PRACTICAL SUPERCONDUCTOR DEVELOPMENT
FOR ELECTRICAL POWER APPLICATIONS

ARGONNE NATIONAL LABORATORY

QUARTERLY REPORT FOR THE PERIOD ENDING DECEMBER 31, 1997

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This is a multiyear experimental research program focused on improving relevant material properties of high- T_c superconductors and on development of fabrication methods that can be transferred to industry for production of commercial conductors. The development of teaming relationships through agreements with industrial partners is a key element of this program.

Technical Highlights

Recent work on microstructural development and current distribution in Bi-2223 powder-in-tube tapes and a novel application for high-temperature superconductors are discussed.

Phase and Microstructural Development in Ag/Bi-2223 Multifilament Conductors

A major focus of our collaboration with American Superconductor Corporation (ASC) has been on optimization of phase chemistry and microstructure in Ag/Bi-2223 composite conductors. ASC's development work in this area has moved along two related paths. One path has involved optimization of the oxide powder (OP) approach, wherein the precursor is a blend of metal oxides; the second path utilizes a metal powder (MP) precursor. The methodologies for fabricating Ag/Bi-2223 conductors from OP and MP precursors are similar. The principal differences between the OP and MP routes are related to the fact that (1) MP composites are subjected to most of their mechanical deformation (by drawing and rolling) while the precursor powder is still in the metallic state and (2) after drawing and rolling, MP composites require a special thermal treatment to convert the metallic precursor into oxide phases prior to initiation of the series of heat treatments that transform the precursor powder into Bi-2223.

We have addressed several key aspects of phase and microstructure development during heat treatment of multifilament Ag/Bi-2223 tapes. The composition of the precursor powders used to manufacture the OP tapes included in this report was $\text{Bi}_{1.7}\text{Pb}_{0.3}\text{Sr}_{1.9}\text{Ca}_{2.0}\text{Cu}_{3.0}\text{O}_y$. The MP precursors were slightly richer in Cu. Heat treatments were initiated by immersing 20-mm-long tapes into a preheated furnace that contained, in most cases, 7.5% O_2 in N_2 . After heat treatment for 10-6000 min, the tapes were quenched in an oil bath. We used the immersion-heating procedure ($\approx 80^\circ\text{C}/\text{min}$) to minimize the time spent below the desired heat treatment temperature. Similarly, we used a rapid quench at the

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end of the heat treatment to inhibit phase transformations that occur during the usual types of post heat-treatment furnace cooling that are employed with Ag/Bi-2223 conductors. These two special procedures (immersion heating and quenching) have allowed us to examine phase formation and kinetics along the isothermal heat treatment plateau without the complicating effects of normal ramp-up and cool-down protocols used in the manufacture of Ag/Bi-2223.

In this report, we present results for 361-filament MP wires and for 361-, 85-, and 19-filament OP wires to provide an MP-versus-OP comparison at the 361-filament level and to examine the effects of filament count and filament architecture on OP composites. In the course of this work, we explored the effect of oxygen partial pressure, pO_2 (0-0.13 atm), heat treatment temperature (790-845°C), and time (0-6000 min) on the kinetics of Bi-2223 formation and conductor microstructure.

Our accumulated experience with the effect of pO_2 on the phase transformation behavior of Ag-sheathed Bi-2223 precursors continues to suggest that a pO_2 of ≈ 0.08 atm is optimal. To check this for the multifilament composites examined in the present study, we obtained phase-conversion data at 815°C for 361- and 85-OP wires at pO_2 between 0 and 0.13 atm, with a processing time of 6000 min. The results of this study (Fig. 1) indicate that a pO_2 of 0.075 atm (the value used in much of our prior work) will support high conversions of the precursor to Bi-2223 for both the 361-OP and 85-OP composites; but it should be noted that the optimal range is decidedly broader for the composite with the higher filament count. Results included in Fig. 1 for the same two composites at a lower temperature (790°C) illustrate how the Bi-2223 content (even after 6000 min) drops off sharply at temperatures slightly below 800°C. These results support the choice of 0.075 atm O_2 as the base case pO_2 for subsequent studies.

Further information about the effect of temperature on the Bi-2223 formation rate for MP and OP composites is presented in Fig. 2 where we show the fractional conversion of Bi-2212 to Bi-2223 for selected specimens processed for 1000 min in 0.075 atm O_2 . This time interval was carefully selected so that no specimens reached 100% conversion before the quench, which allowed us to map the relative reaction rates at all test temperatures. The results in Fig. 2 for the 85-, 19-, and monofilament OP composites were obtained with specimens that exhibited the same powder stoichiometry (base case) and phase chemistry. Results are also shown for two 361-filament MP composites with an unknown pre-heat-treatment phase chemistry because of the as yet unresolved consequences of the post rolling oxidation step used to convert the MP precursor to a predominantly oxide form prior to heat treatment. Putting this uncertainty aside, it is clear that the 85- and 19-filament OPs exhibited a

Fig. 1. Fractional conversion of Ag/Bi-2223 precursor to Bi-2223 in 85- and 361-filament OP composites at 815 and 790°C for various values of pO_2 .

Fig. 2. Fractional conversion of precursor to Bi-2223 in selected MP and OP Ag/Bi-2223 composites after 1000 min in 0.075 atm of O_2 . Number after slash is the filament count.

broader lower-temperature-shifted reaction profile than the monofilament OP and that the 361-filament MPs exhibited an even broader profile and appreciably enhanced kinetics.

Measurements of the length of the Ag/ceramic powder interface and the cross-sectional area of typical filaments (obtained from scanning electron microscopy [SEM] images), which essentially define the surface-area-to-volume ratio of these filaments, provide a correlation for the results shown in Fig. 2. The interface-to-cross-section ratio of the OP monofilament is 40 mm^{-1} , whereas the ratios of the 85- and 19-filament OPs are 112 and 102 mm^{-1} , respectively. The ratio for the two 361-filament MPs is difficult to measure (for reasons that will become apparent later), but the value of this ratio is $\approx 300 \text{ mm}^{-1}$. We believe that increased proximity of the precursor powder to the Ag sheath is the principal reason why the peak reaction rate increases, the overall profile broadens, and the peak of the profile seems to shift to a lower temperature as the filament count increases. This result is consistent with our earlier monofilament work, which showed that, in the initial stages of heat treatment, Bi-2223 formation proceeds more rapidly at the Ag/ceramic powder interface than in the bulk of the specimen. In fact, the earlier results implied that the reaction was complete at the interface before it even started in the mid-core region of a given filament.

Another series of measurements was conducted to explore the fractional conversion of Bi-2212 to Bi-2223 as a function of time at the base case ($p\text{O}_2 = 0.075 \text{ atm}$) and several selected temperatures in peak regions of the reaction profiles shown in Fig. 2. In Fig. 3, we show conversion-versus-time results obtained at 810°C from one of the 361-filament MP composites presented in Fig. 2. The results in Fig. 3, and similar results from other MP and OP composites were analyzed by the model developed independently by Avrami and Erofe'ev. In this much-simplified presentation, the fractional conversion to Bi-2223, F , is equated to time by the relationship $\ln(1-F) = kt^n$, where k is a rate constant, t is time, and n is the exponent of time. In our work, we express F as being equal to $B/(aA + B)$, where A is the X-ray diffraction (XRD) peak intensity of the (008) reflection of Bi-2212, B is the intensity of the (0010) reflection of Bi-2223, and a is a coefficient to correct for differences in XRD scattering factors and other related effects. A value of 0.44 for a has been generally recommended, but to account for the fact that the expression for F in terms of A and B ignores the fugacity of secondary phases in the reacting ceramic powder, a can be taken as 0.88. In the past, we have treated our results as if a were equal to 1.0, and have found that the key conclusions to be drawn from analysis of results like those in Fig. 3 are not appreciably affected by this choice for the value of a .

Fig. 3. Fractional conversion of precursor to Bi-2223 in 361-filament MP composite heat treated at 810°C for various times in 0.075 atm of O₂. Results of the Avrami-Erofe'ev analysis are also summarized.

The results of the Avrami-Erofe'ev modeling are presented in Table 1. Whereas the value of k tends to vary widely, depending on wire type (MP or OP), filament count, and temperature, the value of n at temperatures near the peak of the reaction profile (820-830°C) is consistently in the range of 1.5 ± 0.3 , regardless of whether we use $a = 1.0$ or $a = 0.88$. Values of n in this range imply general conformance to several mechanisms, including one that is fully consistent with many observations in the published literature about phase evolution and microstructure development in Ag/Bi-2223 composite conductors, namely, diffusion-controlled, two-dimensional (platelike) growth under conditions of steady to decreasing nucleation rate. At temperatures near the onset points of the reaction profiles in Fig 2, the value of n often drops below 1.0 (see, e.g., Table 1), possibly indicating a different Bi-2223 formation kinetics mechanism at temperatures below the temperature region of the most rapid kinetics.

Table 1. Summary of Fitting Parameters from Avrami-Erofe'ev^a Analyses of OP and MP Kinetic Data

Temperature (°C)	Filament Type/Count	Fitting Parameters			
		<i>a</i>	<i>n</i>	<i>k</i>	1-Sigma
810	MP/361	1.00	1.19	7.6E-05	0.026
	MP/361	0.88	1.16	1.0E-04	0.028
	OP/85	1.00	0.94	8.4E-04	0.040
	OP/85	0.88	0.93	9.9E-04	0.042
825	OP/19	1.00	1.60	1.4E-05	0.050
830	MP/361	1.00	1.71	1.2E-05	0.086
830	MP/361	0.88	1.69	1.4E-05	0.086
830	OP/85	1.00	1.76	1.3E-06	0.075
830	OP/85	0.88	1.67	2.6E-06	0.083

^aFit to $\ln(1-F) = -kt^n$ for $F = [\text{Bi-2223}] / (a [\text{Bi-2212}] + [\text{Bi-2223}])$.

The principal microstructural differences between the monofilament and multifilament conductors examined in this study are related to the overall cross-sectional area of individual superconducting filaments and the ratio between the volume of each filament and its surface area. Both of these properties tend to decrease with increasing filament count for composites with the same transverse dimensions. When the cross-sectional area of the filaments is reduced, the largest nonsuperconducting secondary phases (usually copper oxide or alkaline earth cuprates that can be as large as 10 μm in major dimension) can begin to occupy a significant fraction of the cross-sectional current path and possibly cause major current path blockages. When the Ag/powder interface area is increased relative to the powder volume (a general consequence of increasing the filament count), the proximity of superconducting material to Ag is increased, as discussed above.

Figure 4 shows SEM images of 19-filament OP composite specimens after immersion heating to 800°C in 0.075 atm O₂ and instantaneously quenching (Image a in Fig. 4), and after immersion heating to 825°C in 0.075 atm O₂ then heat treating for 300 min (Image b in Fig. 4) and 4000 min (Image c in Fig. 4). With data like those in Fig. 3 for this particular 19-filament conductor as a guide, the instantaneously quenched sample should contain no Bi-2223, the 300-min specimen should contain $\approx 10\%$ Bi-2223, and the 4000-min sample should contain 90+% Bi-2223. These results show that second-phase crystallites (the dark particles) are present in the immersion-heated OP specimens during the initial stages of the Bi-2223 formation

Fig. 4.
SEM images (transverse view)
of 85-filament Ag/Bi-2223 OP
composite specimens processed
in 0.075 atm of O₂ for (a) 0 min
at 800°C, (b) 300 min at 825°C,
and (c) 4000 min at 825°C prior
to quench.

reaction. They also reveal how grain growth progresses as the Bi-2223 formation reaction moves toward completion. The size and relative abundance of the secondary phases and their progressive dissipation as the reaction to form Bi-2223 proceeds are also noteworthy.

The broader processing window for multifilament composites, indicated by the reaction profiles in Fig. 2, raises questions about the influence of heat treatment temperature at temperatures where the Bi-2223 formation kinetics are sufficiently rapid to allow complete conversion to Bi-2223 in ≈ 100 h. In Fig. 5, we show results for two 85-filament specimens processed in 0.075 atm of O_2 for 4000 min at 830 or

Fig. 5. SEM images (transverse view) of 19-filament Ag/Bi-2223 OP composite specimens processed at 830 and 810°C in 0.075 atm of O_2 for 4000 min.

810°C. The percent conversion of Bi-2212 to Bi-2223 was $\approx 90\%$ at 830°C and $\approx 85\%$ at 810°C, but the microstructures differed considerably. The specimen heat treated at 830°C contained a better-developed Bi-2223 grain structure, but also contained domains of appreciable second-phase growth (dark regions) and exhibited numerous platelike protrusions of Bi-2223 grains into the Ag sheath. The Bi-2223 grain structure in the specimen heat treated at 810°C was not nearly as well developed, but the second phases were smaller and there were no platelike growths into the Ag sheath. It was subsequently determined that a heat treatment temperature of 820°C produced the best critical-current properties for fully processed versions of this particular composite.

The microstructure of Ag/Bi-2223 multifilaments produced by the MP processing route was unusual. SEM images of a 361-filament MP composite at three stages of heat treatment are shown in Fig. 6. In all three transverse views, it is difficult to distinguish one filament from another, in that all of the filaments seem to run together. There are several possible causes for this filament-blurring effect, e.g.: (1) virtually all of the mechanical deformation on an MP wire occurs prior to heat treatment, and blurring is perhaps a consequence of the close match of the mechanical properties of the Ag sheath and the mostly metallic precursor powder; (2) formation of liquid metal phases that partially dissolve the Ag sheath during the special oxidation step that is used to convert the metallic precursor powder to an oxide prior to what we refer to herein as heat treatment. Specific influences of each of these factors on the microstructure observed in Fig. 6 need more detailed study.

The 361-filament MP composites exhibited the broadest reaction profile of any sample examined in this study, as can be seen in Fig. 2. Within the broad temperature window shown in Fig. 2, we found two noteworthy effects, which are illustrated by the SEM results in Fig. 7. Comparing XRD data for the two specimens processed at 820 and 840°C for 1500 min (in 0.075 atm of O₂), we find that the percent conversions to Bi-2223 are essentially the same (76% and 79%, respectively). As can be seen in Fig. 7, the microstructures are also much the same in terms of the size, content, and spatial distribution of second phase. However, a comparison of the SEM results for the two specimens processed at 820°C for 1500 and 4000 min indicates that second-phase growth at the longer processing time is significantly greater. The determination that the specimen processed for 4000 min at 820°C is $>80\%$ converted to Bi-2223 (relative only to Bi-2212), together with the appearance of large regions of second-phase growth in the corresponding SEM image, suggests that there may be an optimal heat treatment time for MP composites beyond which Bi-2223 phase content begins to drop off due to decomposition.

(UR→)/361-MP/AS-RECEIVED/0% Bi-2223

(LR↘)/361-MP/810°C/300 MIN/
0.075 ATM O₂/33% Bi-2223

(LL↓)/361-MP/830°C/1500 MIN/
0.075 ATM O₂/100% Bi-2223

Fig. 6. SEM images (transverse view) of 361-filament MP composite specimens processed as indicated in upper left inset (UR = upper right, LR = lower right, LL = lower left).

(UR→)/361-MP/820°C/1500 MIN/

0.075 ATM O₂/76% Bi-2223

(LR↘)/361-MP/840°C/1500 MIN/

0.075 ATM O₂/79% Bi-2223

(LL↓)/361-MP/820°C/4000 MIN/

0.075 ATM O₂/88% Bi-2223

Fig. 7: SEM images (transverse view) of 361-filament MP composite specimens processed as indicated in upper left inset (UR = upper right, LR = lower right, LL = lower left).

Energy-dispersive X-ray analyses performed in conjunction with the SEM measurements revealed that the second-phase material seen in Figs. 6 and 7 is mostly CuO, with some $(\text{Ca,Sr})_2\text{CuO}_3$ (2/1 alkaline earth cuprate). In particular, most of the dark phases near the Ag/ceramic interfaces and in the bulk Ag are CuO. The microstructures seen in Figs. 6 and 7 (before and after heat treatment) were investigated in greater detail by transmission electron microscopy (TEM). The image in Fig. 8 shows a longitudinal view of a typical filament in a 361-MP specimen after the MP oxidation step but prior to heat treatment. This view reveals the severe sausaging that was commonly observed for the MP composites prepared for this study.

Finally, some information about texture development in high-filament-count Ag/Bi-2223 conductors was gained from XRD measurements on specimens of 361-OP and 361-MP composites that were heat treated under conditions that favored nearly complete formation of Bi-2223. XRD can be used to probe the extent of texturing of the Bi-2223 phase in Ag/Bi-2223 composites, and, in the case of the 361-OP and 361-MP specimens prepared for this particular investigation, we found that the 361-MP composite exhibited considerably better texture than the 361-OP composite after heat treatment at 820°C in 0.075 atm of O₂ for 4000 min. This result is illustrated in Fig. 9, which shows XRD patterns for the 361-OP and the 361-MP

Fig. 8. TEM image (longitudinal view) of as-received 361-filament MP composite, illustrating a commonly observed region of sausaging.

OXIDE POWDER (OP) VS METAL PRECURSOR (MP) MULTIFILAMENTS

Fig. 9. X-ray diffraction patterns for a 361-filament OP composite and a 361-filament MP composite, heat treated for 4000 min at 820°C in 0.075 atm of O₂.

specimens. The higher count levels and stronger (00l) texturing of the 361 MP are clearly evident. The ceramic filament cores in heat-treated 361-MP specimen were found to be less porous in both transverse and longitudinal SEM views than any of the OP multifilaments we investigated. This observation can be appreciated by

comparing the images in Figs. 4 and 5 with the ones in Figs. 6 and 7. The superior texturing of 361-MP composites appears to be a beneficial consequence of their compact microstructure.

Magneto-optical Imaging (MOI) of $(\text{BiPb})_2\text{Sr}_2\text{Ca}_2\text{Cu}_3\text{O}_x$ Composite Conductors

Significant bend strains occur for powder-in-tube tapes during magnet fabrication. The heat-treated superconductor is wrapped into a coil, which is the most economical fabrication technique. For the react-and-wind method, superconductor core damage is not fully repaired by further heat treatment. Previous studies have demonstrated that strain tolerance of Ag-clad BSCCO tapes exceeds that of monolithic bulk bars. Based on these observations, we developed a current-shunting model to account for mechanical-property differences among composite superconductors. The model was verified by monitoring discrete voltage distributions in Ag/BSCCO composite structures and by magneto-optical imaging (MOI).

Using a high-resolution MOI technique, we studied the flow of transport currents around cracks in monofilament $(\text{BiPb})_2\text{Sr}_2\text{Ca}_2\text{Cu}_2\text{O}_x$ composite conductors. Shunting through the Ag clad was observed directly for the first time. The current in the Ag decays exponentially with distance from the crack, as expected for a one-dimensional distributed-resistance model. At 77 K, the decay length l is ≈ 0.4 mm, implying an interface resistivity of $4 \times 10^{-8} \text{ Wcm}^2$. This decay length increased with increasing temperature.

The ceramic nature of the superconducting cores makes them very susceptible to strain-induced cracks. In bend strain tests, it has been observed that strains of 0.2–0.4% in monofilament conductors and 0.6–1.5% in multifilament conductors cause irreversible reductions of the critical current. In addition, otherwise unperturbed conductors can exhibit residual cracking from the mechanical deformation steps in the powder-in-tube process. Cracking and current shunting through the Ag sheath have been shown to play an important role in determining the transport properties of the composite conductors. It has been predicted that a characteristic healing length l governs the current flow around the cracks in Bi-2223 composite conductors.

Here, we present a high-resolution MOI study of the transport current flow in cracked monofilamentary Bi-2223 composites. This technique yields maps of the component B_z of the local magnetic induction that is normal to the imaged sample surface. The response of the sample to an applied field and to a transport current was studied. In the applied-field mode the strength of persistent supercurrents is detected, while in the transport current mode the technologically important flow patterns of an externally imposed transport current and current sharing with the Ag sheath are seen.

A monofilament Bi-2223 conductor with an isolated crack and a critical current I_c of 35 A at 77 K was studied. Current pulses of 21 A and 0.2-s width were synchronized with the shutter of a high-resolution CCD camera. To enhance the optical resolution, the top face was polished flat without breaking into the superconducting core, leaving $\approx 30 \mu\text{m}$ of Ag (Fig. 10).

In fields below 35 G, the applied field was completely shielded, indicating that the crack was partial. An applied field of 107 G (Fig. 10b) penetrated preferentially along the crack, a behavior expected for any weak link where the critical current is reduced. The image shows that the crack meanders; moreover, near the bottom

Fig. 10.

(a) Geometry of experimental monofilament conductor. (b) MOI at 30 K due to applied field of 107 G. Brightness and darkness represents high and low field, respectively. Width of frame corresponds to length of 3 mm.

edge of the conductor, additional fractures produced more complicated field patterns. Because the crack near the top edge was almost straight and uniform, we chose this area for the study of the flow of transport current.

Figure 11 shows field maps of a transport current in zero applied field at 30, 90, and 110 K. At 30 K, 21 A was below critical and all of the current flowed in the superconducting core. At 110 K, all of the current flowed in the Ag, causing the contrast at the edge (upper dashed line in Fig. 11a). At 90 K, the current flowed partially in the Ag. The intensity of contrast and its spatial extent provide a measure of the degree of current sharing and of the characteristic healing length. The two bands of dark contrast (marked by diagonal dashed lines in Fig. 11b) on either side of the crack indicate sharp concave turns in the supercurrent path around the crack and into the Ag, as schematically indicated by the arrows. The strong positive fields at the location of the crack were caused by the convex curvature of this current path.

Quantitative profiles of the field due to a current of 21 A are shown in Fig. 12. The data at 30 and 110 K show the strong minima of the local field at the edges of the superconductor and the Ag. The trace at 90 K along the crack, corresponding to Fig. 11b, clearly shows the superposition of both features and indicates current sharing. Also shown is a field profile at 90 K, due to a current of 21 A flowing in an external field of 107 G. This profile is taken at a distance of 1.5 mm from the crack. At this distance, the effects due to the crack have largely decayed. The minimum of B_z at the Ag edge indicates that in a field of 107 G, a current of 21 A is above critical for the bulk and that current sharing occurs along the entire conductor length. Figure 12 also includes simulations of these data, based on the numerical integration of Biot-Savart's law for the present conductor geometry, assuming a uniform current density in the Ag and in the superconductor. The experimental data are well defined, with a current sharing of 5 A flowing in the Ag and 16 A flowing in the superconducting core. We note, however, that due to the complicated current path, this fitting procedure is not applicable in the cracked region.

An alternate way of determining the total current I in the Ag clad that does not rely on fitting the entire field profile is based on the change in slope DS of $B_z(y)$ on either side of the minimum at the Ag edge (see inset of Fig. 12) according to: $1 = (A k \Delta S) / \mu_0$. Here, $\mu_0 = 4\pi \cdot 10^{-7}$ Vs/Am, $A = 7.6 \times 10^{-3}$ cm² is the Ag cross-sectional area, and k is a geometrical factor. Inasmuch as the Ag edge is far enough away from the superconducting core, the magnetic fields due to supercurrents contribute only to a smooth background. Therefore, this method is not affected by details of the supercurrent path. At the surface of a uniform conductor with a rectangular cross section, carrying a uniform current, $k = 2$. As indicated in Fig. 10a, the Ag clad of our sample forms a hollow conductor with rounded end faces. Furthermore, because of residual unevenness of the sample surface, images are

Fig. 11. MOI of fields due to current of 21 A at (a) 30, (b) 90, and (c) 110 K. Horizontal dashed lines mark edges of superconducting core and Ag. Gray arrows in frame b indicate current path around crack. Brightness and darkness represent high and low fields, respectively. Width of frames corresponds to length of 3 mm.

always taken at a finite distance above the sample, estimated to be as large as 20 μm . Therefore, the value of k that corresponds to this geometry is calibrated experimentally by analyzing the field profiles due to various known currents at 110 K, giving $k = 4$. It is noted that, for 90 K and 107 G applied field, this method yields the same current as does the fitting to Biot-Savart's law.

Profiles of $B_z(y)$, such as shown in Fig. 12, are generated at various positions x along the conductor length and yield the dependence of I on the distance from the crack as shown in Fig. 13 for various temperatures and fields. The current in the Ag exhibits a sharp peak at the location of the crack ($x = 0$) and decays rapidly on either side. This behavior is expected in a one-dimensional resistor model for the composite, which predicts that $I(x) = (I_0 - I_{cr}) \exp(-|x|/l)$. Here, I_0 is the total transport current, and I_{cr} is the temperature- and field-dependent critical current across the crack.

$$l = \sqrt{(r_i A) / (r_{Ag} L)},$$

where l is the healing length given in terms of the Ag cross section A , the circumference of the superconducting core L , the resistivity of the Ag r_{Ag} , and the interface resistivity r_i between Ag and the superconductor. Fits of the data to this expression are shown as solid lines in Fig. 13. At 77 K and 0 G, ≈ 4 A flow in the Ag at the location of the crack. This value increases to 8 A at 90 K because of the decrease in I_{cr} with increasing temperature. The healing length at 77 K is 0.4 mm,

Fig. 12. Profiles of normal component B_z of the magnetic field due to a current of 21 A. Traces along the crack at 30, 90, and 110 K are shown. Also included are a trace at 90 K in a field of 107 G, taken at a distance of 1.5 mm from the crack and a fit to the data, based on numerical integration of Biot-Savart's law. The inset illustrates determination of change in slope of $B_z(y)$ at the Ag edge.

which increases to 0.8 mm at 90 K. Using the expression for l with $r_{Ag} = 0.29 \mu\text{Wcm}$ at 77 K, we estimate an interface resistivity of $r_i = 4 \times 10^{-8} \text{ Wcm}^2$ at 77 K. This value for a composite conductor is in good agreement with studies on sinter-forged Ag/Bi-2223 sandwiches.

In conclusion, a high-resolution MOI technique for transport currents was used to determine the current path in cracked Bi-2223 composite conductors. The images directly show the extent of current sharing between the Ag clad and the superconducting core in the vicinity of the crack. The current in the Ag is found to decay exponentially with distance from the crack. At 77 K, the healing length is ≈ 0.4 mm and increases to 0.8 mm at 90 K. These findings are in good agreement with the predictions of a one-dimensional distributed-resistance model.

Fig. 13. Dependence of current shunted into Ag clad on distance from the crack at various temperatures and fields. Solid lines are fits according to the one-dimensional resistor model.

HTS Flux Pump SBIR Project with Ability Engineering Technology

The preliminary design of the HTS flux-pump concept demonstration apparatus is complete. A schematic representation of the pump is shown in Fig. 14. The demonstration flux pump will charge a high-temperature superconductor (HTS) load coil to 210 A with a 25-A input current. HTS tape conductor for the flux pump coils and the load coil is being procured from Intermagnetics General Corporation (IGC), with delivery scheduled for January 1998. Coil-winding procedures, components, and equipment are being developed and will be evaluated during practice winding.

Interactions

Vic Maroni attended a Wire Development Group meeting at NIST headquarters in Gaithersburg, MD, on Oct. 9-10, 1997. He presented a report of progress during the previous four months. ASC, ORNL, LANL, and U. of Wisconsin representatives also attended.

Balu Balachandran visited IGC on October 13, 1997 to discuss the BSCCO and YBCO conductor development activities.

Fig. 14. HTS flux pump concept demonstration circuit.

On October 17, 1997, George Crabtree organized a scientific meeting at Argonne on grain boundaries. It included the groups of Larbalestier (U. of Wisconsin), Salama (TCSUH), Suenage/Welch (BNL), and ANL. In addition to information exchange, several activities were proposed and are under active consideration: a special edition of the MRS bulletin on HTS grain boundaries; and a conference/workshop on HTS grain boundaries. Dr. Salama also presented a seminar on melt-processed materials.

Three engineers from S&C Electric, Chicago, visited Argonne on October 21, 1997 to discuss possible collaboration in the area of HTS for electric power applications.

On Nov. 6, Dr. Michael Staines, The New Zealand Institute for Industrial Research and Development, visited ANL and discussed their activities in the area of powder-in-tube and bulk fabrication of HTS.

The Industrial Overview Committee meeting for the three superconductivity technology centers (ANL, LANL, ORNL) was held at Argonne during Nov. 17-18, 1997.

Two senior engineers from Picker International, Inc., Cleveland, OH, visited ANL to discuss the status of HTS technology.

Balu Balachandran attended the MRS Fall Meeting in Boston, Dec. 1-4, 1997.

Steve Dorris presented a paper at the U.S./Japan Workshop on High- T_c Superconductors in Tallahassee, FL on Dec. 8-10, 1997.

List of Publications:

Published:

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