IN SITU FORMED MULTIFILAMENTARY COMPOSITES

PART I: COUPLING MECHANISMS, STRESS EFFECTS AND FLUX PINNING MECHANISMS

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Abstract

Recent developments on in situ formed multifilamentary composites are reviewed and their superconducting and mechanical properties discussed in terms of the underlying physical mechanisms. The evidence is presented for a strong size dependence of the strengthening, flux-pinning and coupling mechanisms and, in turn, the composite normal-state and superconducting transport properties. The importance of the composite microstructure and micro-geometry is illustrated with data on Cu-Nb, Cu-Nb.Sn and Cu-YGa conductors. In particular closely spaced interfaces are shown to interact effectively with both matrix crystal dislocations and flux-line lattice, resulting in strongly anisotropic material properties. The importance of the proximity-effect coupling is discussed for Nb-Nb.Sn-based composites below the microstructural percolation threshold where the self-field critical current densities (normalized to the filament volume fraction) reached values of 1.4 x 10^10 A/cm^2. At high fields, the performance of Cu-YGa in situ composites is significantly better than that of Cu-Nb.Sn conductors, with typical normalized values of J_c of 4 x 10^5 A/cm^2 at 18 Tesla and 4.2 K. Possible use of Cu-Nb in situ composites in high-field magnet design is also discussed in view of their remarkable strength (up to 2.9 GPa at 77 K) and high normal-state conductivity.

Introduction

Since 1974, when in situ formed filamentary composites were first introduced at the Applied Superconductivity Conference, the interest in these materials has grown considerably. Composites developed over the last few years have excellent critical-current characteristics, are remarkably strong and insensitive to mechanical strain and are relatively easy to fabricate. Moreover, significant progress has been made in scaling up the processing of these materials from a few grams of laboratory samples to quantities of 10 kg and more.

The physical mechanisms which govern the properties of in situ composites with very small, closely spaced and discontinuous filaments can be quite different from those in conventional composites and are not yet fully understood. It is clear, however, in that these effects can play an important role in both transport and mechanical behavior of these materials. Some of these effects can be directly related to the composite microgeometry (filament size, spacing, dislocation, volume fraction) and are reflected in the experimentally determined flux-pinning and coupling mechanisms. Others are synergetic in nature - a mere presence of one phase can induce interactions which may radically change the properties of the adjacent phase. Since the interfacial density in in situ composites is extremely high there are typically 10^5 to 10^6 filaments/cm^2 - those effects may completely dominate the behavior of these materials.

The purpose of this paper is to review some of the work done at Harvard and elsewhere over the past few years which is pertinent to the understanding and optimization of the practical properties of in situ composites. In particular, we focus on the inter-filament coupling mechanisms and the interactions between the flux-line lattice and filament-matrix interfaces. The former is of special importance when considering composites in which the volume fraction of superconducting filaments is below the percolation threshold. Since many properties of the practical composites (e.g. stability, ac losses, bending characteristics) depend sensitively on the matrix-to-filament volume ratio, it is clear that quantitative understanding of these mechanisms is necessary in order to optimize the properties of in situ conductors. Coupling mechanisms are further considered in Part II of this paper, which deals exclusively with ac loss behavior.

Because of the rapid progress in this field no attempt has been made to summarize or tabulate results obtained by other groups working with in situ composites. The up-to-date reports from their respective investigations can be found elsewhere in these proceedings.

Sample Preparation and Microstructure

In situ techniques for preparation of superconducting filamentary composites use as a starting material two-phase alloys (Cu-Nb, Cu-V), produced either by means of powder metallurgy or by rapid cooling of a liquid solution of two components which are mutually insoluble in solid phase. Since both phases are ductile, such two-phase alloys can be mechanically processed (sawing, drawing, rolling) to large reductions in cross-sectional area until the in situ formed filaments are sufficiently small (< 100 Å). In order to form high-field A-15 compounds, Sn (or Ga) can be added either during the initial mixing process or plated on the already drawn wires and diffused into Nb (or V) filaments during the final, high-temperature anneal.

The optimum amount of tin (or gallium) necessary to convert niobium (or vanadium) filaments into A-15 compounds by external diffusion is considerably higher than the stoichiometric amount. Consequently, the excess tin (gallium) remains in the matrix which has typically resistivity ratio of 2 or less. This high matrix resistivity is, in fact, desirable in some cases, such as ac conditions, since it reduces interfilament coupling and in turn ac current losses. For high matrix conductivity and improved stability under dc conditions the bronze route appears to be more advantageous. In this latter case, the appropriate amount of tin is added already during the initial melting process, resulting in a two-phase Cu-Sn-Nb alloy. The drawback of this approach is the need for frequent intermediate anneals due to the matrix work hardening, particularly at high reduction of the cross-sectional area. However, since tin is uniformly distributed in the matrix and diffusion distances are very short (on the order of the interfilament spacing) less tin is needed and the average tin concentration in the matrix after annealing may be as low as a few hundredths of a percent. In fact, in the cleanest matrices fine particle scattering from the filament-matrix interfaces becomes the dominant scattering mechanism.

A characteristic feature of all composites after mechanical reduction is a dense distribution of ribbon-like filaments. Their shape can be related directly to deformation mode during composite formation. Bcc crystals are known to develop <110> fiber texture in which only one of the four <110> directions is deformed, as schematically illustrated for four <110> fiber texture in which only one of the four <110> directions is deformed, as schematically illustrated for four
short-filament conductors, the relative strength of the coupling is less, and the critical current may indeed be limited by the onset of superconductive phase slippage between adjacent filaments. Further studies of this phenomenon are certainly warranted, since recent measurements\(^9\) and qualitative theoretical arguments\(^10\) both suggest that ac losses are reduced in composites with shorter filaments and lower filament volume fractions, which reduce the interfilamentary coupling. This observation indicates that wire structure might be optimized not by maximizing interfilamentary coupling but rather by making it no stronger than necessary to give satisfactory \(J_c\) under dc conditions.

Turning back to the long-filament in situ composites and their potential usefulness, one is particularly interested in whether their resistance will be zero or at least close enough so that they could be used as a magnet wire. Given the notion of a percolation threshold\(^11\) at some volume fraction (\(~15\%\) in 3-dimensions, \(~50\%\) in 2-dimensions) one might expect strictly zero resistance above that fraction, and finite (but small) resistance at smaller volume fractions. But the effective superconducting volume fraction will be larger than the nominal one, and will depend on temperature, current density, and magnetic field because of the "proximity effect", which allows superconducting electrons to diffuse roughly a normal coherence length \(\xi\) from the superconducting filaments themselves. This length will be of the order of a fraction of a micron under typical conditions, and hence will become important when filament diameters are reduced to approach a similar scale. But because the coherence energy of the wire function only falls exponentially, as \(\exp(-L/\xi)\), some superconductive coupling will continue to larger distances, until this energy falls below the thermal noise level. Thus, at low temperatures and with small measuring currents, one expects to find true perfect conductivity, as in a bulk superconductor, and one does, even for nominal superconducting volume fractions well below the percolation threshold.

But the technically interesting case is one of high current densities, \(T \gtrsim 4\,K\), and in the presence of strong magnetic fields. All these influences reduce the proximity-effect coupling, shrinking the effective superconducting volume fraction closer to the nominal one. Thus, it is useful to consider first a simple "worst-case" limit, in which one assumes that the proximity-effect coupling is completely suppressed, and that the superconducting volume fraction \(f_s\) consists of a collection of long thin cylinders of length \(L\) and diameter \(d\), formed from individual precipitate grains by the drawing process, and hence aligned along the wire. (In Part II of this paper, symbol \(f_s\) is replaced with \(\lambda\) in order to be consistent with the standard notation in the ac loss literature.) The rest of the volume is assumed filled with a completely normal Cu matrix.

For a qualitative understanding, it is convenient to idealize the deformation in the drawing process so that, if the area reduction ratio is \(R_e\), all transverse dimensions are reduced by \((R_e)^{-1/2}\), while lengths increase by a factor of \(R_e\) to conserve volume. Thus the filamentary aspect ratio \(d/L\) scales as \((R_e)^{-1/2}\), while the filamentary separation scales as \((R_e)^{-1/2}\). (If the deformation is not ideal, one can define an effective \(R_e\) which would produce the actual \(d/L\) ratio.) For large deformations, it is a reasonable approximation to model the current flow as purely axial in the filaments and purely radial in the copper matrix. It then follows that \(J_f\) in the filaments is \(J_f (d/\xi)^{-1/2} (R_e)^{-1/2}\), where \(J_f\) is the overall average current density carried by the wire, while the transverse current density in the copper matrix (evaluated midway between the filaments) is \(J_s \sim J_f (d/\xi)^{-1/2} (R_e)^{-1/2} \approx 10^{-4} J_f\) for \(d/\xi \approx 0.1\) and \(R_e \approx 100\).

For composites discussed in this paper the volume fraction of niobium (or vanadium) in the initial two-phase alloys ranged from 0.075 to 0.20. Although increasing the volume fraction of the superconducting phase obviously results in higher overall critical current densities, other practical considerations (stability, tensile and bending strain tolerance, ac loss behavior) dictate keeping it below \(~25\%).

Most samples described in this paper were prepared in small quantities (a few mm in length). In one instance a 100 m long single piece of Cu-NbSn wire was prepared in order to permit the calorimetric measurement of ac losses\(^8\) and to explore the feasibility of scaling up the process to larger quantities.

**Coupling Mechanisms**

The a priori consequence of discontinuous filaments is that current must flow through a nominally normal copper matrix to get from one filament to the next. The coupling mechanisms, which can lead to a vanishingly small resistance in these composites, are not yet fully understood and they continue to stimulate further theoretical and experimental studies. The emphasis of this paper, however, is on composites with large area reduction ratio \((R_e \gtrsim 1000)\), and for this limiting case there is sufficient experimental evidence that the critical current density \(J_c\) is not limited by the interfilamentary coupling strength, but rather by the flux-pinning strength of the filamentary material itself. In the

**Fig. 1.** Scanning electron micrographs of etched cross-sections of in situ Cu-NbSn wire (a) and tape (b), below the percolation threshold.
Re = 10^3. This shows how extremely effective the long thin filamentary segments are in reducing the current density in the copper. This has two crucial implications:

First, the energy dissipation in the copper is reduced by a factor of \((J_s/J)^2\), leaving an effective remnant resistivity given by

\[
\rho_{\text{rem}} = \frac{1}{f_s} \frac{1}{\rho_s} \frac{1}{\rho} \left( \frac{d}{h} \right)^2 \text{(1)}
\]

where \(\rho\) is the resistivity of the matrix. This resistance reduction ratio would be \(\sim 10^{-8}\) for the typical values of \(f_s = 0.1\) and \(Re = 10^3\), reducing the resistance below the level of detectability except with a SQUID voltmeter. A result similar to (1) was found by Callaghan and Toth,12 apart from the numerical factor. By contrast with this physically reasonable result, conventional effective medium theory is not useful in dealing with such highly anisotropic inclusions of superconducting material14 since it predicts zero resistance for any \(f_s > d^2/\pi^2 \sim 10^{-6}\).

The second crucial consequence of the small \(J_s/J\) ratio is that \(J_s\) can easily be so small that it can be carried as a super-current by proximity effect in the copper matrix. Again taking \(f_s = 0.1\) and \(Re = 10^3\), a \(J\) of \(10^5\) amperes/cm\(^2\) translates into \(J_s \sim Jf_s^{-4}(Re)^{-3/2} \sim 10^2\) amperes/cm\(^2\), small enough to be carried by even a weak residual proximity-effect superconductivity in the copper. It is this geometrical reduction of \(J_s/J\) that causes the effective \(J_s\) to be limited by the filaments rather than the matrix coupling if \(Re\) is large enough.

Since the proximity-effect critical current density \(J_{cl}\) rises continuously from zero at \(T_c\) of the superconducting inclusions, there may be an intermediate temperature region between \(T_c\) and some lower \(T_{cl}(J)\), at which \(J_{cl}\) has become large enough to carry the \(J_s\) for a given applied \(J\), causing the resistance to fall all the way to truly zero. In this intermediate regime, the continuous \(T_c\)-dependence of the resistance is controlled by two distinct influences: the geometrically random array of coupling strengths between filamentary segments (described as a percolation problem) and the thermodynamic fluctuations which would cause a gradual onset of phase locking even in a geometrically regular array. If percolation is dominant, one would expect the remnant resistance to vanish when the effective superconducting fraction \(f_s\) reaches a critical value \(f_c\).

Making a phenomenological generalization of (1), we have

\[
\rho_{\text{rem}} = \frac{1}{f_s} \frac{1}{\rho_s} \frac{1}{\rho} \frac{d^2}{h} (1 - \frac{f_s}{f_c}) \text{(2)}
\]

\[
= 0 \quad \frac{f_s}{f_c} > 1
\]

where \(f_c\) is a critical exponent which is expected to depend on dimensionality, and \(f_s\) is assumed to decrease smoothly with increasing current at temperature, and magnetic field. Since the stiffness of the coupling against thermodynamic fluctuations would also be smoothly controlled by these same parameters, the observable behavior of \(R(T)\) would be qualitatively similar, whichever rounding mechanism is dominant.

Measurements of the remnant resistivity as a function of temperature and current2,13,14 (Fig. 2) and of applied magnetic field15 on samples containing Nb and Nb\(_3\)Sn filaments are in good qualitative agreement with the above model. An important idea here is that, by varying \(T_c\), or \(I\) one can vary \(f_s\), and therefore interfilament coupling, without introducing in the process undesired microstructural changes and secondary effects.

The crucial test of the model, however, and of the usefulness of the in situ material are the measurements of the critical currents. A logical extension of (1) is that in the long-filament limit, the composite critical-current density, \(J_c\), should scale with \(f_s\) even in composites below the microstructural percolation threshold. However, early experimental data, reported by various investigators, did not confirm this predicted behavior. At high fields in particular, a decrease of \(f_s\) by a factor of two resulted typically in a decrease of \(J_c\) by an order of magnitude.

Because composites with low \(f_s\) are particularly useful for those applications where good stability, low ac losses and high tolerance for bending strain are required, and because they provide at the same time a sensitive test of our model, we have recently reexamined the superconducting behavior of Cu-Nb,Sn composites with only 7.5 vol.\% Nb (volume fraction of Nb\(_3\)Sn filaments in the reacted composites was 0.098). Our Results suggest that some of the conclusions stemming from earlier experimental tests are not valid, partly because of the difficulties in preparing uniform Cu-Nb samples with low Nb concentration16 and partly because of the secondary or side effects introduced by varying the volume fractions of the superconducting filaments.

In the case of Cu-Nb,Sn composites, it is now known that the intrinsic properties \(T_c\) \(H_c\) of Nb\(_3\)Sn filaments depend rather sensitively on the matrix-to-filament volume ratio.17 This ratio determines the magnitude of the compressive stress exerted on the filaments by the matrix due to differential thermal contraction. Reducing \(f_s\) will therefore not only reduce \(J_c\) in a
direct material but will also lower $H_{c2}$ of the filamentary structure. A more meaningful way to compare $J_c$'s of two composites with different $f_0$ is therefore to plot normalized critical current density $J_c^* = J_c / f_0$ versus reduced critical field $h = H / H_{c2}$. This is done in Fig. 3 for two composites containing 0.23 and 0.10 volume fraction of Nb$_3$Sn filaments (original Nb$_3$Sn volume fraction before plating and annealing was 0.182 and 0.075, respectively). Furthermore, care was taken that the average matrix composition after annealing was the same in both composites. Although $J_c$ vs $H$ plot (Fig. 3a) shows almost an order of magnitude discrepancy between the two samples at high field, the experimental points fall on a single master curve when plotted as $J_c^*$ vs $h$ (Fig. 3b).

Fig. 3. a) overall and b) normalized critical-current densities for two in situ Cu-Nb$_3$Sn tapes, containing different volume fractions of superconducting filamentary, plotted versus applied and reduced field, respectively.

At lower fields, proximity-effect induced coupling becomes increasingly more important and magnetization measurements in fact show that the samples can become completely superconducting. Consequently, high critical-current densities can be achieved even in composites with low filament volume fraction. The self-field $J_c$ of the Cu-Nb$_3$Sn tape with $f_0 = 0.096$, discussed earlier, was found to be $8 \times 10^5$ A/cm$^2$ at 1 mV/cm voltage criterion. Increasing the sensitivity by two orders of magnitude to 0.01 mV/cm (corresponding to resistivity criterion of $2 \times 10^{-14}$ $\Omega$-cm) resulted in a decrease of the measured $J_c$ by only 10%. This indicates that the sample either behaves as a true superconductor or at the very least that its remnant resistivity is low enough to allow technological applications. An added bonus is exceptional conductor stability; samples of this composition could be repeatedly quenched, without a shunt protection, even in zero applied field with no apparent damage.

**Flux Pinning Mechanisms**

The primary source of flux pinning in bronze-processed A-15 materials is known to be grain boundaries. Dependence of the pinning force $F_p(h)$ on the reduced magnetic field $h$ can be described adequately by $F_p \propto h^{1/2} (1-h)^{2}$ in the range of $H > H(\eta=\infty)$. In materials with very small grain size, however, this range is further restricted to fields close to $H_{c2}$, where the condition that the flux-line lattice parameter is much smaller than the pinning-site spacing is satisfied.

(This restriction is particularly applicable to in situ composites where relatively low annealing temperatures and short annealing times result in very small grain size of the A-15 material.) At lower fields, the field dependence of the flux-line lattice parameter leads to a more complicated $F_p$ vs $h$ relationship and $F_p^{max}$ can sharply increase.

Even with this restriction, the scaling laws are not expected to hold well in those in situ composites where the filament thickness becomes comparable to the grain size. Under these conditions, one can no longer assume a statistical distribution of A-15 grains which forms a basis for the derivation of the scaling laws. Moreover, pinning of flux lines by interfaces is expected to contribute significantly to the total pinning force, and, therefore, the orientation of the ribbon like filaments with respect to the applied field becomes important.

In multifilamentary tapes, where the filaments are aligned in the rolling plane, this leads to anisotropy in critical properties and flux-flow behavior, which is particularly pronounced at high fields (Fig. 4). It should be noted that the aspect ratio of many filaments is greater than that of the tape itself due to the fact that the filaments deform in the plane strain mode already during the wire drawing process. As expected, the strongest pinning occurs when the fluxoids are parallel to the filament surfaces. The anisotropy factor $\alpha(H) \equiv (J_c^*(H)) / (J_c^*(0))$ depends strongly on the magnitude of the applied field and can reach almost an order of magnitude in Cu-Nb$_3$Sn tapes with high aspect ratio, measured at high fields ($H \approx 0.9 H_{c2}$).

Fig. 4. Field dependence of the anisotropy factor $\alpha(H)$ for a Cu-Nb$_3$Sn multifilamentary tape with the aspect ratio of 27.

Another microstructural feature which should not be ignored in composites with very thin filaments is interface roughness. Unless perfectly flat, interfaces will pin even when normal to the applied field since movement of the lattice will then require changes in the length of the vortices. This type of pinning might be important in overannealed filaments with lance-like morphology and may partly account for the observed rapid increase of low-field $J_c$'s with decreasing filament thickness.

In Cu-Nb filamentary tapes with single-crystalline Nb filaments, one would expect the relative contribution of the interface pinning to the bulk pinning force to be even more important. Our recent study of transport behavior of Cu-Nb composite tapes indeed reveals a rapid increase of $\alpha(H)$ as a function of $H$, with additional effects...
due to the anisotropy in $H_{c2}$.

Turning back to the high-field Cu-Nb$_3$Sn composites, the combined effect of small grain size, surface pinning and proximity effect results in very high critical-current densities, particularly in highly reduced in situ tapes. The self-field critical-current density normalized to the volume fraction (0.23) of the Nb$_3$Sn filaments was as high as $1.4 \times 10^6$ A cm$^{-2}$, and the maximum flux pinning force $F_p$ at 37 exceeded $7 \times 10^{10}$ N m$^{-3}$. These values are comparable to the highest values of $J_c$ and $F_p$ in Nb$_3$Sn thin films and layered composites. It should be again stressed that $J_c$ behavior in the long-filament limit is dictated by the properties of the filamentary material, including filament-matrix interfaces. While proximity effect coupling is not crucial to keep $F_p$ low, it nevertheless enhances $J_c$, particularly at low fields, presumably through $F_p$ increased.

We have recently extended the \textit{in situ} approach to preparation of Cu-V$_3$Ga composites. In contrast to earlier published work, we have demonstrated that Cu-V$_3$Ga conductors with excellent critical properties can be produced by \textit{in situ} techniques. Transition temperature (15.5 K midpoint) and upper critical field (22.4 T) in particular are among the highest reported in the literature even for bulk V$_3$Ga and indicate that stoichiometric V$_3$Ga can be grown at relatively low temperature in the presence of the copper matrix. Typical overall $J_c$ for these samples, containing only 20 vol.4 V, was $10^4$ A cm$^{-2}$ at 18 T and $10^5$ A cm$^{-2}$ at 21 T and clearly exceeded the $J_c$ of Cu-Nb$_3$Sn composites with comparable $F_p$ and diameter at fields above 10 T (Fig. 5). Qualitatively similar results have been now reported also by other investigators, using both \textit{in situ} approaches.

\begin{figure}[h]
\centering
\includegraphics[width=0.8\textwidth]{fig5.png}
\caption{$J_c$ of Cu-V$_3$Ga and Cu-Nb$_3$Sn in situ composites with comparable cross-sectional area ($\approx 4.4 \times 10^{-4}$ cm$^2$) and filament volume fraction (23-25 vol.%) (from Ref. 27).}
\end{figure}

\begin{figure}[h]
\centering
\includegraphics[width=0.8\textwidth]{fig6.png}
\caption{Ultimate tensile strength of \textit{in situ} Cu-Nb composites as a function of true strain $\phi$ (from Ref. 6).}
\end{figure}

\begin{figure}[h]
\centering
\includegraphics[width=0.8\textwidth]{fig7.png}
\caption{Ultimate tensile strength of \textit{in situ} Cu-Nb composites as a function of true strain $\phi$ (from Ref. 6).}
\end{figure}

\section*{Stress Effects and Strengthening Mechanisms}

\textbf{Cu-Nb Composites}

One of the most striking characteristics of the ultrafine filamentary composites prepared by \textit{in situ} techniques is their exceptional mechanical strength. Our initial observations of enhanced yield stress and ultimate tensile strength of Cu-Nb$_3$Sn composites, and of enhancement of $J_c$ in composites under bending strain, stimulated a more detailed study of strengthening mechanisms in the Cu-Nb system. We found that in composites with very small filaments ($d < 1000$ Å) the ultimate tensile strength increases anomalously even at very low volume fractions (10-18%) of filaments. The highest values (2.2 GPa at R.T.; 2.9 GPa at 77 K) were found to approach the estimated theoretical strength of the material and equal the strength of the best copper whiskers (Fig. 6). This remarkable behavior is linked to the presence of densely spaced interfaces which inhibit dynamic recovery in both the matrix and the filaments and result in accelerated work hardening and extremely high dislocation densities ($N_{\text{disl}}$). Since the flow stress in work-hardened materials scales with $N_{\text{disl}}^{1/2}$, it is clear that increasing $N_{\text{disl}}$ by a factor of 40-50 over the values attainable in a single-phase material should result in a dramatic increase in strength. In the smallest composites (25 μm in diameter), however, one finds an interesting coexistence of two high-strength components with diametrically opposite defect structure: Nb filaments with virtually no dislocations and therefore whisker-like behavior, and the matrix, packed with so many dislocations and other defects that the stresses, required to move dislocations from one interface to another, approach theoretical strength.

Although Cu-Nb composites cannot be used as high-field superconductors because of the relatively low $H_{c2}$ of niobium filaments, their development may nevertheless be of crucial importance in the current efforts to achieve higher magnetic fields. The pulsed normal-state magnets used for this purpose are limited in their performance by the combination of strength and conductivity.
of the material from which they are constructed. In terms of these two properties, the Cu-Nb conductors (and other similar composites currently studied at Harvard) are far superior to the best presently known conventional materials. Moreover, in contrast to other high-strength and high-conductivity materials, whose electrical conductivity is limited by the impurity scattering and, therefore, virtually independent on the temperature, the main sources of scattering in Cu-Nb composites are dislocations and interfaces. Lowering the temperature to, say, 77 K will therefore result not only in 25-30% increase in strength but also in 2-3 fold increase of electrical conductivity.

![Graph showing the low temperature (~ 10 K) resistivity of annealed Cu-Nb composites as a function of inverse wire diameter.](image)

**Fig. 7.** Low temperature (~ 10 K) resistivity of annealed Cu-Nb composites as a function of inverse wire diameter.

Other possible applications for Cu-Nb (and similar) composites, proposed in the past, include the use of these materials as simultaneous stabilizing and strengthening elements in the design of high-field superconducting magnets, and as conductors in rapidly rotating electrical machinery. Recent progress in developing high-$T_c$ Nb$_2$Ge tapes also suggests the use of Cu-Nb tapes as high-strength substrates for superconducting compounds prepared by CVD techniques. For all these applications the optimum combination of strength and conductivity can be achieved by varying filament volume fraction $V_f$, reduction ratio $R$, and annealing treatment.

Our recent study of electron scattering mechanisms in in situ composites shows that both dislocation and boundary scattering contributions to the total resistivity scale with inverse wire diameter. The functional dependence of the boundary scattering (Fig. 7), in particular, in both high- and low-temperature limits, is in good agreement with Dingle’s expressions derived for thin wires, assuming that most electrons scatter diffusively at the interfaces. The contribution of both boundary and dislocation scattering to the resistivity in the smallest composites (25 μm in diameter) can exceed 1 μΩ·cm at room temperature and hence outweigh the phonon and impurity contributions. Judging from Figs. 6 and 7 and from dislocation resistivity data in Ref. 35, the optimum trade-off between composite strength and conductivity should be achieved in composites with somewhat reduced interfacial or dislocation density. The proper choice of processing parameters will be, of course, dictated by the temperature of the application and specific performance requirements.

**Cu-Nb$_3$Sn and Cu-V$_3$Ga Composites**

The matrix dislocation density in Cu-Nb$_3$Sn and Cu-V$_3$Ga composites is, of course, much lower than in Cu-Nb due to the necessary high temperature anneal. The resistance to plastic flow and the ultimate tensile stress are nevertheless very high due to small interfilament spacing and strong filament-to-matrix bonding. Although the yield stress and the ultimate tensile strength again scale with $D^{1/2}$, as in Cu-Nb composites, the underlying physical mechanism is rather different. The composite will fail mechanically whenever the stress concentration at the tip of dislocation pile-ups in the matrix exceeds the shear strength of the filaments. Since the stress necessary to generate a sufficient number of dislocations in a pile-up will increase with decreasing interfilament spacing, the overall strength of the composites is expected to increase with reduction ratio. The experimental results are in fair agreement with this prediction, considering the oversimplified treatment of a rather complex problem.

![Graph showing $H_c^2$ dependence on the applied uniaxial stress for Cu-Nb$_3$Sn and Cu-V$_3$Ga in situ composites with comparable diameter (~0.25mm) and filament volume fraction (~0.23 - 0.25).](image)

**Fig. 8.** $H_c^2$ dependence on the applied uniaxial stress for Cu-Nb$_3$Sn and Cu-V$_3$Ga in situ composites with comparable diameter (~0.25mm) and filament volume fraction (~0.23 - 0.25).

The most striking difference between in situ and conventional composites, however, is not in their strength but rather in the stress/strain tolerance of their critical properties. Our experiments show that both Cu-Nb$_3$Sn and Cu-V$_3$Ga composites can be subjected to stresses close to their respective $\sigma_{UTS}$ without any permanent degradation in $J_c$ or $H_c^2$. In fact, in Cu-Nb$_3$Sn composites, whose $H_c^2$ depends sensitively on the applied stress (Fig. 6), one can substantially reduce the negative effect of the compressive stress exerted on the filaments by the matrix by simply pre-stressing composites to ~60-70% of their ultimate tensile strength. The net effect of this procedure is a permanent increase of $J_c(H)$ to the peak values of $J_c(0)$ curves (Fig. 9, also Fig. 5). In contrast, $H_c^2$ of Cu-V$_3$Ga composites depends only weakly on the applied stress (Fig. 9) and, consequently, stress effects on $J_c$ are much smaller but, again, fully reversible. Qualitatively similar results have been reported also by other investigators.

**Mechanical Properties of Twisted In Situ Composites**

As discussed in Part II of this paper, twisting can substantially reduce ac losses in in situ filamentary
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