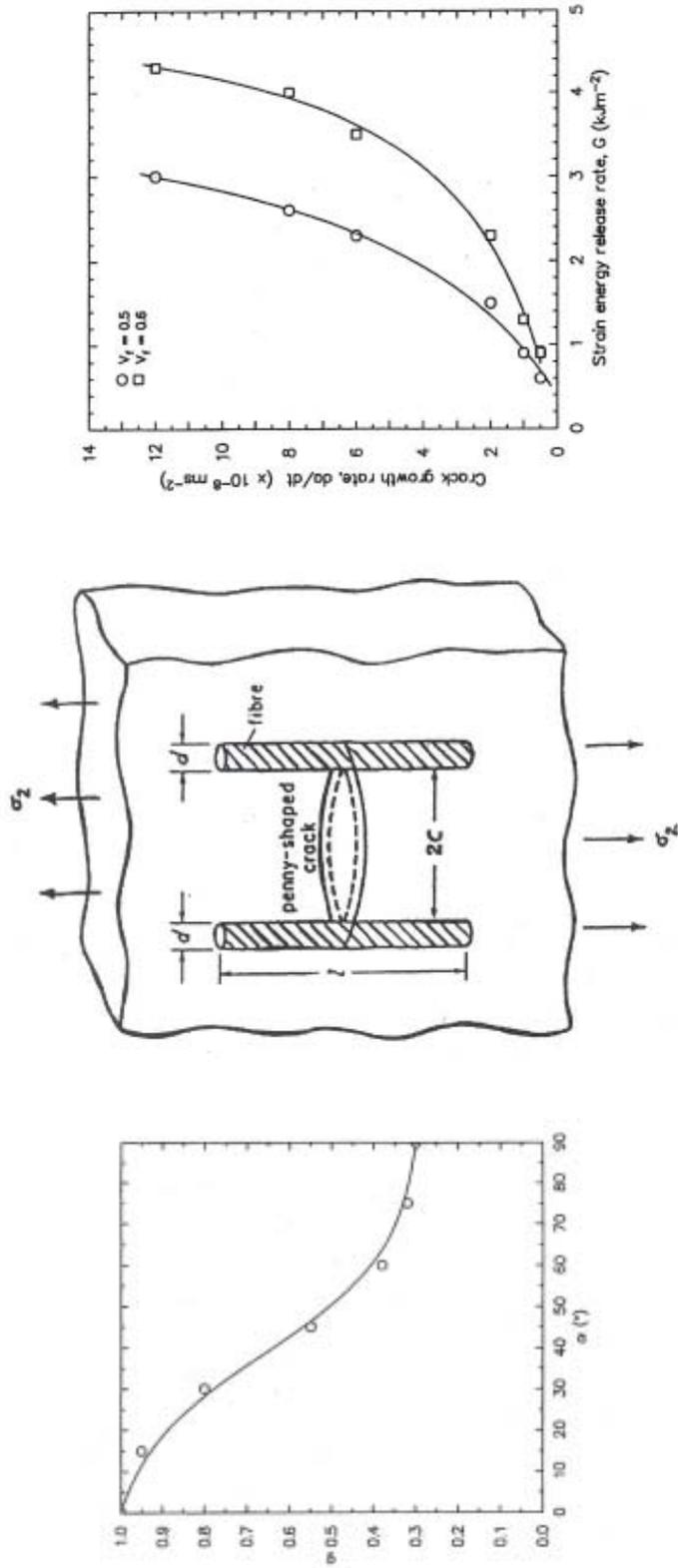


Fig.4: Plot of orientation factor versus orientation angle

Fig.5: A crack-model for calculating the second transition stress

Fig.6: Effect of G on crack growth-rate for two values V_i

In NFRC, the crack growth is irregular, as the direction of the crack-growth and the shape of the crack-front varies considerably from point to point, and these changes are due to local variations in fibre-packing and indicate that crack-growth through the resin may have a significant effect on the overall crack growth-rate [18]. The average crack growth-rates in Fig.6 can be represented by the relationship:

$$\frac{da}{dt} = CK_1^n \quad (17)$$

where C and n are constants. For $V_f = 0.6$, $n = 3.6$, while for $V_f = 0.5$, $n = 4.2$.

For direct comparison, it has been reported that da/dt in 0.6 N HCl is higher than in 1 N H_2SO_4 for a given K_1 [17]. The changes in appearance of the fracture-surface with increasing K_1 are consistent with the model for stress-corrosion cracking proposed by Hogg and Hull [18].

INTERFACIAL BONDING IMPROVEMENT

The two common methods of improving wettability of fibres in matrix are: chemical treatment and mechanical treatment. Caustic ammonia or soda will remove silica, while crushing the fibres increases surface-area for bonding. The cost of chemically removing the silica-rich layer at the surface of the epidermis, or plasticising the fibre, may outweigh any improvement in mechanical properties. To improve interfacial fibre-matrix bonding, wettability of fibre by resins must be studied. Fibres treated with solution of NaOH result in modification of surface, leading to increased wettability and minimum floatation and segregation of fibres in resins, since strong bonding improves mechanical properties. The chemical treatment causes the removal of cuticle and tyloses from the surface of the fibre, resultign in rough fibre-surface, with regular spaced pits. Untreated fibres tended to float in the polyester, whereas alkali-treated fibres are uniformly dispersed in the matrix. The effects of alkali treatment on the microstructure, surface topography and tensile strength of fibre have been investigated [19], and the results confirm that the mechanical properties of coir polyester composites are improved by soaking the fibres in NaOH (3 weight %) for more than 72 hours.

DISCUSSION AND CONCLUSIONS

The benefits of using natural-fibre reinforced composites include, but are not limited to, improved plastic shrinkage, settlement cracking, low permeability, greater impact, abrasion, and shatter-resistance of the matrix material. However, it is clear that the problems of durability are associated with compatibility of fibres and matrix-material and environmental effect on stress-corrosion resulting from interfacial debonding.

The analysis of interfacial debonding, considered from the Pullout theory and Eshelby's equivalent method, can provide basic mechanical data required for design of composite materials without expensive and time-consuming experimentation.

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