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The stress–strain behaviour of glass-fibre reinforced cement composites

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Equations describing the stress–strain behaviour of continuous, aligned, brittle-fibre–brittle-matrix composites have been modified to take account of the construction of practical chopped-strand glass-fibre reinforced cement composites. The effects on the composite strain to failure of a number of factors have been considered and some detailed comparisons made between theoretical predictions and experimental results.

1. Introduction

Glass-fibre reinforced cement (GRC) is a practical and quite extensively used example of a brittle-fibre–brittle-matrix composite material. In normal weathering conditions, or in moist environments, cementitious materials continue to hydrate over many years and it is well known that mortars and concrete gain strength with age. Similar changes occur in the strength of the cement or mortar matrix in GRC composites. However, these composites are particularly interesting examples of brittle-fibre–brittle-matrix materials because, in addition to the changes in matrix strength, there are also long-term changes in the properties of the fibre reinforcement and in the nature of the fibre–matrix bond and its environment.

Although a highly alkali-resistant glass fibre is used as the reinformcent in practical GRC, there remains a degree of interaction at the fibre–cement interface which continues slowly over many years in moist conditions. This leads to some loss in fibre strength and (probably) to an increase in fibre–matrix bond. In addition, the glass fibres are normally incorporated as multifilament bundles which initially contain considerable voidage together with regions of poor fibre–matrix contact. As ageing proceeds the cement hydration products grow into the interstices in the fibre bundles, leading to a considerable increase in the area of fibre–matrix contact [1].

The strength changes which occur during the weathering of typical alkali-resistant (AR) glass-fibre composites have been extensively studied and are well documented [2–4]. Separate studies have also been made of the direct tensile strength of glass fibres when stored in a cement environment [5] and it has been shown that the strength of the composite relates directly to the changes in fibre strength [6]. The strengths of composites are also controlled by the fibre content [7].

There is a reduction in GRC strength and this is accompanied by a significant diminution in strain to failure. This may be explained, at least partly, on the basis of the reduction in fibre strength [8] but an increasing number of workers have suggested that other factors, such as fibre bundle filling and an increase of fibre–matrix bond strength, may be of equal or greater importance in controlling the strain to failure changes in the composite [9–15]. This paper seeks to address that question by analysing some rather sparse, but detailed, tensile stress–strain data on GRC composites in relation to a slightly modified version of the theoretical model provided by Aveston *et al.* [16].

2. Theoretical background

Aveston *et al.* [16] considered the stress–strain behaviour of an idealised brittle-matrix–brittle-fibre composite material containing continuous, individual fibres aligned in the direction of loading. Provided the strain to failure of the fibre exceeded that of the matrix and the fibre content exceeded a critical value, the stress–strain curve assumed the three-part form indicated in Fig. 1 below on the basis of the following model.

Initially fibre and matrix are strained equally and stress rises linearly until the matrix cracks; there is then a horizontal region of considerable strain in which fine parallel cracking develops in the matrix, and all the load is transferred to the fibres at the crack positions. Finally the stress rises again, being carried by the fibres, and composite stiffness and strength in this region are governed entirely by fibre content, modulus and strength.

Aveston *et al.* [16] derived quantitative expressions for many aspects of the stress–strain curve, including the strain at the end of the (horizontal) multiple cracking region (ε_{mc}) and the final failure strain of the composite (ε_{cu}) in terms of fibre and matrix properties.

Practical machine-spray or hand-spray GRC composites differ from this ideal material in that fibres are normally incorporated as chopped strands, i.e. they are in the form of bundles containing ~ 200 fibres and of length 30 to 40 mm [17]. The bundles themselves are approximately randomly oriented in the plane of the GRC sheet. It is easy to show, geometrically, that the strain, $\varepsilon_{f(\theta)}$, in a fibre lying at an angle θ to the composite strain direction is less than that in an aligned fibre, being given by

$$\varepsilon_{f(\theta)} = \varepsilon_c \cos^2 \theta \qquad (1)$$

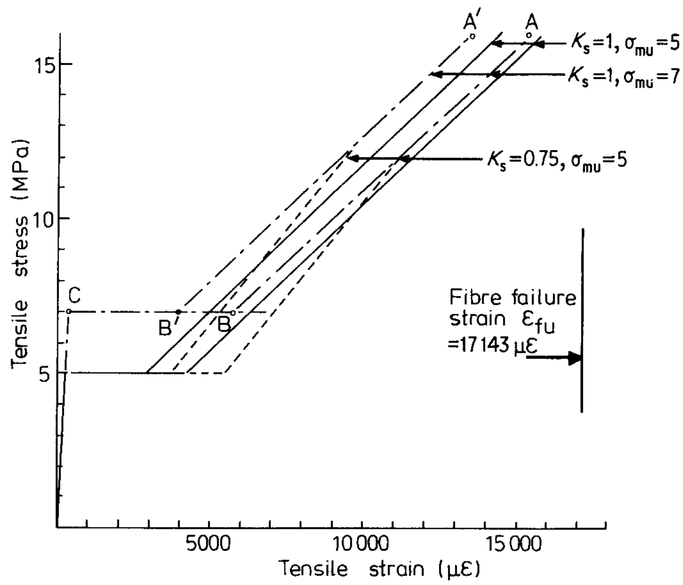

Figure 1 Typical stress–strain curves for GRC materials calculated from Equations 4, 6, 7 and 8.

where ε_c is the strain in the composite. Thus orientation of fibres at an angle to the applied load reduces their effectiveness as reinforcement, and several workers (e.g. [18, 19]) have made numerical estimates of $\sim 1/3$ for the value of an orientation factor K_0 which is used as a multiplier on the fibre content when estimating the strength or stiffness of planar random fibre composites. Other similar "efficiency factors" are often used in composites calculations to take account of effects, such as poor bonding or the use of fibres in bundles, which may also lead to a degree of ineffectiveness in the utilization of fibre properties. Oakley and Proctor [17] adopted that approach when comparing the strength, post-cracking stiffness and crack-spacing behaviour of GRC composites with the Aveston model, and it is used again here in order to compare the observed and predicted stress–strain curves in more detail.

Aveston *et al.* [16] pointed out that when the matrix cracks an additional load, equivalent to that being borne by the matrix, is placed on the bridging fibres. If fibres and matrix volume fractions are respectively V_f, V_m, fibre and matrix stiffness E_f, E_m and matrix cracking strain ε_{mu}, this leads to an additional strain $\Delta\varepsilon_f$ in the fibres at the crack position given by

$$\Delta\varepsilon_f = \alpha\varepsilon_{mu} \qquad (2)$$

where $\alpha = E_m V_m / E_f V_f$. As one moves away from the crack position the additional fibre stress is transferred back into the matrix and the additional fibre strain reduces to a value between zero and $\Delta\varepsilon_f/2$. Thus the total composite strain at the end of multiple cracking (ε_{mc}) lies between

$$\varepsilon_{mc} = \varepsilon_{mu}(1 + \alpha/2)$$

and

$$= \varepsilon_{mu}(1 + 3\alpha/4) \qquad (3)$$

for crack spacings of $2x$ or x, respectively, where x is the minimum possible crack spacing [16].

For the case of a random fibre mat of chopped strands with an efficiency factor for orientation of K_0, and for strand effects (such as poor bonding to some fibres and length effects) of K_s, the effective stiffness of the fibre mat is reduced and the additional strain is increased to

$$\Delta\varepsilon_f(\text{mat}) = \alpha\varepsilon_{mu}/K_0 K_s$$

The expressions for strain at the end of multiple cracking then become

$$\varepsilon_{mc}(\text{mat}) = \varepsilon_{mu}\left(1 + \frac{\alpha}{2K_0K_s}\right)$$

to

$$= \varepsilon_{mu}\left(1 + \frac{3\alpha}{4K_0K_s}\right) \qquad (4)$$

for crack spacings $2x$ to x, respectively. Note that if the values for K_0, K_s are obtained empirically from observed composite behaviour the most relevant data to be used will be those from the efficiency of use of fibre modulus in the post-cracking stiffness of the composite (rather than the use of fibre strength in the ultimate composite strength, see below) since the parameters used in deriving Equation 3 relate to load transfer during the early stages of crack development rather than at final failure.

At the end of multiple cracking the composite strain (as given by Equations 3 or 4) is also equal to the strain in the fibres at the crack position *less* an amount due to matrix restraint (since load is transferred back from fibre to matrix between the cracks). The amount of matrix restraint can therefore be obtained from the difference between fibre strain at the matrix cracking stress level and that given by Equations 3 and 4. As the stress on the composite is further increased, failure will finally occur when fibre strain reaches its failure value at the crack position. The composite strain will then be equal to the fibre failure strain *less* the previously defined matrix restraint. Aveston *et al.* [16] give expressions for the range of aligned composite failure strain

$$\varepsilon_{cu} = (\varepsilon_{fu} - \alpha\varepsilon_{mu}/2)$$

to

$$= (\varepsilon_{fu} - \alpha\varepsilon_{mu}/4) \qquad (5)$$

for crack spacings of $2x$ and x, respectively, where ε_{fu} = fibre failure strain.

For random chopped-strand composites the second terms are modified by $K_0 K_s$ efficiency factors as before, since they have been derived from the matrix-restraint and load-transfer expressions used to obtain Equation 4. From Equation 1, fibres lying at an angle to the load are strained less than those aligned with the load. The latter will be the first to fail. Their failure will throw additional load on the remaining fibres and rapidly initiate failure of the whole mat. The ε_{fu} term therefore does *not* need to be modified by a K_0 factor. If, however, fibre strength is used ineffectively due to poor bonding or length effects then K_s may still apply; this is now most appropriately derived empirically from ultimate strength data rather than post-cracking stiffness data.

The modified expressions for composite failure strain thus become

$$\varepsilon_{cu} = \left(K_s \varepsilon_{fu} - \frac{\alpha \varepsilon_{mu}}{2K_0 K_s}\right)$$

to

$$= \left(K_s \varepsilon_{fu} - \frac{\alpha \varepsilon_{mu}}{4K_0 K_s}\right) \quad (6)$$

for crack spacings of $2x$ to x, respectively.

Before making detailed comparisons between calculated and observed behaviour for a number of experimental composites it is worth inserting some typical values in Equations 2 to 6 to illustrate the general pattern of stress–strain behaviour expected from such composites.

Consider a GRC composite containing 4% by volume (V_f) of glass-fibre strands whose strength (σ_{fu}) is 1200 MPa and Young's modulus (E_f) is 70 GPa. Assume that these strands are randomly oriented in the plane of the GRC sheet so that $K_0 = 1/3$, and further assume initially that there is no inefficiency due to their use in chopped-strand form, i.e. $K_s = 1$. Let the fibres be incorporated in a cement or mortar matrix with a cracking stress (σ_{mu}, commonly called the bend-over point or BOP [17]) of 5 MPa and Young's modulus 20 GPa. If the composite fails by breaking of the fibres bridging one of the matrix cracks when the aligned fibres reach their failure strain, then the composite strength σ_{cu} is given by

$$\sigma_{cu} = K_0 K_s \sigma_{fu} V_f \quad (7)$$

The range of stress–strain behaviour for such a composite, computed from the above data and from Equations 4, 6 and 7, is shown by the continuous lines in Fig. 1 for the possible range of crack spacings $2x$ to x.

If the fibres in the strand are used less effectively, e.g. $K_s = 0.75$, then the composite strength and strain to failure are significantly reduced whilst the strain to the end of multiple cracking is increased, as shown by the region of calculated properties defined by the dashed lines in Fig. 1.

Two main changes are to be expected on prolonged wet ageing of the composite: the strength of the matrix (σ_{mu} or BOP) will increase and the strength of the fibres will reduce. Consider these in two stages for the case when $K_s = 1$.

(a) Assuming that the BOP increases on ageing from 5 to 7 MPa, this leads to a small increase in the strain at the end of multiple cracking (ε_{mc}) and a small decrease in the strain to failure, as shown by the chain-dotted lines in Fig. 1.

(b) Loss of fibre strength on ageing directly controls the ε_{fu} term in Equations 6 and 7. The composite failure point will move down the lines AB or A′B′ until Point B or B′ is reached. The fibres are then no longer strong enough to bear the load when the matrix cracks, i.e. their strength has fallen below a critical value of σ_f given by

$$K_0 K_s \sigma_f V_f < \sigma_{mu} \quad (8)$$

and the composite faliure strain drops suddenly to Point C. The effect of the increase of BOP on ageing (from 5 to 7 MPa in these examples) is to increase the stress and strain levels at which this sudden drops occurs. Although not illustrated in Fig. 1 it is important to note that an increase in fibre content will reduce the value of σ_{crit} calculated from Equation 8 and hence delay the onset of sudden embrittlement by the BC or B′C transition on ageing.

More detailed comparisons with experiment are given below, but it will be noted that these curves are approximately of the general form and proportion of tensile stress–strain curves for sprayed GRC (e.g. [17]) – much more so than the unmodified Equations 3 and 5 which predict a smaller horizontal multiple cracking region and a steeper post-cracking region with higher failure stress and post-cracking stiffness.

The failure strain calculated here does not include, and is not dependent on, any gross fibre pull-out which may or may not occur at final failure. Fibre pull-out may help to give a controlled crack growth once final failure has started, but it does not necessarily give rise to a strain deformation distributed through the gauge length of the specimen prior to final failure.

It can also be seen that composite failure strain is directly related to fibre failure strain and the efficiency of using fibre strain (or strength) through the first term in Equation 6, which predominates for "unaged" composites containing strong fibres. Fibre content and orientation affect the second term in Equation 6 and these factors become more significant in influencing failure strain as (or if) fibre strength and strain reduce on ageing and the second term in Equation 6 becomes comparable with the $K_s \varepsilon_{fu}$ term, when increasing fibre content leads to increased strain to failure. Other factors, such as matrix strength and stiffness, or the detailed crack spacing (x or $2x$ on the Aveston model [16]) have relatively little influence on final failure – again except in the case of an aged composite, where fibre strength is reduced to near Point B or B′ and an increase in matrix strength and/or lower fibre content may cause the critical fibre stress to be exceeded and result in a sudden reduction in failure strain from Point B or B′ to C.

3. Detailed comparisons with experiment

In order to make a detailed comparison with experiment it is necessary to have results from careful tensile stress–strain measurements in which both stress and strain at BOP, end of multiple cracking and final failure have been recorded. In addition it is necessary to known fibre content, orientation and fibre strength in the composite. Not all these data are easy to obtain accurately; some must necessarily be a matter of a certain judgement, and since tensile testing is tedious and time-consuming it is not often carried out. However, the comparison has been attempted below using some detailed information available to the author [20, 21] and some rather less detailed information taken from published papers [7, 22] for Cem-FIL fibre-reinforced, neat cement-paste matrix composites.

The test data were obtained from spray-dewatered GRC boards produced on a machine which imparted a degree of preferred orientation to the glass fibres. In many cases tests were performed on samples cut from the boards in perpendicular directions, longitudinal (L) and transverse (T), giving two components K_L and K_T of the orientation factor K_0. Where sufficient information is available values of K_L and K_T have been computed separately assuming

$$K_L + K_T = 2K_0 \qquad (K_0 \text{ taken as } 1/3) \tag{9}$$

and using values of the strength and/or post-cracking stiffness of longitudinal and transverse samples where

$$\sigma_L = K_L K_s \sigma_f V_f; \qquad \sigma_T = K_T K_s \sigma_f V_f \tag{10}$$

$$E_L = K_L K_s E_f V_f; \qquad E_T = K_T K_s E_f V_f \tag{11}$$

When possible separate estimates of efficiency for both strength, $K_L K_{s(\sigma)}$, and stiffness, $K_L K_{s(E)}$, were made and were used in Equations 4 and 6 as indicated above.

The first group of detailed comparisons is given in Table I for composite specimens taken from two boards stored in air in the laboratory for periods up to 12 months [20], from a 1-month air-stored board [22] and from two 5-year air-stored boards with markedly different fibre contents (4.4 and 8.2%, [7]). The fibre modulus was taken as 70 GPa throughout [17] and fibre strength was estimated from SIC data [5, 17, 23]. The values of σ_f and the efficiency factors used are also noted in Table I. For Board 2 neither the BOP strain nor the strain at the end of multiple cracking was recorded. To enable calculations to be carried out for this board, values of 25 GPa (L) and 19 (T), similar to the 1-month and 3-month values for Board 1, were assumed for the initial Young's modulus. For the published data of Singh *et al.* [22] and Majumdar *et al.* [7] separate L and T data were not available, so that K_s and K_0 (L or T) could not be separated by Equations 9 to 11. A value of $K_s \sim 0.6$, similar to that for Boards 1 and 2 at later ages, was assumed. For the high glass-content board of Majumdar *et al.* [7] it was assumed that the orientation factor was similar to that for the 4.4% V_f board, leading to a low $K_s \sim 0.45$.

The comments just made indicate the extent of the detailed stress–strain data required (and rarely observed or recorded) in order to make meaningful quantitative comparisons between theory and experiment. One of the objectives of this paper is to draw attention to just that point and hopefully stimulate further detailed experimental work.

An immediate observation on inspection of Table I is that the observed value of strain at completion of multiple cracking (ε_{mc}) is, with only two exceptions, below or at the lower bound of the calculated range. Secondly, the observed strain to failure ε_{cu} lies at or above the upper bound of the calculated range. Both these apparent errors imply too high a value for the term $\alpha\varepsilon_{mu}/K_0K_s$ in Equations 4 and 6. Since these disparities are most marked when measured values of the BOP strain ε_{mu} are used in calculations, and since in practice there is nearly always some non-linearity below the BOP leading to a high value for ε_{mu}, it is tempting to suggest this as a cause. However, reducing the value of ε_{mu} used would imply a proportional and compensating increase in E_m and hence in α. Careful observation of actual crack spacings at different stages of the stress–strain curve and correlation of these with theoretical estimates may help to throw light on the factors influencing the development of strain in the composite, and hence resolve these anomalies.

The other obvious way to bring the estimated values closer to those observed would be to use a larger value for K_0K_s. This would lead to disparity with the estimates of ultimate strength unless the fibre strength values used were significantly in error, but this seems unlikely since K_0K_s values were not noticeably higher when estimated from the post-cracking stiffness data (which is independent of fibre strength effects). One final possibility is that load transfer from fibre-strand to matrix between cracks is much more efficient within the small blocks of matrix between cracks than it is overall, thus giving a higher K_0K_s term in Equations 4 and 6 than that estimated from final strength or stiffness. Again, detailed crack-spacing measurements should throw light on this aspect of the behaviour of strain development in the composite.

The behaviour of a second group of water-stored or weathered composites, involving ageing behaviour, is summarized in Table II. In this case there were no records of measured ε_{mc} values to compare with the predicted values, and the absence of other observations is apparent from the table. However, from comparison with Table I it would again appear that the calculated values of ε_{mc} are probably too high.

It is clear that the theoretical estimates predict a significant reduction in composite failure strain ε_{cu} on ageing, due solely to a reduction in fibre strength and strain. However, there is some indication that the observed changes for Boards 3 and 4 are rather greater than predicted, moving from above the upper bound to just below the lower bound of calculated values. In one case (Board 4; 6 months, T) the actual strain was significantly below the predicted range, and in one other case (Majumdar *et al.* [7]; 5 years, *c*) the actual strain significantly exceeded the predicted range. In both these cases the estimated fibre strength was close to the critical value for multiple cracking calculated from $\sigma_{mu} = K_0K_sV_f\sigma_f$(crit) and an error in this parameter could explain the apparent anomaly.

In a more extensive, but more qualitative, survey of a range of GRC composites having matrices of

TABLE I

Specimen (age, type)	σ_{mu}		σ_{fu} (SIC) (MPa)	$10^6 \varepsilon_{mc}$		$10^6 \varepsilon_{cu}$		$K_0 K_s$		K_s		K_0	
	Stress (MPa)	Strain $\times 10^6$		Observed	Calculated range ($2x$ to x)	Observed	Calculated range ($2x$ to x)	(E)	(σ)	(E)	(σ)	(E)	(σ)
Board 1													
1 month, L	8	311	1245	5200	4539 to 6652	13 680	7813 to 9927	0.276	0.304	0.575	0.677	0.48	0.45
3 months, T	5.33	280	1245	6360	7739 to 11 469	10 630	4582 to 9312	0.107	0.147	0.575	0.677	0.186	0.22
6 months, L	7.04	494	1245	3390	4650 to 6728	10 300	6676 to 8754	0.254	0.251	0.569	0.609	0.446	0.416
6 months, T	5.67	565	1245	3940	7358 to 10 754	8580	4039 to 7436	0.125	0.155	0.569	0.609	0.22	0.254
12 months, L	6.82	380	1245	3920	4421 to 6441	8940	5866 to 7887	0.253	0.221	0.567	0.557	0.446	0.397
12 months, T	5.68	375	1245	4860	7198 to 10 609	9560	3084 to 6496	0.125	0.150	0.567	0.557	0.22	0.269
Board 2													
1 month, L	8.07	–	1245	–	4970 to 7294	10 850	8621 to 10 944	–	0.294	–	0.746	–	0.394
1 month, T	6.30	–	1245	–	5590 to 8219	10 000	8010 to 10 639	–	0.203	–	0.746	–	0.272
3 months, L	7.44	–	1245	–	5639 to 8310	6980	5171 to 7841	–	0.236	–	0.591	–	0.399
3 months, T	6.13	–	1245	–	6895 to 10 182	7680	3940 to 7236	–	0.158	–	0.591	–	0.267
Singh et al. [22]													
1 month	8.4	350	1245	~5000	5088 to 7459	8200	4934 to 8303	0.304	0.257	–	~0.6	–	0.44
Majumdar et al. [7]													
5 years (4.4%)	5.6	–	1245	~4300	2831 to 4131	10 100	8073 to 9373	0.334	0.248	–	~0.6	–	0.41
5 years (8.2%)	5.6	–	1245	~2300	2001 to 2885	10 200	6236 to 7120	0.253	0.184	–	~0.45	–	0.41

TABLE II

Specimen (age, type)	σ_{mu} Stress (MPa)	Strain $\times 10^6$	σ_{fu} (σ_f(crit)) (MPa)	$10^6 \varepsilon_{mc}$ Observed	Calculated range ($2x$ to x)	$10^6 \varepsilon_{cu}$ Observed	Calculated range ($2x$ to x)	$K_0 K_s$ (E)	(σ)	K_s (E)	(σ)	K_0 (E)	(σ)
Board 3 (water stored)													
1 month, L	7.79	–	1200	–	5292 to 7782	11 670	6849 to 8979	–	0.265	–	0.669	–	0.396
1 month, T	5.39	–	1200	–	5340 to 7857	9090	6399 to 8934	–	0.180	–	0.669	–	0.269
3 months, L	8.51	–	1000(784)	–	5686 to 8372	6260	4429 to 7114	–	0.268	–	0.686	–	0.390
3 months, T	6.70	–	1000(875)	–	6316 to 9322	2790	3789 to 6794	–	0.189	–	0.686	–	0.276
Board 4 (water stored)													
3 months, L	9.06	–	1000(888)	–	6419 to 9460	7660	3488 to 6529	–	0.240	–	0.670	–	0.358
3 months, T	7.82	–	1000(893)	–	6455 to 9506	6850	3471 to 6521	–	0.206	–	0.670	–	0.307
6 months, L	7.39	–	850(828)	–	5904 to 8733	1060	1276 to 4105	–	0.210	–	0.571	–	0.368
6 months, T	6.01	–	850(827)	–	5889 to 8714	480	1285 to 4109	–	0.171	–	0.571	–	0.299
Majumdar et al. [7]													
5 years, a*	9.5	–	620(646)	$\sigma_f < \sigma_f$(crit); failure at BOP strain predicted and observed									
5 years, b*	8.0	–	500(544)	$\sigma_f < \sigma_f$(crit); failure at BOP strain predicted and observed									
5 years, c*	9.5	–	620(630)	$\sigma_f < \sigma_f$(crit) but BOP exceeded		3200	980 to 2483	–	0.253	–	0.45		
5 years, d*	12	–	500(795)	$\sigma_f < \sigma_f$(crit); failure at BOP strain predicted and observed									

*Specimen a = 4.4% fibre stored in weather, b = 4.4% fibre stored in water, c = 8.2.% fibre stored in weather, d = 8.2% fibre stored in water.

varying sand content from zero to 3:2 sand:cement which give a variation in BOP stress levels of more than a factor of two, Oakley [24] investigated the time to embrittlement when stored in water at room temperature. "Embrittlement" was defined as a failure strain < 0.2%. Oakley concluded that the retention of ductility was primarily controlled by the glass strand strength rather than the bond strength, and that composites would retain ductility as long as the strand strength (σ_f) was greater than

$$\sigma_f > \text{BOP stress}/K_4 V_f$$

where K_4 is essentially equal to $K_0 K_s$ in this paper. The behaviour of samples *c* and *d* [7] at the foot of Table II supports this view.

The term $K_s \varepsilon_{fu}$ in Equation 6 is the most significant factor influencing changes in composite failure strain. Changes in fibre strength are obviously very important, and can in fact account for most of the losses in failure strain noted in Table II. Further and more detailed tensile stress–strain measurements on composites over a period of wet ageing and/or weathering are required to determine whether other factors, such as bundle filling, have any significant effect.

What that effect might be is a matter for some conjecture. Simplistically an increase in K_s might be expected on bundle filling, leading to an *increase* in composite failure strain from Equation 6 (i.e. an increase in strain to maximum stress rather than controlled crack opening). However, it is possible that the effective fibre strength may be reduced in a well filled and bonded system due to stress concentration effects at the tip of a propagating crack; and it was that view which prompted the original speculations [9–15].

In interpreting experimental results it should be remembered that hydration and bundle-filling may also be associated with BOP increases, and that matrix modifications which lead to less dense hydration products and less bundle filling may be related to low BOP stress levels as well as poorer fibre–matrix bonding. Changes in BOP level will affect the value of σ_f(crit) and hence the age of the material at which sudden loss of failure strain along the line BC (Fig. 1) occurs.

4. Conclusions

Minor modifications to the multiple cracking model of Aveston *et al.* [16] to take account of orientation and strand effects lead to tensile stress–strain curves of the general form and proportion observed with practical sprayed GRC materials. The equations provide valuable guidance on the important factors governing changes in stress–strain behaviour.

Attempts to make detailed quantitative comparisons between theory and experiment reveal a paucity of appropriate experimental data and some discrepancies in strain predictions.

The reductions in strain to failure which occur on wet ageing of GRC can largely be predicted from the expected changes in fibre strength. There is, however, an indication that other factors may be causing a somewhat greater reduction in composite strain than that predicted from fibre strength alone.

Additional detailed experimental measurements of stress–strain behaviour, its reduction on ageing, and the effects of fibre content, orientation and durability, supplemented by measurements of crack spacing against strain and microscopic studies of bundle filling and hydration, are required in order to fully assess the effectiveness of the model and the mechanisms of strain reduction.

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