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Fiber–matrix interfaces

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Abstract

The bonding mechanisms in FRC are reviewed, with reference to latest experimental and analytical studies. It is shown that bonding characteristics can not be adequately accounted for by considering only the interfacial bond strength in aligned fibers. A variety of additional factors should be considered, such as bending effects (influencing the orientation efficiency), lateral stresses and the special characteristics of the interfacial microstructure. All these should be included in a comprehensive approach to advance high performance FRC.

Keywords: fibers, interfaces, bond, fiber efficiency, pull-out, durability.

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Contents

1 Introduction

It is well established that interfacial effects in fiber reinforced composites play a crucial role in controlling the overall performance of the material. In considering interfacial effects attention should be given to two major characteristics: (i) the micro-mechanics of the physical and chemical processes taking place at the interface, (ii) the microstructure of the composite which develops at the interfacial zone. Obviously these two are related and should be analyzed together. The interfacial interactions in fiber reinforced cements (FRC) are unique and can be quite different from those occurring in polymer or ceramic composites, for two main reasons: (i) the microstructure of the cementitious composites is quite unique as its interface can not be characterized by a thin layer around the fibers, but

rather by a relatively wide interfacial transition zone where the microstructure is quite different from that of the bulk matrix. This is the result of the particulate nature of the cement matrix; (ii) in the brittle cementitious composite, the reinforcing effect of the fibers is largely materialized in the post-cracking zone; bridging across a crack can evoke a range of micro-mechanical processes which are quite different from simple pull-out. As a result of these two characteristics the bonding in cementitous composites can be weak, by order of magnitude or more smaller than other composites, and in addition to that the micro-mechanical processes at the interface are quite complex, and can not always be adequately accounted for by the simple pull-out models; other influences such as local bending across the crack and variable microstructure of the matrix should be considered.

In view of theses considerations the interfacial characteristics of FRC can not be discussed only in terms of simple parameters such as interfacial bond. If high performance composites are to be advanced it is crucial to understand and resolve the complex interactions between the fiber and matrix. This issue is perhaps more crucial than in other composites, because in the cementitous composites the interface is inherently weak and the volume of fibers that can be incorporated is limited [1,2]. The object of the present paper is to provide an overview of the special interfacial microstructures in FRC and the variety of micro-mechanical processes that should be considered simultaneously to resolve the interfacial interactions. After reviewing these issues an analysis of the influences that should be considered in high performance FRC is presented. The object of this analysis is to resolve issues that should be dealt with, and opportunities which are available, for further development of high performance FRC. The present paper provides an overview of this topic and it is interlaced with up to date information provided by contributions of the members of this group, which are acknowledged in the references.

2 The Microstructure of the Interfacial Transition Zone

It is now well accepted and documented that in cement composites, whether with fiber or aggregate inclusion, the matrix in the vicinity of the inclusion can be quite different in its microstructure than the bulk cement matrix [3]. It is characterized by a width that can be as high as 50 to 100μm, and the microstructure in it is not uniform; strictly speaking it should be treated and described in terms of microstructural gradients. The preferred term used to address this zone is the interfacial transition zone, ITZ. This is to emphasize and make the distinction with other composites (e.g. ceramic and polymer matrices), where the interface can be treated as a relatively thin boundary layer separating the fiber and the matrix, rather than a wide zone with gradients in microstructure. This difference is a significant one with regard to the control and modeling of the fiber-matrix interactions, which is usually referred to by the global term "bonding". The ITZ in FRC can assume various types of morphologies, depending on the fiber composition, geometry, surface treatment, matrix composition and processing of the composite in the production processes. In this section we will provide an overview of these microstructures, attempting to set general concepts, rather than treat different fibers individually. To achieve this goal it is best to classify the fibers into 3 types, depending on their geometry:

(i) Macro-fibers where the cross section dimension is much greater than that of the cement grains, which are typically smaller than 70μm (an average size of about 10-20 μm). This is characteristic of commercially available steel and polypropylene fibers used for concrete reinforcement, where the diameter is in the range of about 0.1 to 1.0mm.

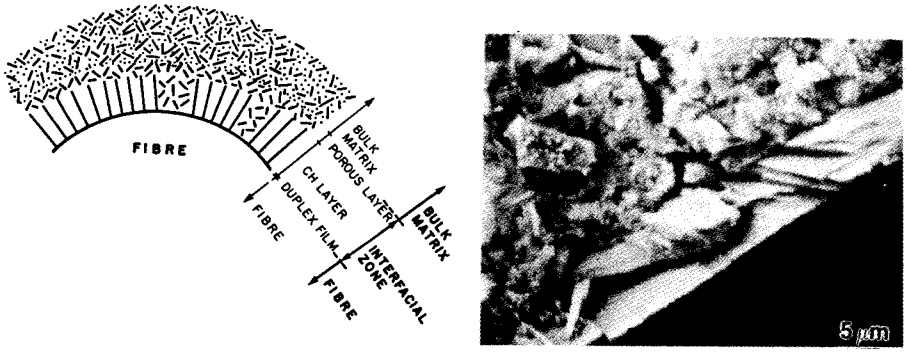
(ii) Micro-fibers having diameters equal or smaller than the cement grains. This is typical of many of the man-made filaments such as glass, carbon, polyethylene fibers, where the diameter is in the range of 5 to 40μm. It has also been suggested to arbitrarily define micro-fibers as these with specific surface area less than $20m^2/kg$ [4]. Steel fibers can also be obtained in this size range but their cross-section is rectangular with rough surface, compared to the circular and usually smooth surface of carbon micro-fibers.

(iii) Bundled fibers which are usually strands consisting of several hundreds or thousands of filaments of micro-fibers. The man-made fibers are usually produced as such bundles, in which the filaments are held together by means of a "size". It is a thin layer of polymer applied on the filament surface by a special surface treatment. Depending on the production process of the composite, the fibers can be dispersed into monofilament micro-fibers (as in (ii) above), or they can remain bundled.

2.1 Macro-fibers

The ITZ microstructure around macro-fibers is quite similar to that observed around aggregates in concrete, and it is shown in Fig.1. It is characterized by high porosity and large deposits of calcium hydroxide (CH). In both systems its formation is the consequence of a wall effect and some bleeding, resulting in inefficient packing of the cement grains in the fresh mix around the much bigger inclusion [3,5]. Thus, a water filled space tends to build up around the fiber (or aggregate) and with the progress of hydration it becomes only partially filled with hydration products; CH tends to preferentially grow in its large cavities. This microstructure can be observed by SEM and can also be quantified by backscattered electron imaging (showing the gradients of porosity at the interface) and by microhardness tests [6] (showing a weak zone extending to about 50μm (Fig.2)). Tests to quantify the interfacial fracture toughness [7] demonstrated that it is considerably smaller than that of the bulk composite matrix, suggesting that it is the weak link, as expected on the basis of the microstructural features. Observations of interfacial debonding by interferometry and fluorescent techniques [8,9] indicated that the debonding is not only at the actual interface, and it could not be described as a simple shear failure; rather microcracks and large shear displacements are observed to develop into a zone 40 to 70μm wide, which is the characteristic width of the ITZ.

In view of the microstructure of the ITZ and its weakness, there is interest in modifying its nature to enhance the bonding efficiency. Continuous curing was found to be effective in densifying the ITZ microstructure: a modest increase in the curing time from 14 to 28 days resulted in almost doubling of the interfacial bond strength and stiffness, as determined by pull-out tests [7,10]. This proportional increase is greater than that expected



(a) (b)

Fig. 1. The structure of the ITZ around macro-steel fiber: (a) schematic description, (b) SEM observation showing the CH rim and a crack arrested in the ITZ before reaching the actual interface (after Bentur [3]).

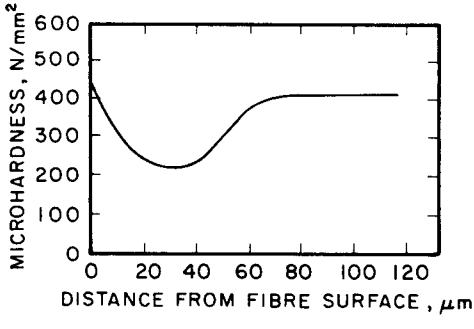


Fig. 2. Microhardness profile of the paste in the vicinity of a steel fiber (after Wei et al. [6]).

to occur in the matrix properties, suggesting that the normal maturity functions for the bulk properties may not necessarily apply to the interfacial properties.

Incorporation of silica fume is known to have a marked influence on the ITZ, as it can pack efficiently at the interface due to its small size. Indeed, additions of this admixture resulted in increase in interfacial bond strength by about 35%, but this was accompanied by reduction in the interfacial fracture toughness [10]. This is consistent with influences of silica fume on densening of the ITZ and reduction in its width: the greater density allows for a more intimate contact between the fiber and the matrix (i.e. higher bond strength) while the reduced width of the ITZ, where microcracking and slippage damage can occur, may reduce the energy absorption capacity, accounting for lower interfacial fracture toughness [10].

Changes in the ITZ microstructure, attempting to improve bond were also obtained by incorporation of polymers in the matrix, either water dispersed acrylics [6,11] or PVA [12,13]. The polymer dispersion consists of minute polymer particles, much smaller than the cement grains which can thus pack more efficiently around the fiber; later on they coalesce into a film at the interface which is intertwined also in the hydration products at the ITZ. The PVA is water soluble and can also approach closely the fiber surface. Its presence in the matrix around steel fibers resulted in the formation of a fine grained interfacial layer. The formation of this microstructure is suggested to arise from the effect of the PVA on the nucleation of CH and C-S-H at the fiber surface, as well as the presence of the polymer itself around the fiber. Its influence was found to be associated with its ability to precipitate into gel during the time period that the cement was hydrating [14]. Both types of polymer modification were found to be effective in enhancing the pull-out resistance of steel fibers by a factor of approximately 2. The PVA treatment was found to be less effective in polypropylene fibers, due probably to the low surface energy of the polypropylene [12,13].

2.2 Micro-fibers

The microstructure around micro-fibers is quite different than that of macro-fibers. It is characterized by a dense transition zone which in the SEM observations (Fig.3) does not seem to be much different than that of the bulk matrix [15,16]. The dense ITZ microstructure developed in such systems can be attributed to the fact that the size of the micro-fiber is of the same order of magnitude as that of the cement grains, thus eliminating the wall effect in the packing in the fresh mix. Also, the formulation of the mix in micro-fiber FRC is such that bleeding is largely prevented. As a result, the two main causes for the formation of the microstructural gradient at the ITZ are largely eliminated in micro-fiber composites. Yet, modification of the matrix microstructure has been reported to have an influence on the bonding characteristics; the presence of silica fume resulted in increase in pull-out resistance by about 20% [17] and in the change of the mode of failure in carbon micro-fiber composites from pull-out in composite without silica fume to fiber fracture in composite with silica fume [16]. These changes have been attributed to the formation of a denser matrix around the micro-fibers. Yet it is difficult to resolve such changes by SEM observations because the ITZ is already extremely dense even without silica fume.

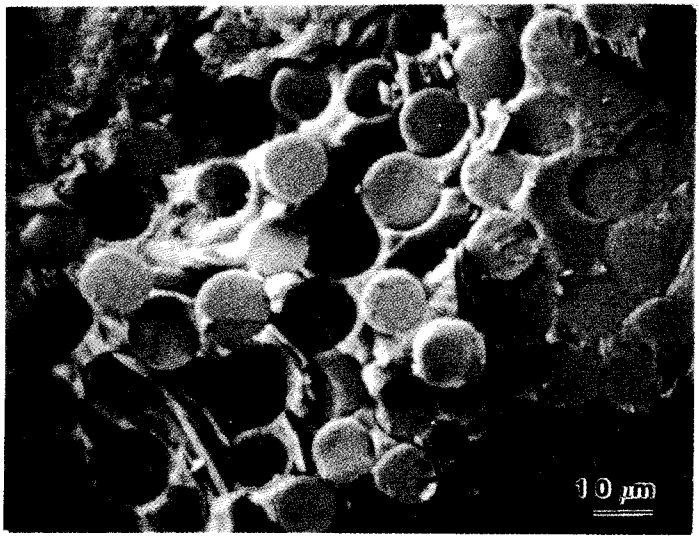


Fig. 3. The dense microstructure around a carbon micro-fiber (after Katz and Bentur [15]).

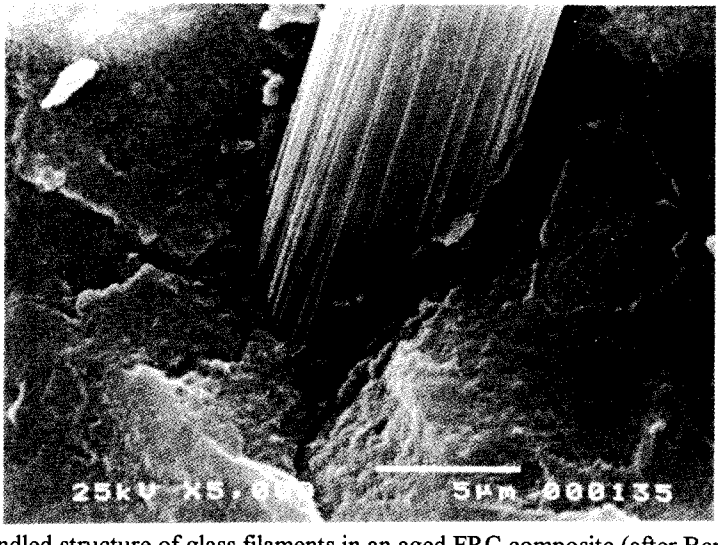


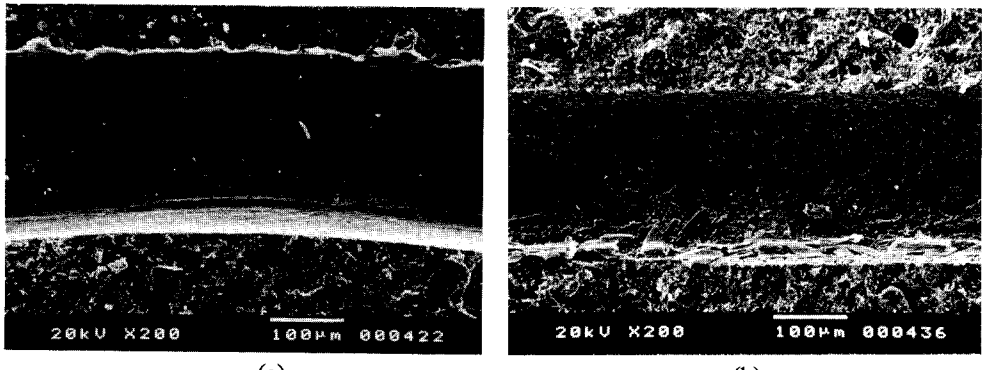
Fig. 4. Bundled structure of glass filaments in an aged FRC composite (after Bentur [18]).

2.3 Bundled Fibers

The bundled structure is characteristic to most of the glass fiber composites (Fig.4). In some instances the reinforcement with carbon, aramid and polyethylene is of this nature. The bundle consists of several hundreds of ~10μm filaments held together; the spaces between the filaments are several μm or less. Thus, cement grains can not penetrate into the filaments and as a result the microstructure obtained is different for the external filaments in the strand (which are in contact with hydration products) and the internal filaments which remain largely unbonded, in particular during the first few weeks [18]. This is a complex microstructure and bonding would be different for the inner and outer filaments. The mechanism of stress transfer is also quite complex, as stresses are transferred from the external filaments, which are in direct contact with the matrix, to the internal filaments, by means of surface contacts which may generate a friction-like mechanism. Prolonged curing results in precipitation of CH between the filaments in the strand, leading to a more rigid and brittle reinforcing unit, which may account for loss in toughness of aged glass fiber reinforced cements [18]. This microstructure and its changes with time can depend on the geometry of the strand and the type of size applied on the filaments surface. It was shown that with a bundle consisting of a smaller number of filaments, or in a bundle which is more open (as obtained in a premix production process), the growth of hydration products into the inner filaments occur at a higher rate, leading to accelerated loss in toughness during aging [19]. The modification of the sizing treatment can result in a change in the nature of the hydration products deposited in the inner filaments (causing them to be more porous, preserving the flexible nature of the strand) and it can also affect the nature of the matrix engulfing the external filaments [20]. It was shown for glass and carbon bundles that epoxy type sizing led to changes in the surrounding matrix (increased microhardness) resulting in improved bonding. The composite produced with this size showed strain hardening behavior which was not obtained when a different size was used [20].

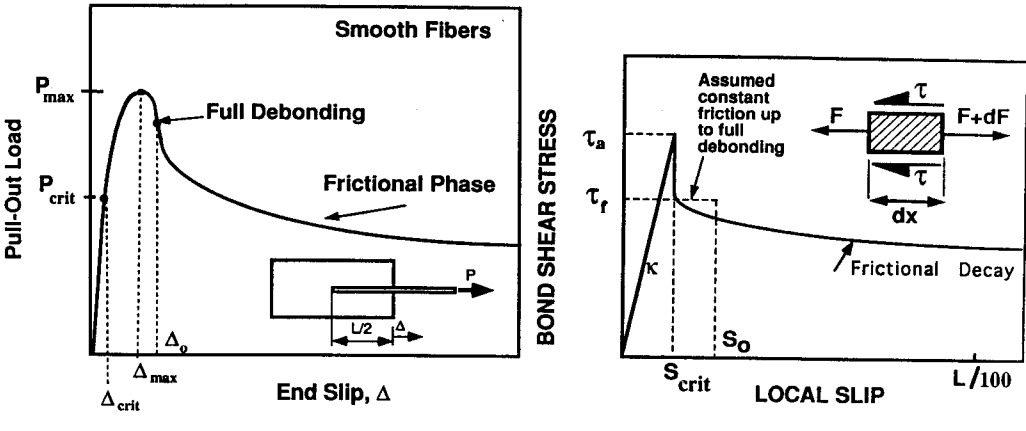
3 Failure and Damage Processes

The damage developing at the ITZ during pull-out and debonding is of particular interest in analyzing the bond efficiency and the parameters which control it. It has already been noted for macro-fiber that the damage observed is not only at the actual interface but it extends throughout the whole width of the ITZ [7,8]. Observations of the grove of pulled out fibers show a greater extent of damage in the portion close to the exit of the fiber, seen as the rupture of the thin duplex film, caused apparently by abrasion (Fig.5) [21]. This gradient in damage along the fiber is consistent with models that have concluded frictional decay with increase in slip. This however is not always the case, and fibers of low modulus may exhibit slip hardening due to different types of damage processes at the interface [22,23]. Wang et al [23] suggested that this is the result of abrasion of the fiber and accumulation of debris at the interface. Geng and Leung [22] observed that the effects occur to a larger extent in hydrophilic fiber (nylon) than in hydrophobic fiber (polypropylene). They suggested that in the hydrophilic fiber there is an increase in the effective diameter during pull-out which is



(a) (b)

Fig. 5. Abrasion of the surface of a grove during the pull-out of a polyethylene fiber: (a) The grove surface prior to the pull-out and (b) after pull-out (after Peled et al. [21]).



(a) (b)

Fig. 6. Pull-out of aligned fibers (after Naaman [28]): (a)Typical pull-out load versus slip response for steel fiber embedded in cement based matrix, (b) Typical bond shear stress versus slip relationship with frictional decay.

due to the peeling of the fiber surface and penetration of water into it, causing swelling and increase in the interfacial compression.

A special type of cracking mode at the ITZ was reported for a situation where a crack is initiated in the matrix, and as it approaches the fiber it is deviated into the weak ITZ in front of the fiber surface. This results in debonding which occurs within the ITZ rather than at the actual interface [1] (Fig. 1b).

A different type of damage has been observed in fibers which are not straight (e.g. hooked, crimped) where sometimes a fracture or yielding of the deformed part of the fiber could be seen (e.g. fracture and yielding of the hook in references [24,25]); when this does not occur the pull-out may be accompanied by considerable cracking in the surrounding matrix which can extend to a zone of a size similar to that of the deformed shape [26]. Additional structural change in such systems has been observed, which is the increase in the microhardness of the steel fiber surface, due probably to the frictional stresses applied during pull-out[27].

A third type of damage that has to be considered is the one resulting from the local bending of a fiber bridging across a crack [3]. The local flexural stresses generated may lead to some crushing of the matrix in the vicinity of the fiber. Kawamura and Igarishi [9] observed with fluorescence microscopy a damaged zone around the intersection of an inclined macro-fiber and the crack, which extended several hundreds of microns into the matrix, i.e. a distance much wider than the ITZ.

4 Pull-Out Resistance and Its Modeling

A significant portion of the studies on pull-out resistance deal with a rather simple system of a straight fiber being pulled out of the matrix, with the pull-out load being aligned with the fiber axis. Our understanding of this system is quite advanced. However, within the composite the actual pull-out process is not that simple, because most of the fibers are oriented with respect to the load. Also there is a need to take into account lateral stresses, as they may be quite different from those occurring in the simple pull-out testing. All these three topics will be addresses in this section.

4.1 Pull-Out of Aligned and Straight Fibers

The concepts of characterization and modeling of pull-out behavior were summarized by Naaman [28] and are presented in Fig.6. The modeling of pull-out behavior of aligned and straight fibers is based on analysis of pull-out curves (Fig.6a) in terms of stress transfer, assuming interfacial shear bond stresses that are elastic to start with. These stresses lead to gradual debonding and the stress transfer across the debonded interface is gradually becoming a frictional one (Fig.6b). Debonding is initiated at a load of P_{crit} where the slip is Δ_{crit} (Fig.6a). This occurs when the elastic interfacial shear stress exceeds the adhesional strength. The debonding zone is gradually extending, and at a slip of Δ_o the whole fiber is debonded and the interfacial stress transfer becomes frictional. In between Δ_{crit} and Δ_o the stress transfer includes a mixed mode of adhesional and frictional stresses. Numerous analytical treatments based on these concepts have been reported (e.g. references 7,

28,29,30,31 and earlier models reviewed in references 1 and 32), combining shear lag and frictional processes. On the basis of such models it is possible to calculate the contributions of the elastic and frictional bond components to the total pull-out resistance force and pull-out energy, as demonstrated by Hansen [31] (Fig.7).

Baggott [33] developed a novel pull-out testing system in which electrical contacts were made with the fiber and a circuit was set to measure resistivity. Initiation of pull-out was identified by increase in resistance. Intermittent increases in resistance were observed in straight, smooth uniform circular cross-section fibers at the bend over point load, suggesting that debonding started at this time. The marked increase in resistivity immediately following the peak load was indication for complete debonding. Such tests confirm the assumptions in the pull-out models.

Different assumptions have been made regarding the nature of the frictional stress transfer, treating it as a constant value throughout the whole slip, or assuming various types of post peak softening with increased slip, which is consistent with the experimental curves obtained and with microstructural observations of steel fibers. However, frictional decay is not always characteristic to all the systems. Wang et al [23] and Geng and Leung [22] reported increase in frictional resistance with increased slip of synthetic fibers. Hansen [31] reported that frictional decay occurred in steel fiber pulled from normal cementitious matrix, but this decay was eliminated when the matrix was extremely high strength cementitious binder (DSP). It has been suggested [28] that for practical purposes, within a reasonable range of slip between Δ_{crit} and Δ_o , the interfacial shear stress can be assumed to be constant. Interfacial strength values of the adhesional bond strength (τ_a) and frictional bond strength (τ_f) as calculated from the different models are presented in Table 1. The table includes also data based on simple analysis in which average bond strength, τ_{ave}, is calculated, by dividing the maximum pull-out load by the surface area of the embedded portion of the fiber. For sufficiently long embedded length, τ_{ave} is approaching the value of τ_f. It can be seen that for steel fibers τ_{ave} and τ_f are in the range of 1 to 3 MPa. Lower values are reported for the polymeric fibers.

An alternative method for modeling the debonding stage in the pull-out process, is based on fracture mechanics approach. The debonded zone is treated as an interfacial crack of length b (Fig.8) and the conditions for its propagation (i.e. debonding) are considered in terms of fracture parameters of the interface (e.g. Morrison et al. [37], Zhou et al. [38], Mobasher and Li [39] and review by Bentur and Mindess [1]) . Critical strain energy release rate calculated for the interface for macro- steel fibers [37] was found to be considerably lower than that of the matrix. This is in agreement with microstructural observations suggesting the ITZ to be weaker than the bulk matrix. In studies comparing between the two approaches, the fracture mechanics and stress calculations Stang et al. [40] and Kim et al. [38,41] reported that the two criteria (fracture mechanics and stress) are comparable. Leung [30] has shown that by defining a parameter called the effective interfacial strength in an appropriate way, strength-based and fracture-based debonding can be described by the same set of equations.

In many of these models implied assumption is made that the matrix at the interface is similar to that of the bulk, taking the same shear modulus for both. However, the

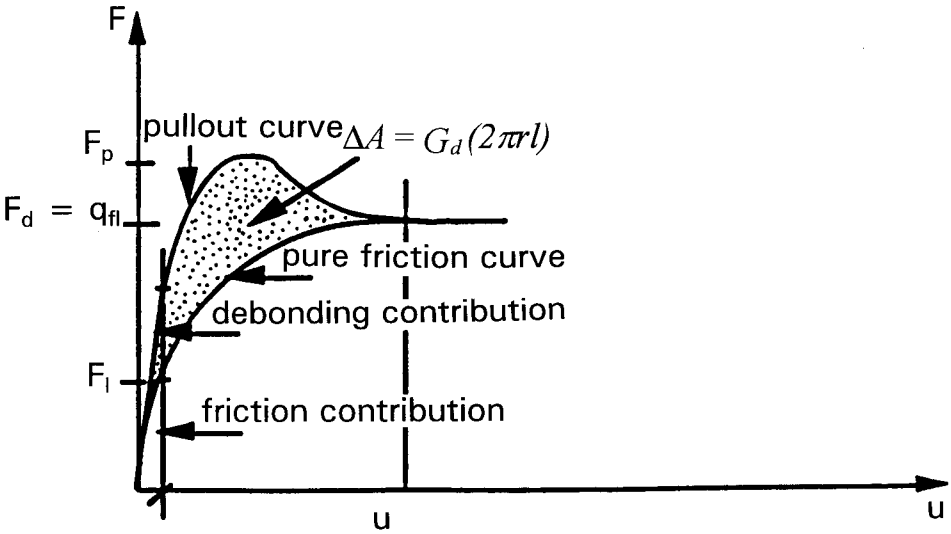


Fig. 7. Separation of the contribution of adhesion and friction to total pull-out force and pull-out energy (after Hansen [31]).

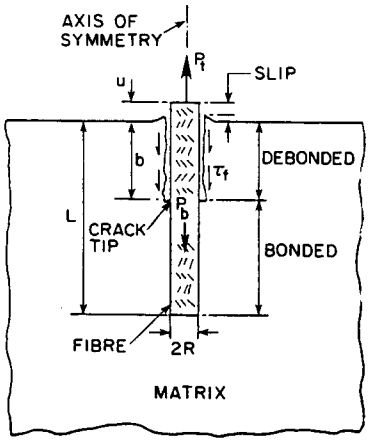


Fig. 8. Schematic description of a model used to consider the pull-out problem in terms of fracture mechanics concepts, with a propagating debonding crack of length b (after Morrison et al. [37]).

microstructural analysis already pointed the special microstructure of the ITZ. Some models attempt to consider such differences. For example, Li et al [7] calculated a parameter related to the stiffness of the ITZ, to resolve its value for different systems. Mobasher and and Li [39] extended a fracture mechanics model to describe the pull-out curves in a more comprehensive way, which included characterization of the curve in terms of adhesional and frictional interfacial bond strength, stiffness of the interface and interface toughness. They demonstrated that all of these parameters influence the pull-out behavior and when considering for example the influence of age and aging the relative change in each of them may be different (Fig.9).

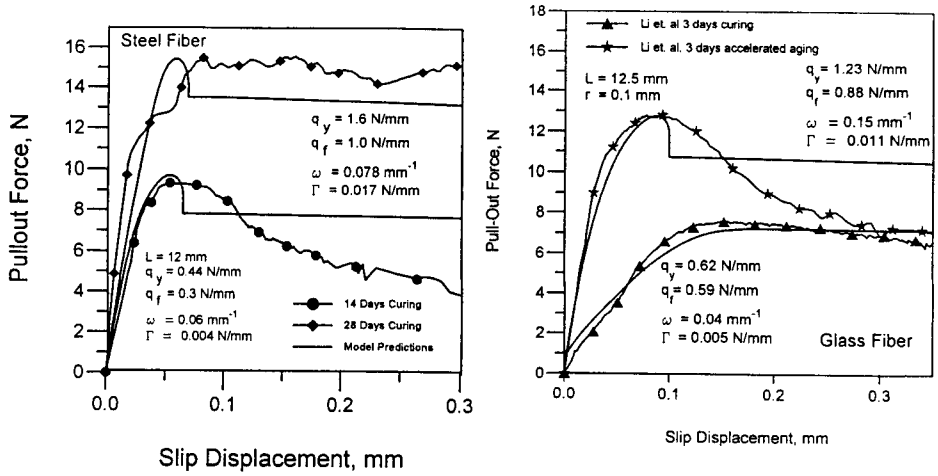
Table 1. Bond strength values of fibers in normal cementitious matrix.

fiber	fiber modulus, GPa	fiber diameter, mm	τ_{ave} MPa	τ_a MPa	τ_f MPa	reference
steel	210	0.1-1.0	-	7.4-94.7	1.2-4.9	[1]*
steel	210	0.1-1.0	0.95-4.2	-	-	[2]*
steel	210		2	-	1.2	[9]
steel	210	0.40,0.76	1.5	-	-	[12]
steel	210	0.19	1.9	-	-	[31]
steel	210		-	1.49	1.49	[34]
steel	210	0.20	-	0.78-1.12	0.43-1.05	[6]
poly-propylene	0.40	0.51	0.45	-	-	[49]
poly-ethylene	0.89	0.25	0.11	-	-	[21]
nylon	6	0.027	0.16	-	-	[35]
kevlar 49		0.012	4.5	-	-	[35]
poly-ethylene spectra	120	0.038	1.02	-	-	[35]
		0.042	0.40-0.63	-	-	[36]
carbon	240	0.010	0.52-0.66	-	-	[36]

* Review of data in the literature

4.2 Influence of Lateral Stresses and Strains

Lateral stresses and strains at the fiber-matrix interface are expected to have considerable influence on the stress transfer characteristics. Three types of lateral effects should be



(a) (b)

Fig. 9. Effect of age and aging on the pull-out curve of fibers, experimental and analytical results: (a) steel fiber, (b) glass fiber (after Mobasher and Li [39]).
q_y ,q_f - adhesional and frictional bond, respectively ($q = 2\pi r\tau$), w - parameter related to interface stiffness k ($w = (k/E_f A)^{1/2}$), Γ - interfacial toughness.

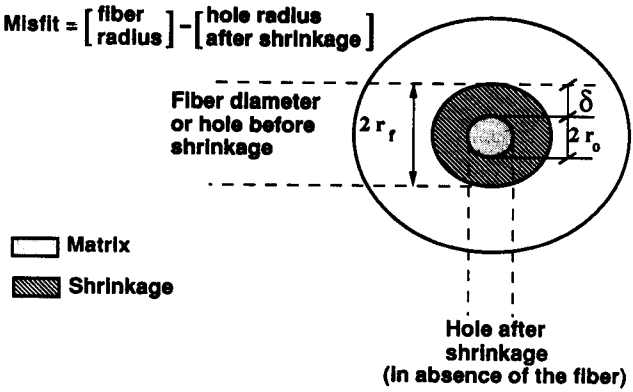


Fig.10. Fiber-matrix misfit (after Naaman et al [29]).

considered: (i) contraction of the matrix around the fiber due to shrinkage and thermal strains, (ii) Poisson effect causing excessive contraction of the fiber from the surrounding matrix, (iii) lateral strains induced during loading of the composite.

Perhaps the most important of the three is the matrix contraction around the fibers which generates a clamping stress by a mechanism which is sometimes referred to as the fiber-matrix misfit (Fig.10) [28]. It has been demonstrated that the frictional stress transfer resolved in the models outlined in section 3.1 is due largely to Columb type friction [28,42,43], as can be seen from the agreement in the frictional bond strength values estimated by Stang (43) considering the clamping stress and coefficient of friction (Table 2) and those estimated from the pull-out models (Table 1).

Table 2: Clamping stresses and frictional bond developed due to autogenous shrinkage (after Stang [43]).

fiber	fiber modulus, GPa	fiber poisson ratio	clamping stress, MPa*	coefficient of friction	bond strength, MPa
steel	210	0.3	18.5	0.08-0.2	1.5-3.7
carbon	390	0.2	17.3	0.05-0.1	0.9-1.7
poly-propylene	4	0.4	8.4	0.05-0.1	0.4-0.8

* after 500 hours of autegenous shrinkage

Of particular interest in the report of Stang (43) is the development of a novel method of in-situ testing to measure the clamping stresses developing during autogenous shrinkage. He calculated the frictional resistance by combining this data with a model developed for this purpose (Fig.11, data in Table 2). The fact that by considering autogenous shrinkage only it is possible to derive frictional resistance similar to that obtained in simple pull-out tests raises a series of questions regarding the significance of the values obtained by the various pull-out models. Perhaps the most critical one is whether the frictional resistance can be considered a basic material property, since it might be affected by environmental conditions, such as humidity and temperature changes. As the frictional resistance is perhaps the more important stress transfer mechanism there is a need for caution in using values such as those in Table 1 as design parameters. It is suggested that greater attention should be given to the environmental conditions during the pull-out testing and that the models developed for using such data should be modified to include environmental conditions. An additional issue that should be considered is the improvement in bond that is being attempted to be achieved by matrix modification. In section 1 of this paper it was demonstrated that silica fume was effective in enhancing bond and this was attributed to the changes in the ITZ microstructure. In view of the data by Stang [43] which was based on testing of a system with 10% silica fume, it may be questioned whether the improvement by silica fume is due to

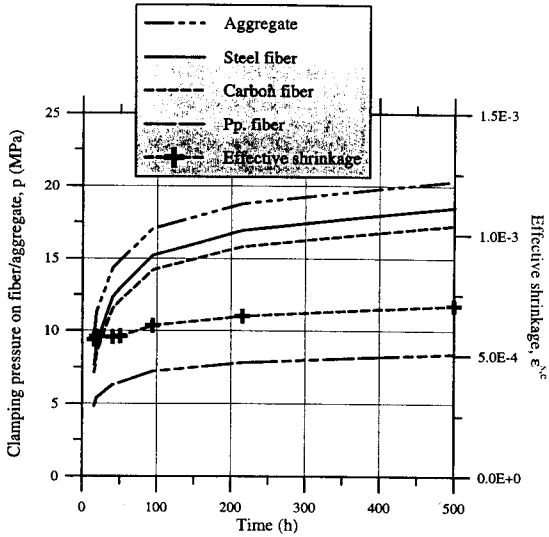


Fig. 11. Development of calculated clamping pressure for different inhomogenities along with the calculated effective shrinkage in cement paste. The elastic constants for the inhomogenities used in the calculation are summarized in Table 2 (after Stang [43]).

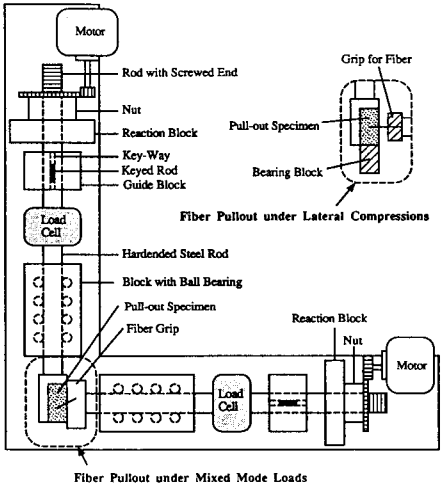


Fig. 12. Experimental set-up for mixed mode fiber pull-out test (after Leung [45]).

the microstructural changes or perhaps it is the result of autogenous shrinkage which tends to be greater in silica fume systems. This is an important question to be resolved in our attempts to enhance bonding by matrix modification.

The Poisson effect has been considered in several models and tests (Mobasher [44], review in Bentur and Mindess [1] and it is obviously of greater significance in polymeric fibers where the modulus of elasticity is low in comparison with the strength. It is of interest to note that pull-out test data reported for synthetic fibers show a trend for lower values of bond for the fibers of lower modulus of elasticity such as polyethylene and polypropylene (Table 1), which might be attributed to the Poisson effect. This is compounded by the lower clamping stresses which develop in the lower modulus fibers, as reported by Stang [43]. In view of these considerations the modulus of elasticity of synthetic fibers should be given a greater attention when being considered for reinforcement in high performance FRC, and the Poisson effect should be included in the modeling of the behavior of composites with such fibers.

The treatment by Mobasher [44] attempts to take all the above influences into account: special properties of the ITZ as a separate boundary phase, clamping and Poisson effect. In his model the interface was modeled as an independent third phase. Finite element calculation was carried out, with clamping pressure applied along the outer layer of the mesh. The state of stress at the interface started out in pure compression, and as the shear stresses increased due to pull-out loading, the compressive stress at the interface decreased due to Poisson contraction the fiber. After debonding, Columb type friction traction was introduced at the contact surface.

Very little attention has been given to the influence of lateral stresses induced during the actual loading of the composite. Leung [45] has emphasized the significance of such effects indicating that in many common failure modes, such as shear failure of beams, there are both shear and opening displacements at the crack which is bridged by the fiber. In situations involving splitting failure, as well as in shear failure, crack bridging fibers can be under significant lateral compression. In a special experimental rig developed by Leung [45] (Fig.12) the effect of mixed mode fiber pull-out was tested. It was shown that the pull-out behavior can be quite different when lateral compressive and shear stresses are applied, and the influence depends on the stress level as well as stress history. It affects both, the peak load during the pull-out test as well as the post-peak curve, as seen in the example in Fig.13.

4.3 Effect of Fiber Orientation

The orientation of the fiber has a considerable influence on the pull-out resistance and this should be taken into account since in the actual composite the fibers are rarely aligned. In cementitious composites it is important to make the distinction between three different situations: (i) fibers in uncracked composite, (ii) ductile fibers bridging the cracks in the cracked composite, and (iii) brittle fibers bridging the cracks in cracked composites. Several relations developed for the effect of orientation prior to cracking show a decrease in the pull-out resistance with increase in orientation angle; as a result the fiber efficiency falls rather sharply at the higher angles [46,47]. Orientation efficiency factors derived from such models for 2 and 3 dimensional random fiber distributions are about 1/3 and 1/9,

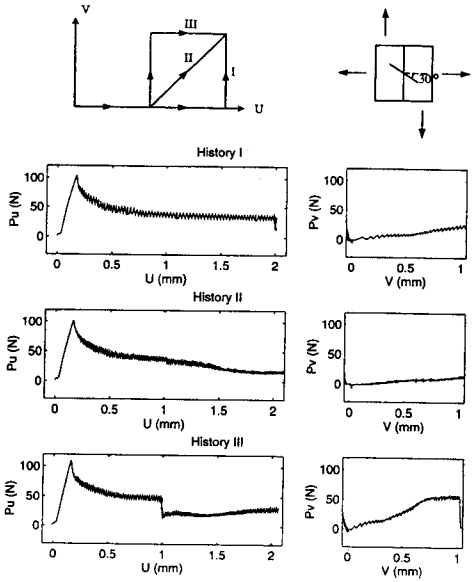


Fig. 13. Mixed mode pull-out results for aligned steel fiber in a mortar matrix (after Leung [45]).

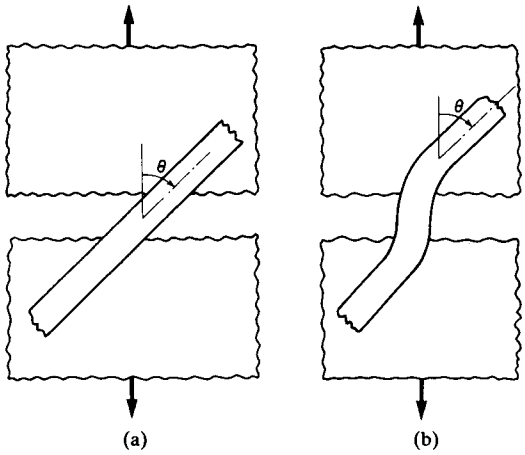


Fig. 14. The intersection of an oriented fiber with a crack assuming (a) constant fiber orientation across the crack; (b) local fiber bending around the crack (after Bentur and Mindess [1]).

respectively [46,47]. However, in a fiber bridging over cracks the influence of orientation angle can be quite different since it is necessary to consider fiber bending due to local geometrical constraints (Fig. 14). As a result of this bending, complex state of stress develops, both in the matrix and in the fiber, as seen in Fig.15. The overall behavior depends to a large extent on the balance between the rigidity of the matrix and the fiber, as seen schematically in Fig.16. If the fiber is ductile and of low modulus it will easily bend and a dowel action may be induced leading to additional pull-out resistance that may compensate for the reduced efficiency when considering only the inclination angle [35,48-53]. If the fiber is brittle and of higher modulus of elasticity, there is a build up of a local flexural stress which when superimposed on the axial tensile stress may lead to premature fiber failure and reduction in its efficiency to levels below that predicted on the basis of the influence of orientation on the fiber axial stresses only [54,55,56]. The response is also dependent on the properties of the matrix in the vicinity of the fiber and its ability to withstand the additional local flexure without cracking. The overall behavior taking into account these processes has been modeled for brittle fibers such as carbon and glass [54,55,56] and ductile ones, such as steel, polyethylene and polypropylene [35,48,49,51,52]. The micro-mechanical processes taking into account include mechanisms such as a frictional pulley effect at the exit point of the fiber (referred to as snubbing effect by Li et al [35]), beam bending on elastic foundation [48,52,54,56], matrix crushing or crumbling which occurs more readily in a brittle matrix [55,56] and energy consumed in the bending and plastic deformation in a ductile fiber [51]. Such models could account for observations of the increase in the pull-out resistance and pull-out energy of oriented ductile fibers and reduced efficiency of brittle ones (Fig. 17).

The ideal case of zero fiber stiffness shown in Fig. 16 may not be achievable in practice. Leung [49] compared results from inclined fiber pull-out tests under two conditions, (i) a free length of fiber is left between the grip and the pull-out specimen (as tested in Li et al [35], and (ii) no free length is left between the grip and the specimen. In the tests, polypropylene, a low modulus fiber, has been employed. If the fiber has no stiffness and indeed stays straight as a string, the two types of specimens should give identical results. However, the experiments indicate that for the tests with the free fiber length, the maximum pull-out load is significantly higher. The results indicate that the ideal case of zero stiffness is not necessarily achieved in practical situations. Care must therefore be taken when testing and modeling inclined flexible fibers.

Leung and Chi [48], in a refined model, attempted to predict the optimal fiber properties to achieve the best pull-out resistance considering the bending effects in ductile fibers. They concluded that optimal crack bridging behavior can be achieved with an intermediate fiber yield stress which is a function of the other composite parameters. If the yield stress is too high, an increased matrix spalling will occur, reducing both the bending and pull-out components of the bridging force.

Katz and Bentur [57,58] considered similar influences in a brittle fiber, and resolved that the orientation efficiency is improved when the fiber modulus of elasticity and diameter are smaller. This could account for observations that the fiber efficiency for low modulus pitch carbon is greater than that of PAN carbon fiber (i.e. the coefficient K in the equation

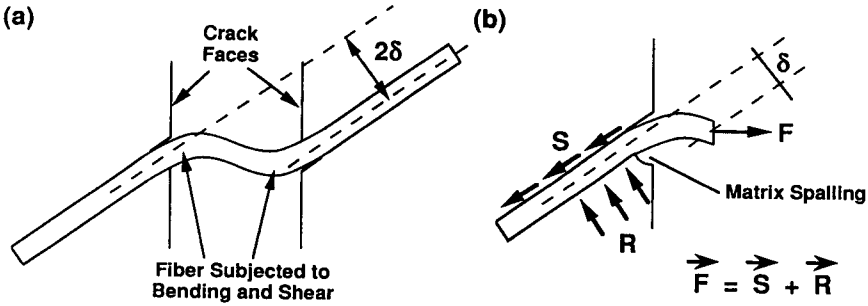


Fig. 15. Bending of fiber across a crack (a) and components of crack bridging force (b) (after Leung and Chi [48]).

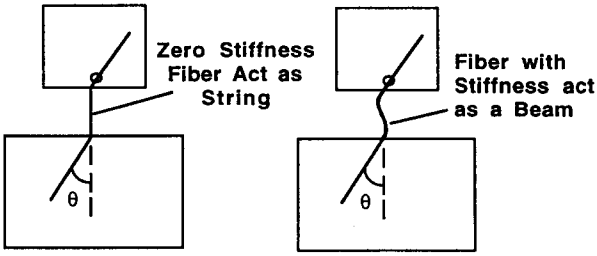


Fig. 16. Difference in behavior between fiber with zero stiffness and finite stiffness (after Leung [49]).

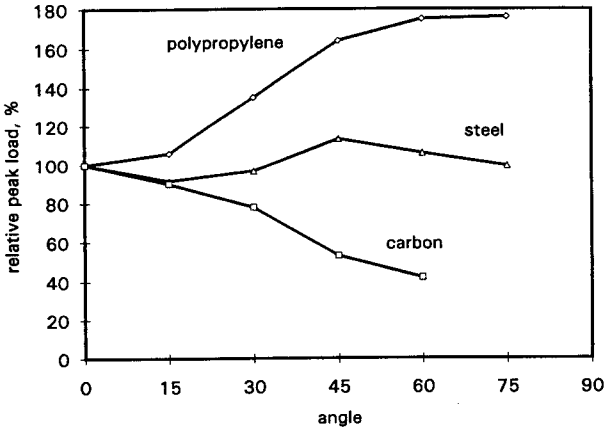


Fig. 17. Effect of orientation on the pull-out of ductile fibers (polypropylene [49] and steel [50]) and a brittle carbon fiber [55].

$\sigma_c = K\ \sigma_f\ V_f$) and the reduction in strength during prolonged aging of PAN carbon fiber composites prepared with a very dense matrix.

With these trends in mind, it should be realized that the optimization and quantification of the bonding behavior of fibers can not be based only on interfacial bond strength values and there is a need to consider also other matrix and fiber parameters.

The orientation effects for the ductile fibers as seen in Fig.17 may be considered to be of significant practical influence on the performance of the composite. Orientation efficiency factors calculated from the data in Fig.17 for a random 2-dimensional distribution of ductile fibers is about 1 or greater, which is about 3 times the efficiency of brittle fibers. Thus, the value of 1/3 usually assumed in calculations of 2-dimensional composites, based on estimates like those of references [46,47] may not be applicable to many of the FRC systems. Optimization of the orientation performance as outlined above can thus be a valuable tool for enhancing pull-out performance, perhaps as effective or more, than means based on modification of fiber surface, aiming to increase the interfacial bond strength.

4.4 Effect of Fiber Bundling and Multiple Pull-Out

In the analysis of pull-out resistance we have considered so far a rather simple case of a single fiber. However in the composite the situation can be quite different. There is a need to take into account influences due to fibers which are not dispersed as monofilaments but remain bundled in strands, or the mutual influences of monofilaments where the stress field developed around each individual filament might affect its neighbors.

4.4.1 Stress-Transfer in Bundled Fibers

As was already indicated in section 2.3, that modern man-made fibers are produced in bundled geometry (described by terms such as strand). They include amongst others carbon, nylon, aramid (trade name Kevlar), high density-ultra high modulus polyethylene (trade name Spectra) and glass. If they are not dispersed in the production process to monofilaments, their stress-transfer mechanisms can be quite different than those described above. Only few studies dealt with the modeling and measurements of the pull-out resistance in this kind of geometry. A simplified approach to analyze the results of pull-out tests can be based on calculations whereby the actual surface of filaments in the bundle which are in contact with the matrix (i.e. mostly the external filaments) is considered. Li et al [35] calculated the ratio between the exposed surface area of filaments in the bundle relative to the total surface area, assuming hexagonal close packing, and found it to decrease with increase in bundle size, from about 0.08 for a bundle of 200 filaments to about 0.04 for a bundle of 2000 filaments. Testing of the pull-out resistance of single filaments and bundles, showed that for nylon and polyethylene (Spectra) the ratio of pull-out resistance of the two geometries was much higher than expected from the model, being in the range of 0.3 to 0.5, regardless of the number of filaments in the bundle (Table 3). This is in agreement with conclusions reached for glass fiber strands, based on microscopical observations [59] (Table 3). The higher experimental value can be readily explained by opening up of the bundle allowing for hydration products to form contact with some surface of the inner filaments in the bundle. In the case of aramid (Kevlar), the ratio between the average bundle bond strength and that of the monofilament was about 0.04 (Table 3), as expected from the theoretical calculation, suggesting that this bundle is not opened up.

Table 3. Effect of bundle structure on average bond values determined by pull-out tests.

fiber	filament diameter, μm	number of filaments in bundle	τ_{single}, MPa	τ_{bundle}, MPa	$\tau_{bundle}/\tau_{single}$	refernce
nylon	27	220	0.16	0.051	0.321	[35]
kevlar	12	1000	4.50	0.198	0.044	[35]
poly-ethylene (spectra)	38	20	1.02	0.328	0.322	[35]
		40	1.02	0.502	0.492	
		57	1.02	0.505	0.495	
		118	1.02	0.352	0.352	
glass	12.5	204	1.1	0.38	0.350*	[59]

* based on microscopic observations

It should be emphasized that the stress transfer of a bundled reinforcement is quite complex and can not be accounted for in terms of average bond. Bartos [60] has proposed a bonding mechanism involving "telescopic" behavior, in which the external filaments which are well bonded to the matrix may fracture, and the internal ones will be engaged in slip. He suggested this concept for the development of a reinforcing unit that could be optimized to provide high strengthening and toughening effects. Bentur [61] suggested that the bundle structure may provide a reinforcing unit that could be flexible, and thus may be able to accommodate the bending deformations in oriented fibers. This may prevent premature fracture in brittle fibers, that when dispersed as monofilaments may break in flexure, as outlined in section 4.3. It was suggested that the embrittlement of glass fiber reinforced cement on aging may be associated with the loss of flexibility of the strand, as hydration products deposit between the inner filaments, eliminating their ability to slide one relative to the other. These mechanisms of bundle action require additional in depth quantitative study, as they have the potential for obtaining improved performance by control of the bundle structure. An example of this kind is the study by Igarashi and Kawamura [19], which showed improved durability performance of glass fiber reinforced cement when a bundle consisting of a greater number of strands is used.

4.4.2 Influence of Spacing Between Fibers

Pull-out tests are frequently carried out in a system where a single fiber is pulled out of the matrix. There is however the concern that as the spacings between the fibers become small there would be a mutual influence of stress fields, and the pull-out resistance may not necessarily be the same as that obtained by pull-out tests. Little attention has been given to

high volume (5-10%) of aligned fibers the enhancement could be as large as a factor of 2 [67-70]. The extent of improvement is obviously dependent on the magnitude of bond. In most fibers an average or frictional bond level in the range of 1 to 3 MPa can be achieved (Table 1 and the data in references [67-70]) suggesting that this is a sufficient bond for enhancement of the first crack stress in FRC with fiber contents of 5 to 10% by volume. However, in low modulus fibers, such as polypropylene, where the bond is extremely low (Table 1, discussion in section 4.2) enhancement in first crack strength was not obtained, unless surface treatments were applied that could presumably increase the bond [68].

An alternative approach to the calculation of the first crack stress is based on energy balance considerations and fracture mechanics, analyzing the conditions leading to the suppression of first crack propagation. Models based on energy balance [71], linear elastic fracture mechanics [72] and R-curve analysis [73]) were proposed. Ouang and Shah [73] and Banthia and Sheng [74] extended the R-curve approach to the modeling of the first crack enhancement in discontinuous short fiber composites. The fracture mechanics concepts reveal that the first crack strength would be expected to increase considerably with reduction in fiber spacing and enhanced bond. Both of these are very sensitive to the fiber diameter. Thus, the use of micro-fibers where the diameter is smaller by more than an order of magnitude than macro-fibers was reported to be an efficient means for enhancing the first crack strength (Fig.18) (e.g. Banthia and Sheng [75], Park et al. [76] and Katz and Bentur [15]) even at an intermediate fiber volume content (3-6%) of dispersed short monofilaments. With intermediate diameter fibers of 0.150mm, similar influences were obtained for higher contents (~10%) of random, short dispersed fibers [77]. The micro-fiber reinforcement is particularly effective for achieving improved first crack strength, not only because of their effectiveness at moderate fiber contents of about 3 to 6%, but also because they can be readily mixed at this volume content, whereas macro-fiber, because of their length (~25mm and ~3 mm in macro- and micro- fiber, respectively) can not be practically incorporated by conventional mixing at contents exceeding 1 to 2% by volume.

5.2 Strain-Hardening Behavior

The discussion in the previous section dealt with one aspect of high performance, demonstrating that the fibers may be effective in enhancing the tensile strength of the matrix [73]. It was suggested that this enhanced strength of the matrix may be carried over into the post cracking range [70], although at considerable reduced efficiency. In this zone it is the bridging effect of the fiber and their pull-out resistance which is of prime importance. Since we are dealing with a zone where crack openings occur, the pull-out resistance of the fibers may be fully materialized.

The efficiency of different types of fibers can be judged on the basis of a value which is the product of the pull-out resistance of individual fiber multiplied by the number of fibers per unit volume of reinforcement. The number of fibers is inversely proportional to the product of their cross section area, A, and their length, L, i.e. $A*L$. Thus, the efficiency of aligned fibers is proportional to $P/A*L$, where P is the pull-out resistance of a single fiber. A comparison on this basis is presented in Fig.19, based on data on Banthia et al. [17,78,79] and Katz and Li [36]. It includes straight macro- and micro-fibers as well as macro-fibers of

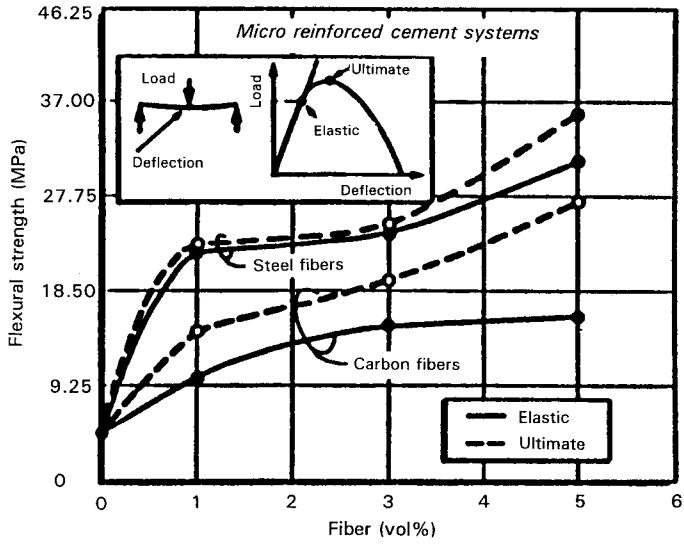


Fig.18. Flexural behavior of steel and carbon micro-fiber FRC: computed elastic and ultimate flexural strength (after Banthia and Sheng [75]).

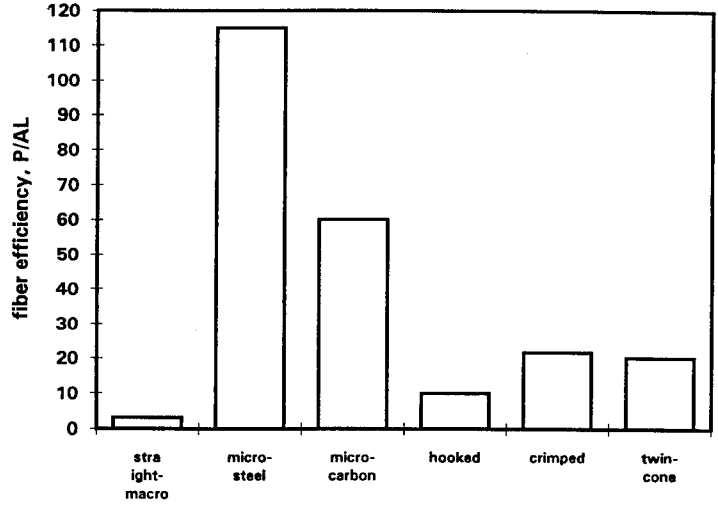


Fig.19. Efficiency of different types of aligned fibers calculated as P/A*L which is a value proportional to the pull-out resistance of individual fiber multiplied by the number of fibers per unit volume of reinforcement (based on data in references [17,36,78,79]).

deformed shapes. When considering straight fibers it can be clearly seen that the micro-fibers are more efficient, by an order of magnitude or more. Efficiency of straight macro-fibers can be improved by enhancing the bond using deformed shape. However, even with this shape the macro-fibers are not a efficient as the micro-fibers (Fig.19).

When considering high performance FRC, attention should also be given to reinforcement by continuous fibers. Although this type of reinforcement was largely used in model studies [67,68,69] it may be a viable method for production of composites by pulltrusion [80] or by fabric reinforcement [81,82], both of which enable incorporation of large volume of reinforcement. In woven fabric reinforcement special attention should be given to the crimped nature of the fiber. The crimped geometry may generate a substantially different bonding behavior, as reported by Peled et al. [21,83], with considerably improved bonding efficiency (Fig. 20). It was shown that the maximum pull-out resistance was a function of the wave length and amplitude of the waves in the crimped structure. An alternative approach to bond enhancement was offered by Li et al. [84,85] using plasma treatment. In the presence of gas plasma, hydrogen atoms are removed from the polymer backbone and replaced by polar groups. The presence of polar functional groups on the surface enhances reactivity and may thus improve the bonding with the cement matrix. The influence of this treatment on the pull-out behavior of two polymeric fibers is presented in Fig.21. In both cases the treatment resulted in considerable enhancement in the average pull-out resistance. In the case of the polyethylene fiber the average bond strength increased from 0.55 to 1.06MPa, but the shape of the curve did not change. In the polypropylene fiber the increase in average bond strength was only 20% but the shape of the pull-out curve changed considerably, exhibiting much greater slip hardening. The treatment in the polypropylene fiber resulted in a much denser interfacial matrix microstructure, and this was attributed to improved wettability of the fiber.

It might be assumed that the improved bonding efficiency in continuous reinforcement is not critical with respect to the strength of the composite, but it would have a significant influence on the strain hardening behavior, in particular the cracking patterns. Baggott and Gandhi (86) showed experimentally and theoretically that with low modulus continuous polymeric fibers, the multiple cracking process was such that no strain hardening was achieved, because of the low bond associated with the Poisson effect. Peled et al [21,68,83] improved the bond in such systems by using a woven fabric (where the fiber is crimped), or straight continuous fibers where the surface was treated. It was demonstrated that such means were effective in achieving strain hardening in composites reinforced with low modulus fibers even at an intermediate fiber content of 5% by volume. The need for a minimum bond to obtain multiple cracking and strain hardening was recently highlighted by Wu and Li [87]. This minimum value was shown to be dependent also on matrix properties. Thus optimization of the interfacial behavior with regards to trade off between strength and toughness requires attention to a range of properties, and pull-out resistance is only one of them.

In view of the discussion above there is a need to address to a greater detail the bonding mechanisms in micro-fiber and fibers of deformed shape, to analyze which of the

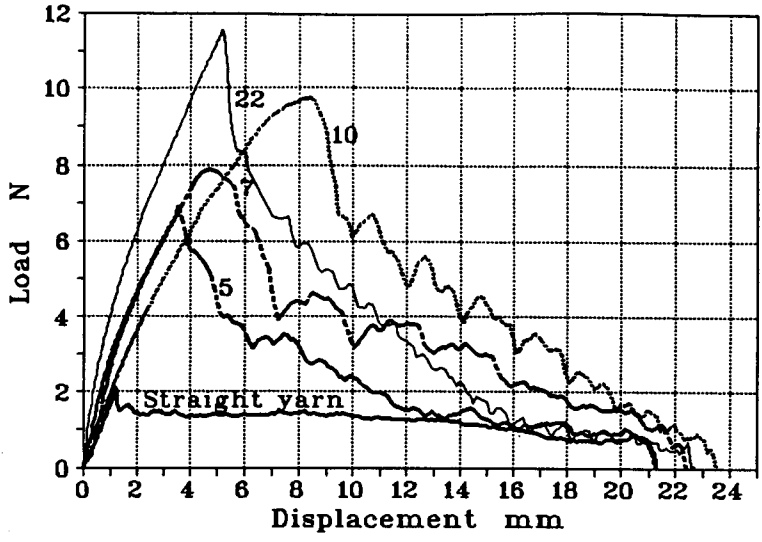
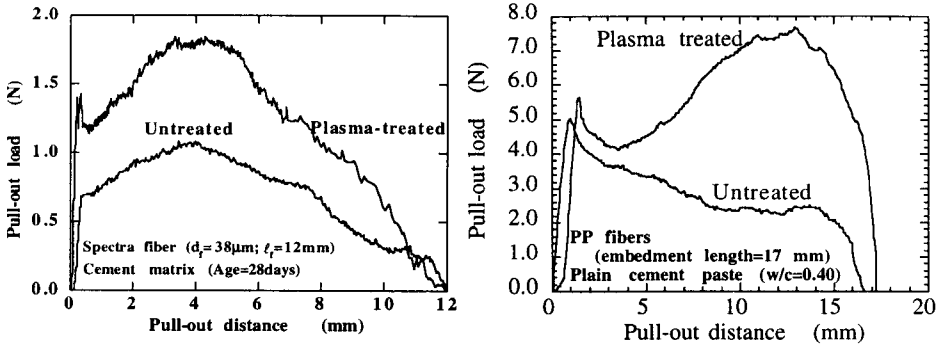


Fig.20. Effect of crimping of polyethylene fibers on the pull-out curves; the number adjacent to each curve indicates the crimp density, crimps per cm (after Peled et al. [82]).



(a) (b)

Fig.21. Effect of plasma treatment on pull-out behavior of polymeric fibers: (a) polyethylene (Spectra), (b) polypropylene (after Li et al. [84,85].

concepts addresses in section 4 are applicable here and what may be their limitations in these systems.

5.3 Bonding in Micro-Fibers

Essentially the bonding mechanisms in micro-fibers can be described according to the models presented in section 4, since the geometry of the fibers is the same as that assumed in these models. Some load-slip curves of such systems are provided in Fig. 22, and they might be analyzed according to the concepts described previously. The difference in behavior could be attributed to microstructural characteristics: The fibers which showed a mild post peak decline (the polyethylene and the 10μm carbon fiber) had both smooth surface and as a result a somewhat lower average bond strength in the range of 0.4 to 1.3MPa. This was sufficiently low to result in failure by pull-out. The slip hardening behavior of the polyethylene fiber was attributed to its softer nature; crumbs from the matrix removed by the abrasion during slip squash into this fiber and are trapped at the interface, leading to a an increase in pull-out resistance. The steel and high diameter carbon fiber (Fig. 22) had both rough surface which resulted in relatively high bond (>2.4 MPa), leading to fiber failure. In the case of the carbon fiber this failure could be prevented in a higher w/c ratio matrix, due to a reduced bond level, to average stresses of about 0.5 MPa. Thus, in this type of reinforcement special attention should be given to the possible detrimental influences of increases in bond strength that may lead to failure by fiber fracture rather than fiber pull-out. This has been reported to occur in some carbon fiber reinforced cements using the brittle PAN fiber, when the matrix is dense (Linton et al. [16]) and when it is aged (Katz and Bentur [15]). This kind of influence has been used to explain the embrittlement of GRC where the interfacial frictional bond can increase over 3.5 MPa during aging, which is of the same order of magnitude which led to fiber failure in Fig. 22. Such high bond values may be associated also with the fine size of the filaments, allowing more intimate contact with the matrix, as explained in section 2. Thus, these levels of bond and their critical influence on the transition from pull-out to fiber fracture failure is a characteristic of micro-fiber that should be taken into consideration.

An important characteristic of pull-out efficiency is the effect of orientation. There is a need to resolve whether with micro-fiber it is different than the influences discussed in section 3. Banthia et al [17,78] reported the results of such a study in steel micro-fibers, showing that the pull-out resistance is decreasing with increased orientation. This is different from the reports for macro-steel fibers [50,78] and polymeric macro-fibers [35,49](Fig.23). The difference in the behavior was attributed to the crumbling of the matrix at the exit point of the inclined fiber, causing a relative large reduction in the embedded length. In macro-fiber similar effect may occur, but the relative reduction in length is perhaps considerably smaller, due to the much larger embedded length (~20mm in macro-fiber and ~3mm in micro-fiber). Special attention should be given to the orientation efficiency of brittle micro-fiber due to risks of premature failure discussed in section 2, caused by the flexural stresses induced in bending. The reduction in the pull-out resistance of such fiber with increase in angle was also reported by Katz and Li [55] (Fig.17).

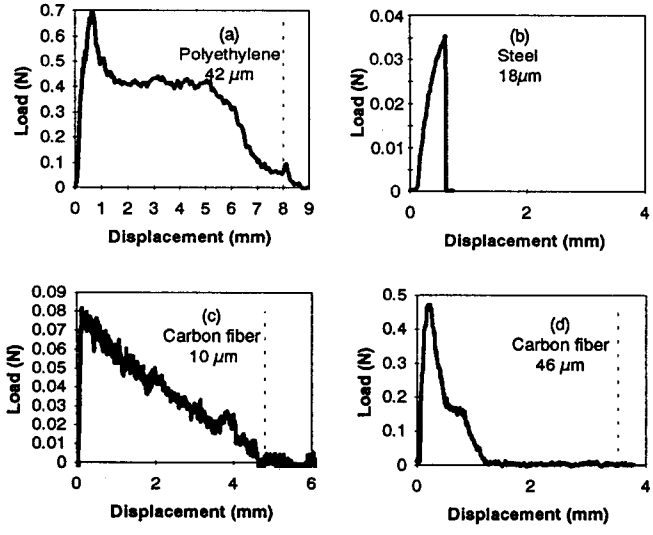


Fig.22. Pull-out curves of micro-fibers from a cementitious matrix (after Katz and Li [36]).

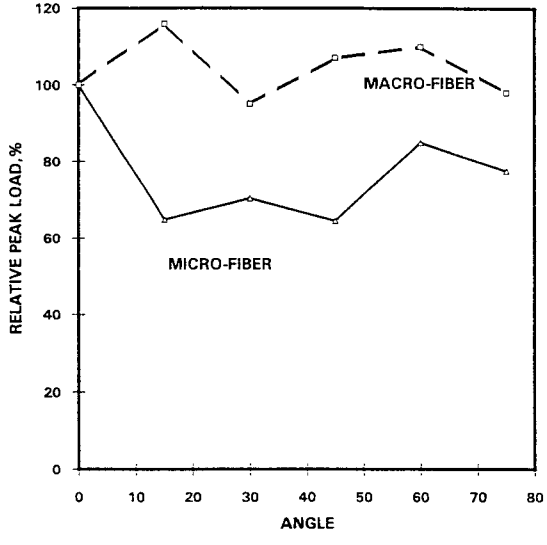


Fig.23. The influence of orientation angle on the pull-out resistance of macro- and micro-steel fibers (after Banthia and Bentur [78]).

In view of the significance of bonding in micro-fibers, and the risks involved in developing bond which is too high, there is a need for a much more in-depth investigation of the influence of lateral stresses, in particular shrinkage and autogenous shrinkage along the testing techniques and models such as that of Stang [43]. The reason for that is that such systems are usually produced with high strength matrices containing silica fume where autogenous shrinkage may be particularly high.

5.4 Bonding in Macro-Fibers of Deformed Shapes

In the case of macro-fibers the bond developed in a straight geometry is insufficient for high performance composites, and therefore most of the fibers used in practice (e.g. steel fibers) are of irregular shape to allow for mechanical (anchoring) effect. In such systems large slips are needed to mobilize the enhanced bond, and therefore the mechanical component of bond becomes effective much after matrix cracking (Naaman [28]). This is significant in terms of improved toughness, energy absorption capacity, the development of multiple cracking and the spread of plasticity in the composite at low to medium straining levels. The mechanical component of bond may be the most effective component to enhance the toughness and the energy absorption capacity.

The differences in pull-out resistance in such fibers can vary considerably (Fig.24). The micro-mechanics of such systems can not be described in terms of the models outlined in section 4. Although some attempts were made to account for mechanical bonding, they were limited to analyzing individual mechanisms such as the yield of the deformed part of the fiber [27,88] and the stress distribution along the fiber [89]. However, the problem is more complex and there is a need to consider the combined effect of these mechanisms together, and also take into account the properties of the matrix and the fiber. The sum of all of these will determine the mode of failure (fiber yielding, fiber fracture, fiber pull-out, and matrix fracture) and the overall pull-out behavior. Thus we are lacking a model that can adequately be applied to optimize mechanical anchoring in terms of the properties of the matrix and the geometry and properties of the fiber. Available experimental results suggest that a range of influences should be considered. Krishnadeve et al. [90] studied the influence of different types of steels on the pullout performance of end deformed fibers. They concluded that the strength of the steel is a more dominant parameter than its ductility (Fig.25). However, the choice of optimum steel properties depends also on the strength of the concrete. Taerwe and Van Gysel [91] reported that the strength of the matrix had a small influence on the pull-out resistance of hooked low carbon steel fiber; however in high carbon steel the increase in matrix strength was associated with increase in peak pull-out load as well as enhanced post peak resistance (Fig.26). Banthia and Trottier [92] reported the influence of the shape of the fiber on the aligned pull-out resistance and on the orientation efficiency for a range of matrix strength. The various influences are summarized in Figs. 27 and 28. Several observations should be made: (i) The fibers were considerably different in their aligned behavior (0° in Figure 27): the one which provided the highest peak load failed by fiber fracture, (ii) In some of the inclined orientations the fibers failed by fiber fracture (fiber F1 at 45° in Fig.27) although in the aligned pull-out they failed by slip, (iii) In some inclined orientation there was matrix failure (fiber F2 at 60° in Fig.28) and (iv) The orientation efficiency seems to be

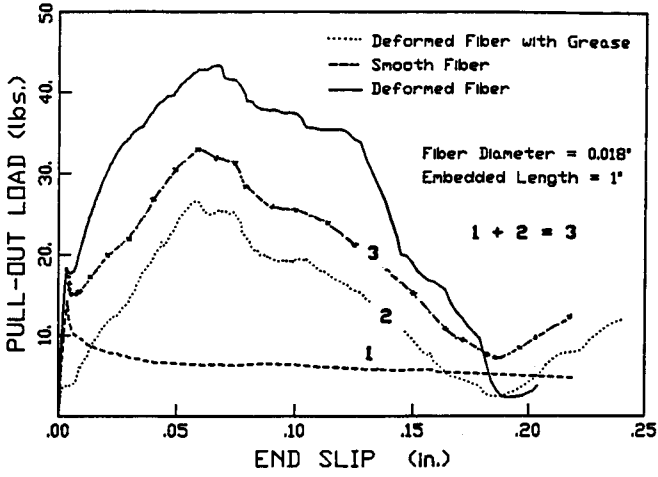


Fig.24. Effect of steel fiber geometry and surface treatment on the pull-out resistance curves (after Naaman [28]).

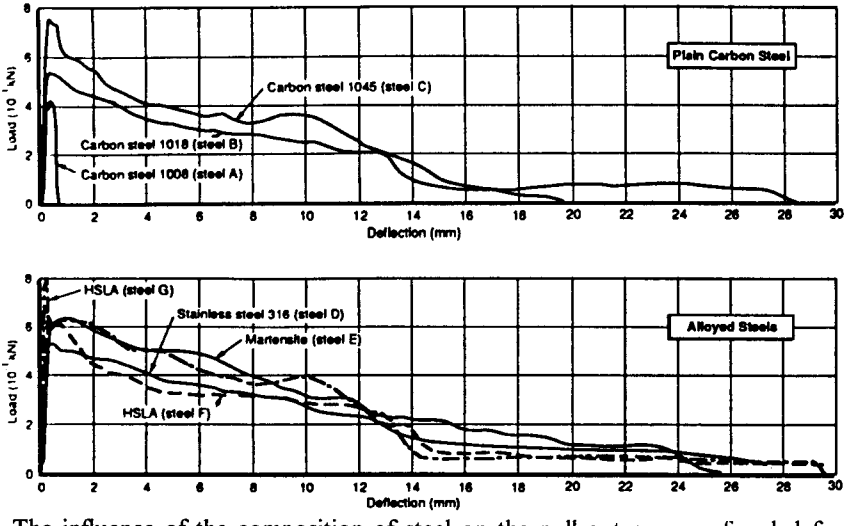
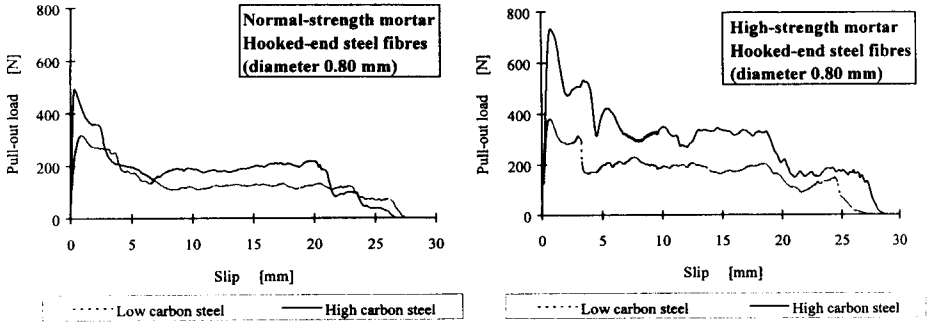


Fig.25. The influence of the composition of steel on the pull-out curves of end deformed fibers: (a) carbon steels; (b) alloyed steels (after Krishnadev et al [90]).



(a) (b)

Fig.26. Influence of matrix and steel strength on the pull-out curves of hooked fibers: (a) matrix strength in combination with low carbon steel fibers; (b) matrix strength in combination with high carbon steel fibers (after Taerwe and Van Gysel [91]).

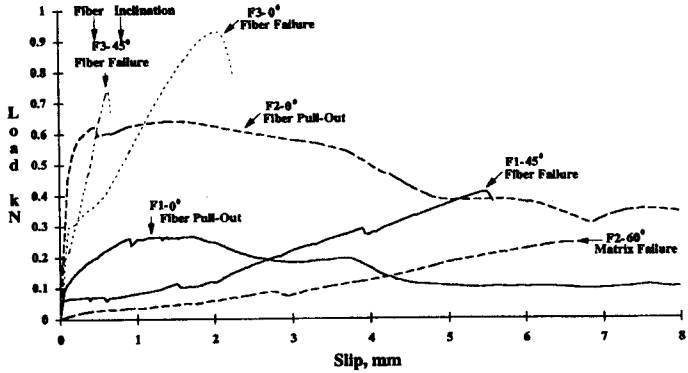


Fig.27. Some pull-out curves in high strength matrix, illustrating possible brittle failure modes (after Banthia and Trottier [92]).

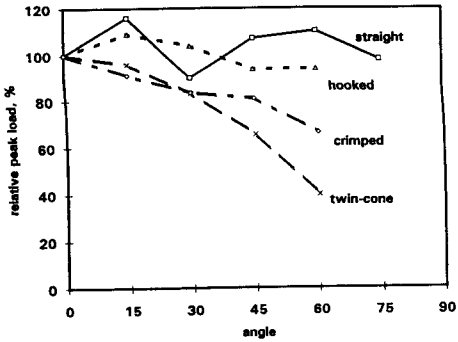


Fig.28. Influence of orientation angle on the pull-out resistance of steel fibers of different geometries (adopted from the data of Banthia and Trottier [92]).

the best in fiber F1 and the lowest in fiber F3 (Fig.28) which is just opposite to their performance in aligned pull-out (Fig.27). The overall improvement in toughness in the actual composite was obtained with fiber F1 [93] although its aligned pull-out performance was the lowest. Thus, in judging and optimizing the behavior of fibers of deformed shape, attention should be given to a variety of characteristics, including the matrix and fiber properties; judging the performance on the basis of the aligned pull-out can be misleading.

5.5 Bond Enhancement by Matrix Modification

In view of the limited bond that is achieved in straight fibers, attempts have been made to enhance the stress transfer by means of matrix modifications. Some of them, using silica fume and polymeric additives were referred to in section 2 of this paper. Data published on the bond improvement which may be achieved by such means is presented in Table 4 for macro- and micro-fibers. The improvement in most of the systems, achieved by treatments such as additional curing, incorporation of silica fume, incorporation of polymers and reduction in water/cement ratio not greater than a factor of 2. Although this improvement is quite impressive it does not provide the kind of enhancement that is obtained by using micro-fibers and fibers of deformed shape (Fig.19). Thus such changes in matrix composition can not provide the means which can drastically change the behavior of the FRC, to elevate it to the class defined in this workshop as high performance FRC. Inspite of that, the role of matrix modification can be quite important due to indirect effect. Higher bond can enable to use shorter fiber without detrimentally affecting their reinforcing influence. The combination of shorter fibers and a modified matrix, which has considerably different rheological properties, enables to incorporate in the FRC a large content of fibers. As a result, a high performance composite can be obtained either by conventional mixing or more sophisticated ones. This has been the basis for production of medium content (up to about 6%) micro-fiber FRC, such as carbon and spectra fibers [74,87,94,95] as well as high volume (>10%) short steel fiber FRC with extremely high strength and strain hardening characteristics [77,96]. Attention should be drawn to special effect reported by Katz and Li [36], in which the use of silica fume with a rough carbon fiber resulted in an increase in bond by a factor of 5 to 10, leading to fiber fracture in pull-out test (Table 4). The increase was attributed to the ability of silica fume to penetrate more effectively into the rough surface, leading to an increase in bond which is much greater than that obtained in smooth fibers.

Table 4. Effect of matrix composition and curing on bond strength

(a) Macro-fiber

diameter, mm	matrix	τ_{ave} MPa	τ_a MPa	τ_f MPa	reference
effect of age					
0.2	OPC-14days	1.45	-	0.84	[9]
	OPC-28days	2.04	-	1.2	
	OPC-14days	-	1.12	1.05	[7]
	OPC-28days	-	2.74	1.97	
effect of silica fume					
	OPC	2.04	-	1.20	[9]
	OPC+10%SF	2.51	-	1.68	
	OPC+20%SF	2.75	-	2.57	
0.19	OPC	1.95	-	-	[31]
	DSP	4.4	-	-	
effect of polymeric additive					
0.40-0.76	OPC	1.53	-	-	[12]
	OPC+ PVA	2.49-2.81	-	-	
0.5	OPC	-	1.49	1.49	[34]
	OPC +Latex	-	9.80	1.82	

(b) Micro-fiber

fiber	fiber modulus GPa	fiber diameter, μm	matrix composition	τ_{ave} MPa	reference
polyethylene (spectra)	120	42	w/c=0.50 SF= 0, 20%	0.40, 0.63	[36]
			w/c=0.35 SF= 0, 20%	0.56, 0.61	
carbon (smooth)	240	10	w/c=0.50 SF= 0, 20%	0.52, 0.66	[36]
			w/c=0.35 SF= 0, 20%	0.80, 1.29	
carbon (rough)	175	42	w/c=0.50 SF= 0, 20%	0.52, >2.44	[36]
			w/c=0.35 SF= 0, 20%	0.39, >3.02	
steel (rough)	210	60x120	w/c=0.35 SF= 10%	3.60-3.98	[17]
			w/c=0.35 SF= 20%	3.73-4.38	

5.6 Durability

Special attention should be given to the long term performance of the high performance FRC, in particular those with micro-fibers. A variety of such fibers have been developed with great attention given to their alkali resistance. Providing such resistance does not necessarily assure durability since the properties of the composite may change over time due to densening at the interface; in the case of micro-fiber such densening, even if it brings about a modest increase in strength, can lead to a change in the mode of the micro-fiber composite failure from a ductile fiber pull-out to brittle fiber fracture. This has been demonstrated by Katz and Bentur [15] to occur in the more brittle PAN carbon fiber when used in particularly dense matrix , and it has been suggested to be a major mechanism in the embrittlement of GRC [18]. A contribution to this effect may come from the bending of inclined fibers, which may lead to their premature failure due to local flexural stresses if the fiber is brittle and the matrix is sufficiently dense. This type of failure is thus dependent also on the matrix and fiber properties and fiber dimensions. Katz [58] developed a numerical model to predict such influences, and some of them are seen in Fig.29. The effect of fiber/matrix moduli on fiber bending has also been addressed by Leung and Li [56].

A variety of durability effects have been reported for cellulose fiber FRC used as asbestos replacement. These composites can be considered as high performance FRC because they can exhibit strain hardening. The cellulose fibers are essentially hollow micro-fibers. Matrix densening of the kind reported above may lead to embrittlement and loss in strength on aging [18,97,98]. However in these composites some interesting aging trends can be observed in carbonating conditions which frequently occur in natural weathering, that may lead to strength enhancement and reduction in toughness. Bentur and Akers [97] showed that such changes are due to petrification of the cellulose cell, as the cell wall and fiber lumen become "impregnated" with carbonated calcium silicates, leading to strengthening of the reinforcing unit and reduction in its flexibility (Fig.30). Soroushian [98] reported that such changes are the ones of greater concern and this conclusion was based on a study of various aging effects induced by wetting/drying, frost attack, alkali attack and conditions which may lead to thermal degradation. The matrix composition of such composites can be finely adjusted by proprietary processes to eliminate aging effects of the kind described above.

6 Conclusions

1. The pull-out resistance evaluated experimentally and analytically assuming a simple aligned fiber which is being pulled out may not always be sufficient to assess the efficiency of the fiber. There is a need to consider the influence of orientation and lateral stresses.
2. If all the processes outlined in conclusion (1) are considered it can be shown that the efficiency of the fiber can not be described only on the basis of interfacial shear strength. To optimize the pull-out resistance there is a need to consider the modulus of elasticity

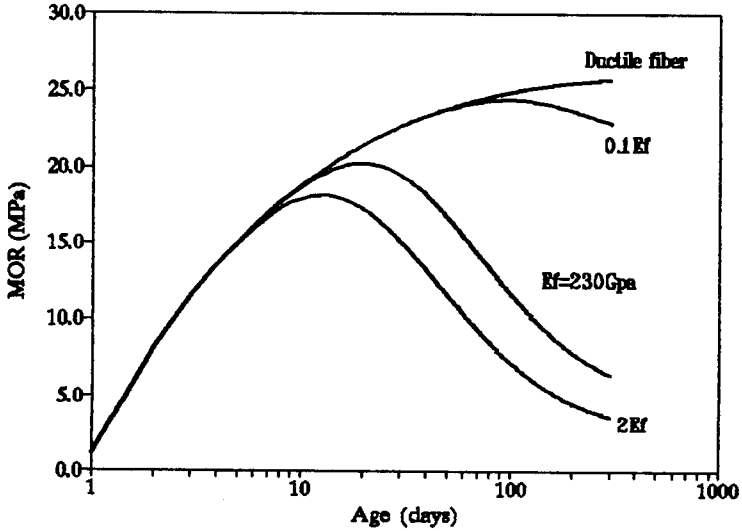


Fig.29. Aging effects in high strength cementitious matrix reinforced with ductile micro-fibers and brittle micro-fibers of different moduli of elasticity (after Katz [58]).

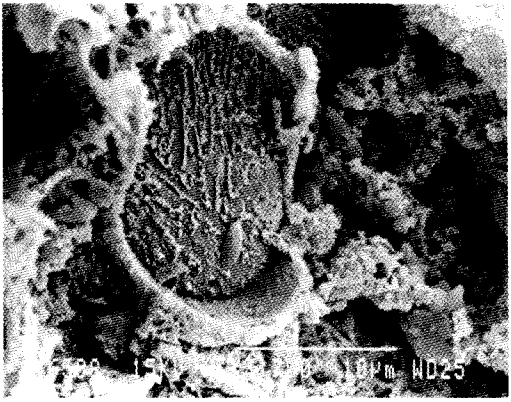


Fig.30. Cellulose fiber after "petrification" which occured during aging in carbonating conditions showing cementitious material impregnating the cell lumen (after Bentur and Akers [97]).

of the fiber and the matrix, the ductility of the fiber and the strength of the matrix, in addition to the interfacial shear bond strength.

3. Greater attention should be given to the orientation efficiency of fibers, which can be considerably different from the predictions based on conventional concepts. It is affected to a great extent by the ductility and modulus of elastictity of the fiber, and can approach 1 for ductile fibers (compared to 1/3-1/6 in random 2-D or 3-D calculated by conventional concepts).
4. Clamping stresses play a major role in controlling the frictional bond in straight fibers. The very low bond in low modulus fibers can be related to the smaller clamping stresses developed, as well as to the larger poisson effect. Such influences are needed to be considered in order to optimize the bonding behavior. For example, autogenous shrinkage should be considered in dense matrices, since its influence may be as important as that of changes in the interfacial microstructure.
5. In view of conclusion (4) special means are needed to be taken to improve the bonding of low modulus polymeric fibers if they are to be used in high performance FRC. Means which induce mechanical anchoring are particularly effective.
6. In view of conclusion (4) there is a need to assess in greater depth the possible influences of environmental conditions that may lead to changes in the clamping stresses.
7. The modification of the cementitious matrix can improve bond by as much as a factor of 2. Although impressive, this is less than achieved by reducing fiber diameter (i.e. going from macro- to micro- fiber) or deforming its geometry.
8. Micro-fibers seem to be more efficient for achieving high performance FRC, in particular when considering influences on first crack strength. They enable to obtain high performance FRC even at moderate fiber contents of 3 to 6% by volume.
9. The orientation efficiency of micro-fibers can be different than that of macro-fibers, since matrix spalling may have a greater influence on the former. There is a need for further in-depth study of such influences.
10. The bonding efficiencies in deformed fibers depend on a variety of factors, other than fiber geometry. They include amongst others the matrix and fiber strength. In addition, the orientation efficiency can be quite different, depending on fiber geometry. There is a need for a more thorough understanding of the bonding mechanisms in fibers of deformed shape, to account for such influences and optimize the geometry and composition of the fibers used.
11. Aging effects in micro-fiber FRC may result from microstructural changes occurring at the interface over time. Thus, alkali resistivity of the fibers is not the only factor to be considered when predicting durability.

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