APPLIED PHYSICS LETTERS VOLUME 83, NUMBER 20 17 NOVEMBER 2003

Reversible axial-strain effect and extended strain limits in Y-Ba-Cu-O coatings on deformation-textured substrates

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(Received 16 July 2003; accepted 25 September 2003)

The dependence of transport critical-current density J_c on axial tensile strain ε was measured at 76 K and self-magnetic field for $YBa_2Cu_3O_{7-\delta}$ (YBCO) coatings on buffered, deformation-textured substrates of pure Ni, Ni–5-at. %-W, and Ni–10-at. %-Cr–2-at. %-W. Expectations have been that the strain tolerance of these composites would be limited by the relatively low yield strains of the deformation-textured substrates, typically less than 0.2%. However, results show that the irreversible degradation of $J_c(\varepsilon)$ occurs at a strain equal to about twice the yield strain of the substrate. Therefore, YBCO/Ni-alloy composites may satisfy axial-strain performance requirements for electric devices, including the most demanding applications, motors and generators in which a strain tolerance exceeding 0.25% is needed. Furthermore, the YBCO/Ni–5-at. %-W conductors showed a reversible strain effect, which may be induced by a reversible strain-field broadening around mismatch dislocations at the grain boundaries. This effect may contribute to the unexpectedly large usable strain range of these conductors.
[DOI: 10.1063/1.1628818]

The ongoing development of high-temperature superconducting wires is stimulated by the prospect of their wide use in electric utility devices and industrial magnets.[1,2] Of significance, transport critical-current densities J_c over 1 MA/cm^2 were recently achieved in $YBa_2Cu_3O_{7-\delta}$ (YBCO) coated conductors in lengths of 10 m.[3] However, a challenging requirement for their application is that such composites must experience little or no degradation of J_c from the longitudinal, transverse, and bending strains to which they are subjected during strand fabrication and handling, magnet winding and operation, and power-line cabling and installation.[4]

A previous study of J_c versus transverse compressive stress, carried out in YBCO coatings deposited on buffered, rolling-assisted, biaxially textured substrates (RABiTS),[5] showed the benefit of using Ni–5-at. %-W (Ni-5W) substrates instead of weaker pure Ni in these composites.[6] It is also known that YBCO/Ni-alloy withstands higher axial *stress* than YBCO/pure-Ni before ceramic breakage.[7] However, for RABiTS, no information is available on the axial *strain* at which fractures occur, nor on the relation of this critical strain to the substrate stress–strain curve.

In this letter, we present the effect of axial tensile strain ε, applied along the conductor axis, on J_c in YBCO films deposited on buffered Ni-5W, Ni–10-at. %-Cr–2-at. %-W (Ni-10Cr-2W), and pure Ni RABiTS. Expectations have been that the strain tolerance of these composites would be severely limited by the relatively low yield strains (proportional limit of elasticity ε_p) of the RABiTS. However, our results reveal that the onset strain $\varepsilon_{\mathrm{onset}}$ where $J_c(\varepsilon)$ first drops below its initial value is substantially higher than ε_p, indicating that fractures in the ceramic layers do not occur simultaneously with substrate yielding for these systems. Furthermore, $\varepsilon_{\mathrm{onset}}$ for YBCO/Ni-alloy is about twice that for YBCO/pure-Ni. This allows the coated conductor-alloy to satisfy many of the strain-limit benchmarks for device applications. For YBCO/Ni-5W, we also found a reversible strain effect that extends significantly beyond $\varepsilon_{\mathrm{onset}}$.

YBCO films were grown on buffered Ni-5W, Ni-10Cr-2W, and pure-Ni RABiTS by the use of a BaF_2 *ex situ* process with either metalorganic deposition precursors (samples A1–A3, B1–B3)[3] or evaporated precursors (sample C1).[8] Thicknesses of the substrate (t_S), YBCO (t_{YBCO}), and Ag (t_{Ag}) layers are listed in Table I. About 10 μm of Ag was added to the original (1 to 3 μm) Ag layer in order to achieve low-resistivity electrical contacts.[9] This procedure was followed for all samples, except that for A3 and B3 the additional Ag was deposited only at the sample ends, where current contacts were attached. This was done to investigate whether t_{Ag} has any effect on strain properties of the composite. Samples had widths of 3 to 4 mm and a length of 3.5 cm.

To measure J_c versus ε, the ends of the sample were soldered to two copper blocks in the test apparatus, which blocks served the dual purpose of electrical contacts and grips for applying strain. One grip was stationary and the other could slide onto a carriageway so that the sample could contract freely during cooling. Strain was measured directly at the sample location using a calibrated extensometer attached to the two grips. All strain data are given in terms of *applied* strain rather than a calculated *intrinsic* strain that takes into account prestress from differential thermal con-

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TABLE I. Electromechanical properties of coated conductors on Ni-5W, Ni-10Cr-2W, and pure-Ni RABiTS. Measurements were performed at 76 K. J_c and I_c values were determined at a 1 μV/cm criterion.

Sample No.	Substrate; t_S[a] (μm)	t_{YBCO}[a] (μm)	t_{Ag}[a] (μm)	Initial J_c (MA/cm^2)	Initial I_c (A/cm-width)	Initial n-value[b]	ε_{onset}[c] (%)	$\varepsilon_{1\%}$[d] (%)	$\varepsilon_{5\%}$[d] (%)	ε_{irr}[e] (%)
A1	Ni-5W; 75	1	14	1.03	103.6	39	0.28	0.35	0.54	0.38
A2	”	1	14	0.98	97.8	31	0.32	0.43	0.62	–
A3	”	1	3	0.87	86.6	30	0.28	0.35	0.54	0.38
B1	Ni-10Cr-2W; 75	1	14	1.14	114.1	39	0.35	0.43	0.62	0.35
B2	”	1	14	1.16	116.2	42	0.26	0.32	0.50	0.26
B3	”	1	3	1.17	117.1	41	0.33	0.41	0.54	0.33
C1	Pure Ni; 50	0.3	11	0.91	29.1	21	0.17	0.22	0.47	0.17

[a]t_S, t_{YBCO}, t_{Ag}: thicknesses of substrate, YBCO, and Ag components, respectively.
[b]n-value: exponent in the fit $V \propto I^n$ of the $V-I$ curves around the critical current I_c.
[c]ε_{onset}: strain at the beginning of the degradation of $I_c(\varepsilon)$.
[d]$\varepsilon_{1\%}$, $\varepsilon_{5\%}$: strain corresponding to 1% and 5% reduction of $I_c(\varepsilon)$ from its initial value, respectively.
[e]ε_{irr}: strain up to which any degradation of $I_c(\varepsilon)$ is reversible.

traction among the components of the composite, which prestress is not known.

Uncertainty in the measurement of strain was about $\pm 0.02\%$. Measurements were made in self-field, in liquid nitrogen at 76 K. J_c was determined for an electric-field criterion of 1 μV/cm. Uncertainties in determining the critical current I_c and the YBCO cross section were, respectively, about 1% and 10%. For stress–strain curves, coated and bare substrate samples (1 cm wide and 30 cm long) were measured with an apparatus designed for thin tapes. Uncertainties in estimating the Young's modulus and yield strength were about 10%.

Samples were of high quality, with an initial J_c of about 1 MA/cm^2 and n-value (exponent in the fit $V \propto I^n$ of the $I-V$ curves around I_c) between 20 to 40 (Table I). Figure 1 shows $J_c(\varepsilon)$ for the two YBCO/Ni-alloy composites. These systems had very similar behavior, with an average ε_{onset} of about 0.3%. We also used 1% and 5% criteria for degradation of $J_c(\varepsilon)$ to define strains $\varepsilon_{1\%}$ and $\varepsilon_{5\%}$, respectively. $\varepsilon_{5\%}$ had an average value of 0.56%. Both samples A3 and B3 ($t_{Ag} = 3$ μm) had similar J_c dependence on ε as compared to samples A1-2 and B1-2 ($t_{Ag} = 14$ μm) up to a strain corresponding to about 10% reduction of $J_c(\varepsilon)$, but exhibited a more precipitous degradation of J_c as ε was increased further. This suggests that a thick Ag layer may provide relatively low-resistance current paths around dissipative strain-induced defects in the YBCO layer, but does not affect the strain characterizing the start of $J_c(\varepsilon)$ degradation.

Figure 2 compares $J_c(\varepsilon)$ for typical samples of YBCO/Ni-alloy and YBCO/pure-Ni composites. It is evident that the latter system is less tolerant to strain compared to the former since ε_{onset} for YBCO/pure-Ni, 0.17%, is only about half that for YBCO/Ni-alloy. A bending strain study showed that a thick YBCO layer is more prone to cracking.[8] Therefore, as t_{YBCO} for YBCO/pure-Ni is one third that for YBCO/Ni-alloy composites, the difference depicted in Fig. 2 is not attributable to the difference in t_{YBCO}, but is probably due to the weaker mechanical properties of pure Ni compared to the Ni-alloy substrates.

Scanning electron microscopic examination of samples after 1.2% to 1.5% strain showed cracks in the YBCO layer transverse to the applied-strain direction. Cracks of lengths from 1 to 15 μm and widths around 50 nm are distributed randomly throughout the sample area. The crack density is higher in YBCO/pure-Ni than in YBCO/Ni-alloy.

The mechanical properties of coated conductor composite and bare substrates are summarized in Table II. Comparison of ε_p of a substrate and ε_{onset} of the corresponding coated conductor shows that the onset of $J_c(\varepsilon)$ degradation is greater than the onset of substrate yielding. This contrasts with, but does not contradict, data obtained for YBCO films coated on buffered Inconel-625 substrates using ion-beam-assisted deposition, in which the two strains were about the same ($\sim 0.35\%$, similar to our results for YBCO on Ni-alloy RABiTS).[10] We suggest that there exists a critical strain ε_c at which YBCO fractures that is coincidentally close to the greater value of ε_p of Inconel-625. For substrates in which $\varepsilon_p < \varepsilon_c$, substrate yielding does not induce YBCO cracking so long as local strain remains smaller than ε_c throughout the entire YBCO/buffers/substrate interfaces. For $\varepsilon_p \ll \varepsilon < \varepsilon_c$, inhomogeneous substrate yielding could drive local strain

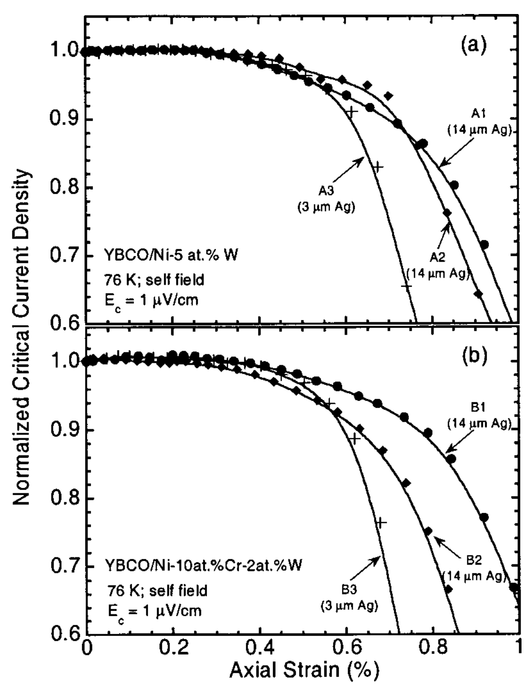

FIG. 1. $J_c(\varepsilon)$ at 76 K and self-field for YBCO films on (a) Ni-5W and (b) Ni-10Cr-2W RABiTS.

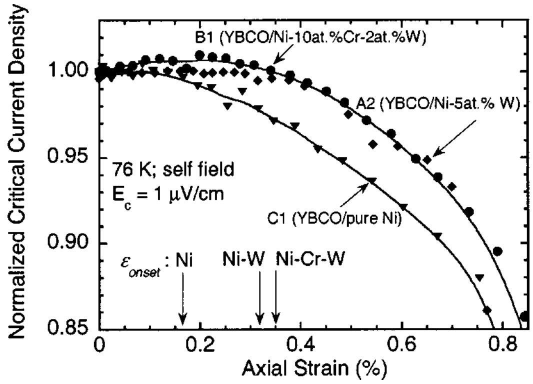

FIG. 2. Comparison of $J_c(\varepsilon)$ at 76 K and self-field for typical YBCO films on pure Ni, Ni-5W, and Ni-10Cr-2W RABiTS. Strain at the onset of $J_c(\varepsilon)$ degradation for YBCO/Ni-alloy is about twice that for YBCO/pure Ni.

higher than ε_c, thus initiating crack formation. This unexpected decoupling of ε_c from ε_p is of particular importance for YBCO on RABiTS since the deformation-textured substrates have relatively low values of ε_p ($\leqslant 0.2\%$). Due to this decoupling, YBCO/Ni-alloy composites may be used in electric devices, including the most demanding of applications: rotating machinery, both motors and generators, which require strain tolerance in excess of 0.25%.[4]

Finally, we investigated whether any *elastic* strain effect on J_c could be found at 76 K and self-field. To determine the irreversible strain ($\varepsilon_{\mathrm{irr}}$) beyond which the conductor is permanently damaged, all samples (except A2) were periodically unloaded and J_c remeasured. Results obtained for sample A1 (YBCO/Ni-5W) are presented in Fig. 3. Data when the sample was loaded and unloaded are labeled by corresponding unprimed and primed letters, respectively. For example, from strain point *A*, the tensile load applied to the sample was released to nearly zero. The corresponding unloaded strain was point A' (the sample did not return to its original length due to the conductor yielding). Remarkably, J_c fully recovered upon unloading. This reversibility continued up to a strain as high as 0.38%. Beyond this strain, J_c degraded irreversibly. Note that even beyond $\varepsilon_{\mathrm{irr}}$, there is a partial recovery of J_c. This reversible regime was also observed for J_c determined at a 0.1 μV/cm criterion. Moreover,

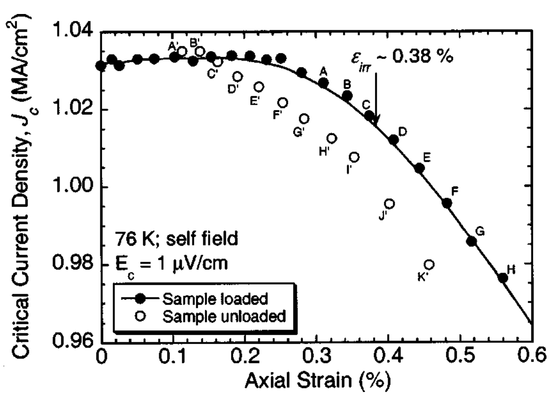

FIG. 3. J_c shows an *elastic* strain effect at 76 K and self-field that extends well beyond the onset of J_c degradation. Data when the sample was loaded and unloaded are labeled by unprimed and primed letters, respectively.

TABLE II. Mechanical properties of coated conductor and RABiTS materials at 76 K.

	ε_p (%)[a]	σ_Y (MPa)[b]	E (GPa)[c]
Ag/YBCO/Ni-5W coated conductor	0.16	208	132
Ni-5W substrate	0.20	254	128
Ni-10Cr-2W substrate	0.17	210	120
Pure-Ni substrate	0.08	59	72

[a] ε_p = proportional limit of elasticity.
[b] σ_Y = tensile yield strength; 0.02% offset criterion.
[c] E = effective Young's modulus (initial slope of stress vs strain).

the same reversible effect was found in sample A3. This suggests that for this composite, fractures of YBCO did not occur until $\varepsilon_{\mathrm{irr}}$, which we identify with the critical strain ε_c. The full recovery of J_c upon releasing strain was not observed for either YBCO/Ni-10Cr-2W or YBCO/pure-Ni composites. The origin of this elastic behavior is still unclear at present. It is known that small grain misalignment nucleates periodically spaced dislocations at grain boundaries, separated by "good" current-conductive channels.[1,11] Strain associated with misfit dislocations can be large and degrades the superconducting properties of the material around dislocations.[11] We suggest that for $\varepsilon \leqslant \varepsilon_{\mathrm{irr}}$, applied strain reversibly widens the strain field around each dislocation core such that the conductive channels are reversibly narrowed. This channel narrowing may occur due to a degradation of the superconducting critical temperature T_c of the material affected by the strain field.[12] Besides its phenomenological aspect, this reversible effect contributes to the significant extension of the usable strain range of these conductors.

This work was supported by the U.S. Dept. of Energy/Office of Electric Transmission and Distribution, and the U.S. Dept. of Energy/High Energy Physics.

[1] D. Larbalestier, A. Gurevich, D. M. Feldmann, and A. Polyanskii, Nature (London) **414**, 368 (2001).
[2] A. P. Malozemoff, D. T. Verebelyi, S. Fleshler, D. Aized, and D. Yu, Physica C **386**, 424 (2003).
[3] D. T. Verebelyi, U. Schoop, C. Thieme, X. Li, W. Zhang, T. Kodenkandath, A. P. Malozemoff, N. Nguyen, E. Siegal, D. Buczek, J. Lynch, J. Scudiere, M. Rupich, A. Goyal, E. D. Specht, P. Martin, and M. Paranthaman, Supercond. Sci. Technol. **16**, 19 (2003).
[4] J. W. Ekin and N. Cheggour, Presentation at the *DOE Wire Development Workshop*, St. Petersburg, Florida, January 2003.
[5] A. Goyal, D. P. Norton, J. D. Budai, M. Paranthaman, E. D. Specht, D. M. Kroeger, D. K. Christen, Q. He, B. Saffian, F. A. List, D. F. Lee, P. M. Martin, C. E. Klabunde, E. Hartfield, and V. K. Sikka, Appl. Phys. Lett. **69**, 1795 (1996).
[6] N. Cheggour, J. W. Ekin, C. C. Clickner, D. T. Verebelyi, C. L. H. Thieme, R. Feenstra, A. Goyal, and M. Paranthaman, IEEE Trans. Appl. Supercond. **13**, 3530 (2003).
[7] J. Yoo and D. Youm, Supercond. Sci. Technol. **14**, 109 (2001).
[8] C. Park, D. P. Norton, J. D. Budai, D. K. Christen, D. Verebelyi, R. Feenstra, D. F. Lee, A. Goyal, D. M. Kroeger, and M. Paranthaman, Appl. Phys. Lett. **73**, 1904 (1998).
[9] J. W. Ekin, *Superconductor Contacts*, in *Handbook of Superconducting Materials* (Institute of Physics, Bristol, 2002), p. 857.
[10] C. L. H. Thieme, S. Fleshler, D. M. Buczek, M. Jowett, L. G. Fritzemeier, P. N. Arendt, S. R. Foltyn, J. Y. Coulter, and J. O. Willis, IEEE Trans. Appl. Supercond. **9**, 1494 (1999).
[11] M. F. Chisholm and S. J. Pennycook, Nature (London) **351**, 47 (1991).
[12] U. Welp, M. Grimsditch, S. Fleshler, W. Nessler, J. Downey, G. W. Crabtree, and J. Guimpel, Phys. Rev. Lett. **69**, 2130 (1992).