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Engineering the precompression of Bi,Pb(2223): the influence of the geometry of the metallic matrix on the mechanical properties of tapes

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***Abstract*—At relatively low tensile strain (ε) levels, Bi,Pb(2223) tapes show a strain insensitive plateau in their $I_c(\varepsilon)$ behavior. The "Irreversible I_c reduction model" explains this regime as due to the irreversible damage caused by the precompression of the ceramic filaments by the metal matrix. In this paper we investigate whether this damage occurs predominantly during the cool-down from the reaction temperature to room temperature, or during the cryogenic refrigeration. Furthermore, we present $I_c(\varepsilon)$ data on a range of tapes with different superconductor-to-matrix ratio and different tape geometry, showing how these factors influence the precompression. A modification of the "Irreversible I_c Reduction Model " is proposed to take into account "connected-grains" domains in the $I_c(\varepsilon)$ behavior.**

***Index Terms*—Bi,Pb(2223) tapes, mechanical properties, tensile strain, irreversible strain, filling factor.**

I. Introduction

Mechanical properties of Bi,Pb(2223) tapes in longitudinal tensile experiments have been widely studied in view to obtain suitable properties for practical applications [1]-[6]. The measurement of $I_c(\varepsilon)$ shows a constant value up to an irreversible strain limit ε_{irr} beyond which I_c falls abruptly. Whereas the appearance of the steep decrease of I_c is commonly attributed to cracks formation and propagation, the microscopic origin of the plateau is rarely discussed.

According to the "irreversible I_c reduction model" presented by ten Haken et al. [1], [2], this plateau is due to the precompression applied by the matrix on the BSCCO core during cooling down from annealing temperature to measurement temperature. This compressive strain causes irreversible damages in the BSCCO core, altering I_c. Further lowering of the critical current occurs when the precompression is completely released. We investigate this model with measurements of tapes made with different filling factors, each of them have been optimized to get higher j_c. Numerical calculations were performed to estimate the precompression of BSCCO all over the cooling process. Ultrasonic fracture experiments give a qualitative estimate of the stresses endured by the ceramic filaments at room temperature and at 77K.

Comparison between the measured ε_{irr} versus filling factor and calculated values suggests a modified model to describe $I_c(\varepsilon)$.

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II. Experimental

Mono- and multifilamentary pure Ag-sheathed tapes were produced by the powder in tube technique (PIT). Coprecipitated precursors made of oxides and carbonates have been calcined in air. Heat treatments were performed in air at 838°C for about 200h. A 10% reduction pressing step was applied after 35 hours. Properties of these tapes are listed in table I.

TABLE I
TAPES PROPERTIES

Sample	Number of filaments	Filling Factor [%]	J_c [kA/cm^2]
Tape A	1	30	20
Tape B	37	21	16
Tape C	19	13	33
Tape D	19	13	21

For the tensile experiment the samples were indium-soldered on a brass support (Fig. 1), which was elongated by a DC motor applying a controlled force. Keeping the force constant, the samples were cooled to 77K. The strain level ε was then stepped up progressively, monitored by an extensometer with four strain gauges clamped on the support. I_c was measured with the standard four-point technique.

The ultrasonic fracture technique has been described by Anderson et al. [7], single filaments were first extracted from the tape by etching in a mixture of NH_4OH and H_2O_2, fracturing was then performed in an acetone bath.

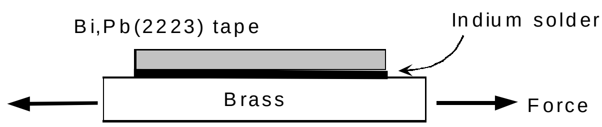

Fig. 1. Schematic I_c vs. strain measurement

III. Results

A. *Estimate of the strains involved during cooling*

We performed numerical calculations to estimate the precompression endured by the ceramic core during the cooling down. The mechanical properties of the Ag matrix and the ceramic filaments over the whole temperature range from heat treatment to measuring temperature are the basic parameters to be introduced in the model. The choice of Young's modulus $E^{Ag}(T)$ and $E^{fil}(T)$, yield strains $\varepsilon_y^{Ag}(T)$ and $\varepsilon_y^{fil}(T)$, thermal expansion coefficients $\alpha^{Ag}(T)$ and $\alpha^{fil}(T)$ have been taken from litterature:

1. *Young's modulus*: $E^{Ag}(T)$ is given by Bell [8], and following this author we assumed a linear dependence of Young's modulus versus temperature. For BSCCO we adjusted the linear parameters to fit data by Goretta et al. [9].
2. *Yield strain*: A linear relation between yield stress σ_y^{Ag} and temperature is suggested by Bell [10], the parameters were adjusted to fit results by Harvey et al. [11] at room temperature and to $\sigma_y^{Ag} = 0$ at melting temperature. The yield strain is obtained using Hook's law. The behavior of the compressed multiphase ceramic core is assumed to stay in an elastic regime
3. *Thermal expansion*: Low temperature behavior of $\alpha^{Ag}(T)$ and $\alpha^{fil}(T)$ are deduced from Okaji et al. [12] and adjusted to fit 19 E-6 K^{-1} for Ag and data by Goretta et al. [9] for BSCCO at room temperature. At higher temperatures we assumed a linear dependence up to 22 E-6 K^{-1} for Ag and 16,5 E-6 K^{-1} for BSCCO at 1110 K.

At high temperatures differential thermal contraction induces an *elastic* deformation of both the matrix and the ceramic core. In this regime the strains for a temperature decrease of ΔT are calculated using "action = reaction" law between the two components (filaments and matrix) of a tape. Using Hook's law we deduce:

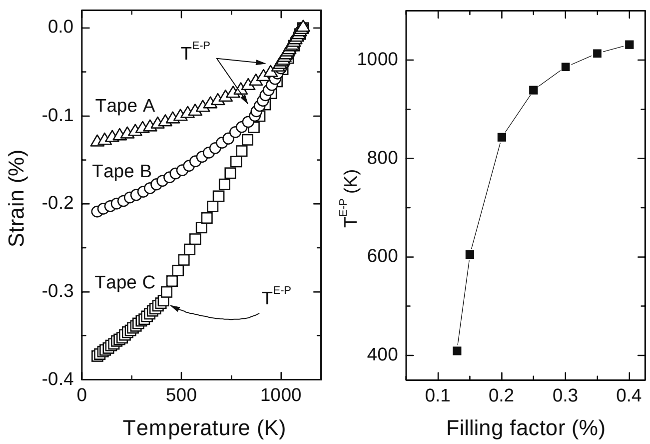

Fig. 2. Calculated evolution of strain in BSCCO during cooling and "plastic to elastic" transition temperature (T^{E-P}) of silver sheath versus filling factor.

$$\Delta\varepsilon^{fil}(T) = [1 + \frac{F \cdot E^{fil}(T)}{(1-F)E^{Ag}(T)}]^{-1}(\alpha^{Ag} - \alpha^{fil})\Delta T \qquad (1)$$

$$\Delta\varepsilon^{Ag}(T) = [1 + \frac{(1-F)E^{Ag}(T)}{F \cdot E^{fil}(T)}]^{-1}(\alpha^{fil} - \alpha^{Ag})\Delta T \qquad (2)$$

where F is the filling factor (%BSCCO / (%BSCCO + %Ag). The values of $\varepsilon^{fil}(T)$ and $\varepsilon^{Ag}(T)$ are given by the integration of (1) and (2):

$$\varepsilon^{fil}(T) = \int_{\substack{Annealing\\Temperature}}^{T} \frac{\Delta\varepsilon^{fil}}{\Delta T} dT \ (3), \quad \varepsilon^{Ag}(T) = \int_{\substack{Annealing\\Temperature}}^{T} \frac{\Delta\varepsilon^{Ag}}{\Delta T} dT \ (4)$$

When $\varepsilon^{Ag}(T)$ reaches the yield strain of silver $\varepsilon_y^{Ag}(T)$ we switch to a *plastic* regime. Below this transition temperature T^{E-P} (Fig. 2) the silver matrix deforms plastically exerting a constant stress σ_y^{Ag} on the BSCCO. The strain is given by:

$$\varepsilon^{fil}(T) = \varepsilon_y^{Ag}(T)\frac{1-F}{F}\frac{E^{Ag}(T)}{E^{fil}(T)} \qquad (5)$$

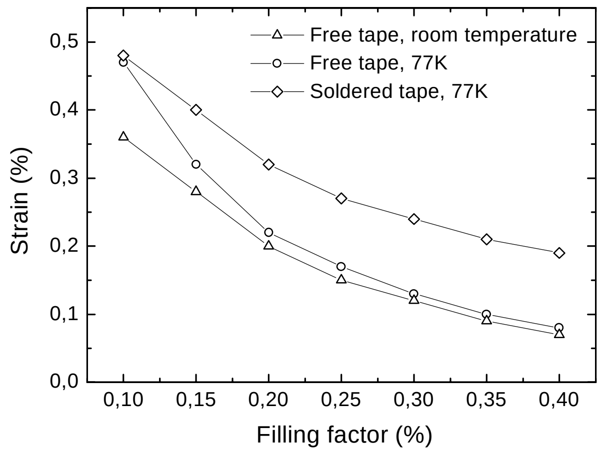

Fig. 3. Calculated strains of BSCCO core for tapes cooled down to room temperature and 77K. BSCCO precompression in tapes soldered on a brass support after cooling to room temperature is calculated assuming the support completely controls the expansion from 295K to 77K.

We also considered the case of tapes soldered on a brass support during cooling down from room temperature to 77K (Fig. 3). In this case the contraction is determined by the support and BSCCO strain is

$$\varepsilon^{fil}(T) = \varepsilon^{fil}(T = 295K) + \int_{295}^{T}(\alpha^{brass} - \alpha^{fil})dT \qquad (6)$$

B. *I_c versus strain*

As shown in Fig. 4 and 5 the irreversible strain depends strongly on the filling factor. Samples C and D are from the same green tape and have been treated at slightly different temperatures. Nevertheless they show comparable irreversible strain limits, regardless of the absolute critical current density.

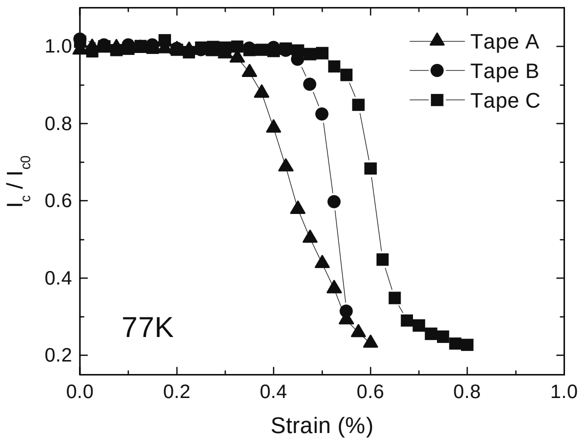

Fig. 4. Normalized critical current I_c/I_{c0} versus strain for tapes A, B and C measured at 77K.

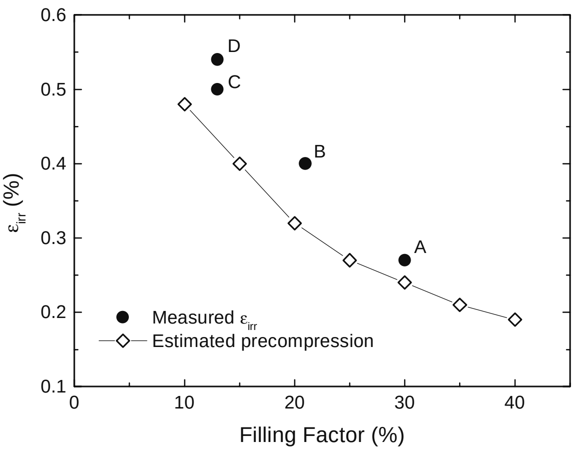

Fig. 5. Irreversible strain limit versus filling factors for optimized tapes measured at 77K compared with estimated precompression (see Fig. 2).

C. Ultrasonic fracture

The tape with the lowest filling factor (tape C) was chosen to study the fracture behavior of filaments, the influence of the matrix being more pronounced. Details of cooling processes and the mean size of the fractured fragments flat surface are listed in table II. The distribution of fragment size is shown in Fig. 6.

TABLE II
ULTRASONIC FRACTURE OF TAPE C

Cooling procedure	Mean area of fragments [μm^2]
Furnace cooled to room temperature	63'790
Furnace cooled to room temperature + quenched in liquid nitrogen	30'680
Furnace cooled to room temperature + slow cooled to 77K (7 hours) + slow heated to room temperature (7 hours)	17'320

Further cooling from room temperature to 77K clearly increases the fracture susceptibility of filaments, especially when cooling is performed slowly.

IV. DISCUSSION

The influence of the filling factor on the irreversible strain limit is verified with optimized tapes. Samples C and D, from the same green tape, show a very close behavior despite their very different J_c values, emphasizing the role of the filling factor.

According to the "Irreversible I_c Reduction Model" [1], [2] the plot of I_c versus ε_{irrev} of optimized tapes should exhibit a decreasing tendency. Our measurements do not follow this

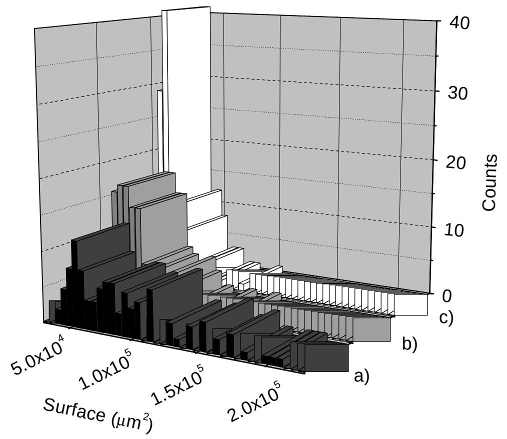

Fig. 6. Normalized distribution of fragment size after fracture of filaments from tape C in ultrasonic bath. Sample a) is furnace cooled to room temperature, b) is furnace cooled to room temperature and quenched in liquid nitrogen, c) is furnace cooled to room temperature, slow cooled to 77K (7 hours) and slow heated to room temperature (7 hours).

behavior, but tend to be randomly distributed (Fig. 7). This observation taken together with the correspondence between precompression ($\varepsilon_{precomp}$, calculated) and irreversible strain limit (ε_{irr}, measured) shown in (Fig. 5) suggests the following modification of the "Irreversible Ic Reduction Model" (Fig.8):

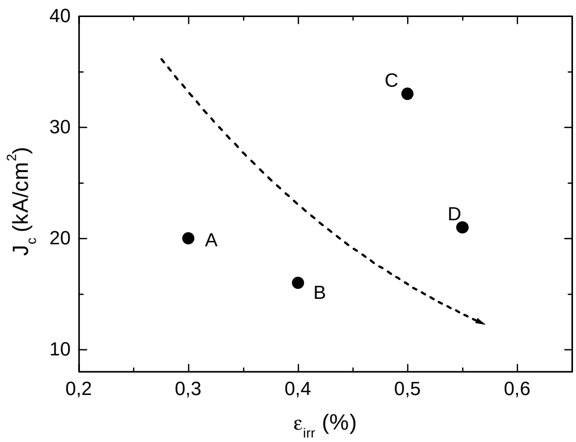

Fig. 7. Distribution of J_c versus irreversible strain of tapes A to D. The dotted arrow indicates the expected behavior following the "Irreversible I_c Reduction Model" and should only be regarded qualitatively.

During cooling the ceramic core is first submitted to a "connected-grains" deformation (i.e. a strain which does not damage grain interconnects) down to a compressive limit ε^*_{comp}, a further compression causes an irreversible lowering of I_c. For a tensile strain to cause irreversible damage, it needs to compensate at least the precompression strain $\varepsilon_{precomp}$. The systematic displacement between the calculated $\varepsilon_{precomp}$ values and the ε_{irr} data furthermore suggests the existence of a small "connected-grains" tensile strain domain ε^*_{elong}. The measured ε_{irr} is then equal to $\varepsilon_{precomp}$ + ε^*_{elong}. Comparison between estimated precompressions and measured irreversible strains gives an estimate of the tensile "connected-grains" limit of BSCCO core:

$$\varepsilon^*_{elong} = \varepsilon_{irr} - \varepsilon_{precomp} \cong 0.1\ \%$$

"Connected-grains" domains in this model are not related to elastic regimes of single BSCCO grains, but to a macroscopic behavior of the multiphase filaments including tilts and bending of composite, so far they do not introduce any damages to connections between grains.

The main consequence of this modified model is the insensitivity of I_c versus small precompression values (up to ε^*_{comp}). Whereas ε_{irr} always reflects the precompression of BSCCO, I_c should be randomly distributed since ε^*_{comp} is not reached. In this framework the distribution of I_c in Fig. 7 is consistent. Previous measurements of free tapes and soldered tapes do not reveal any significant differences in I_c (77K), also suggesting the occurrence of a "connected-grains" compressive domain. Further experiments with reinforced tapes reducing the precompression of filaments are underway. As expected first results with adequate reinforcements show a smaller ε_{irr} whereas I_c is not increased.

Ultrasonic fractures performed on tape C show a modification of fracture behavior after the sample has been cooled down to liquid nitrogen. The sample cooled slowly to 77K exhibit a stronger effect. We attribute this feature to more homogenous stresses experienced all over the tape length. Since this experiment was carried out with pieces from the same tape, comparisons of above fragment size distributions can be made (same filament thickness and width, same heat treatments, same deformation steps).

As expected for the sample cooled at lower temperature (77K), higher stresses are built up causing smaller fragments.

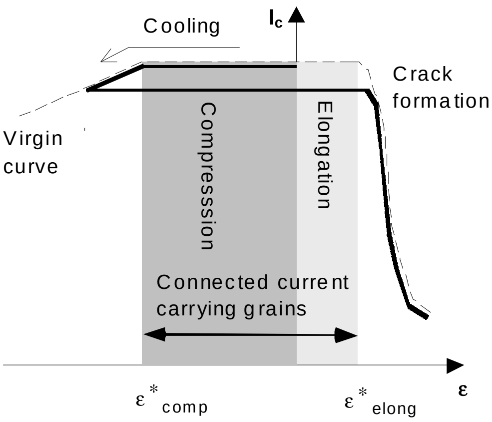

Fig. 8. Model describing the effect of cooling on the critical current and core strains of Bi,Pb(2223) tapes.

V. Conclusion

We measured the influence of filling factor on the mechanical properties of optimized tapes with low temperature tensile strain experiments. The evolution of stresses in the filament between room temperature and 77K has been studied on ultrasonic fractures of single filaments.

The measurements are compared with numerical calculations and a new $I_c(\varepsilon)$ model is presented, taking into account "connected-grains" domains in the macroscopic behavior of BSCCO filaments.

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