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Critical current in silver clad Y-Ba-Cu-O superconducting wires

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Abstract. Silver clad wires of high T_c superconductor $Y_1Ba_2Cu_3O_{7-x}$ have been fabricated through the powder metallurgy technique. The reacted wires show a midpoint T_c of 84 K. A critical current density of 26·4 A cm^{-2} (77 K, 0 T) is obtained in these wires. The wires, however, turn complete normal only at a current density of 280 A cm^{-2}. The reasons for low critical current density obtained in these wires are discussed.

Keywords. Silver clad wire; critical current density.

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The discovery of superconductivity in Y-Ba-Cu-O system around 92 K by Wu *et al* (1987) has opened up a new area of research and raised the expectation that the new superconducting devices can now be operated at liquid nitrogen temperature. Therefore, efforts to develop thermally stabilized high T_c superconducting wires are going on an unprecedented scale. So far the superconducting wires have been used most widely in the production of magnetic fields. The stringent requirements of a superconducting material useful for magnet application are that they should carry large currents in high magnetic fields and should be able to be produced in a flexible form such as a wire or a tape. Because of the tremendous commercial potential of these wires we took up a programme of developing these high T_c stabilized wires indigenously. In this brief communication, we report our preliminary results on silver clad $Y_1Ba_2Cu_3O_{7-x}$ wires.

So far, limited results have been published on these high T_c superconducting wires. Cava *et al* (1987) reported a critical current density (J_c) value of $1{\cdot}1 \times 10^3$ A cm^{-2} (77 K, 0 T) in bulk Y-Ba-Cu-O material. High values of $J_c > 10^5$ A cm^{-2} (77 K, 0 T) have been reported by Chaudhari *et al* (1987) in epitaxial film of $Y_1Ba_2Cu_3O_{7-x}$ on $SrTiO_3$ substrate. Similar high values of $J_c = 3 \times 10^6$ A cm^{-2} (4·5 K, 0 T) were inferred from the magnetization experiments by Dinger *et al* (1987) on single crystals of the compound $Y_1Ba_2Cu_3O_{7-x}$ when the applied magnetic field was perpendicular to the Cu-O plane (induced screening current being along the Cu-O plane). These experiments do indicate good potentiality of these materials to carry large J_c if grown in ideal conditions. The J_c results obtained on polycrystalline wires are, however, not so encouraging.

Yamada *et al* (1987) at Toshiba Japan prepared 2 mm diameter wire of the compound $Y_{0{\cdot}4}Ba_{0{\cdot}6}Cu_{1{\cdot}0}O_{3-x}$ in a copper sheeth through powder metallurgy and cold working. The copper sleeve was removed before the wire was finally sintered. The bare wire carries a current density of $7{\cdot}25 \times 10^2$ A cm^{-2} (77 K, 0 T). Almost

identical attempts have been made by Malik *et al* (1987). They extruded short lengths of 0·3–2.4 mm diameter copper cladded wires of the $Y_1Ba_2Cu_3O_y$ compound. Copper was etched away before the core material was oxygen reacted. A maximum value of $J_c = 90\,A\,cm^{-2}$ (77 K, 0 T) has been reported for specimen reacted for prolonged period. The wires are short and brittle. Extremely low J_c values ($1·9\,A\,cm^{-2}$ at 4·2 K, 0 T) have been reported by Matsuzaki *et al* (1987) on liquid quenched $Y_1Ba_2Cu_3$ alloy foil oxidized in air, even though the T_c is 91 K. More recently Jin *et al* (1987) succeeded in making metal clad $Y_1Ba_2Cu_3O_7$ wires with a zero field J_c of $\simeq 175\,A\,cm^{-2}$ at 77 K. We also tried a few metals for cladding but most of them were found reacting with the compound. Silver was found to be the most acceptable material. In fact we carried out systematic studies on the addition of silver to the $Y_1Ba_2Cu_3O_{7-x}$ compound and found that large addition of silver does not destroy superconductivity in this material (Sharma *et al* 1987).

Appropriate quantities of the Y_2O_3, $BaCo_3$ and CuO as per the formula $Y_1Ba_2Cu_3O_x$ were thoroughly mixed, ground and calcinated at 800° C a few times. The calcinated mass was finally ground and sintered in flowing oxygen at 850° C overnight and cooled slowly to 200° C. The sintered powder was checked for its diamagnetic behaviour at 77 K (as repelled strongly by the magnet) and the resistive transition. The sintered material which took the shape of a irregular rod and had a density of 3·43 g/cc showed a superconducting transition (midpoint) at 92 K. This sintered rod was again crushed to fine powder with an average particle size of 3·8 μm. This O_2-reacted powder was designated as Y-26 and will be referred to in the text as such. Two geometries were used for making wires. In the first geometry a copper tube (O D 8 mm and I D 6 mm) was lined with a silver sheet 0·5 mm thick, packed with the Y-26 powder and sealed at the two ends. The composite billet was cold worked to a wire of cross section of $1·2\,mm^2$. This wire is designated as Y-26 Ag(1). In the second geometry a tube (O D 12·5 mm, I D 8 mm) cast out of a commercial silver was packed with the same powder and rolled down to $1·4 \times 1·4\,mm^2$ cross section wire, designated as Y-26 Ag(2). The copper sleeve of the Y-26 Ag(1) wire was removed leaving the silver cladding intact. Both the wires were sintered in flowing oxygen at temperature between 800 and 850° C for varying periods.

The transition temperature was determined by the usual four-probe resistive method. A criterion of 1 μv/cm was employed for determining the critical current I_c. The critical current density, J_c, is then calculated by dividing I_c by the cross sectional area of the powder core as determined by the optical microscope. SEM picture were obtained on the cross section of the wires as also on the pellet made of the same compound to find out the grain size and the intergrain separation. Energy dispersive spectrum analysis was also carried out to determine the elemental percentage in the grains as well as on the grain boundaries. The X-ray diffraction pattern obtained on the $Y_1Ba_2Cu_3O_{7-x}$ powder used in the wire conformed to a distorted orthorhombic phase as reported by Cava *et al* (1987) and Rao *et al* (1987).

From the rolled wires of a few metres, specimens about 10 cm long were cut for sintering. Specimens from the first wire Y-26 Ag(1) (hereafter referred to as Ag(1)) were reacted at 825° C for 15 h in flowing oxygen. The wires showed a sharp superconducting transition with a midpoint T_c of 84 K (figure 1).

These Ag(1) wires were not pursued further for critical current studies sometimes part of the silver was etched away alongwith the copper sleeves leaving the material bare. Only Y-26 Ag(2) wires (referred to as Ag(2) hereafter) which had only silver cladding were pursued for critical current studies.

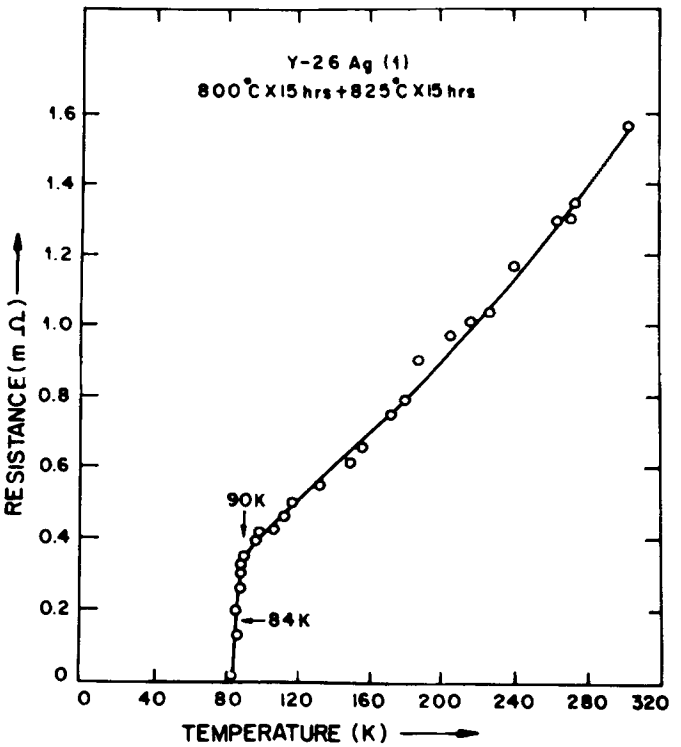

Figure 1. The resistance vs temperature plot of a silver clad $Y_1Ba_2Cu_3O_{7-x}$ wire Y-26 Ag(1). The reaction temperature and duration is given in the figure.

The Ag(2) wire reacted at 850° C for 15 h and followed by another reaction at 830° C for 38 h under flowing oxygen shows a J_c value of 26·4 A cm^{-2} (inset in figure 2). The resistance of the wire, however, shows a current dependance and saturates only at a current density of 280 A cm^{-2}. This indicates the presence of a superconducting phase with this J_c value.

As seen from the R-I plot for the Ag(2) wire in figure 2 R rises sharply at a lower current density just above the J_c value. The SEM picture taken on a $Y_1Ba_2Cu_3O_{7-x}$ pellet shows a randon orientation of the elongated grains. Individual grains are typically 15 to 25 μm long, 4 to 5 μm wide and the intergrain separation varies between 0 and 2 μm as seen from figure (3). It is known from various studies (for example Cava *et al* 1987) that the coherence length $\xi(T)$ in these superconducting compounds is only 22 A° which is much smaller than the limit of the intergrain separation $\simeq 2\,\mu$m observed in our samples. It appears, therefore, that the sharp rise in R just above J_c value is caused by the breakage of a large number of superconducting weak links. This is followed by a slow increase of resistance with current reaching a saturation—current independent value—at a higher current density. This slow increase of R should have been caused by the non-uniformity of the J_c of the grains. It has been shown by Dinger *et al* (1987) that $Y_1Ba_2Cu_3O_{7-x}$ crystals possess large anisotropy of the critical current with respect to the magnetic field direction. The ratio of the critical current along the Cu-O plane to that in the perpendicular direction is as much as 20 at 4·5 K. The anisotropy increases still further with the increase in

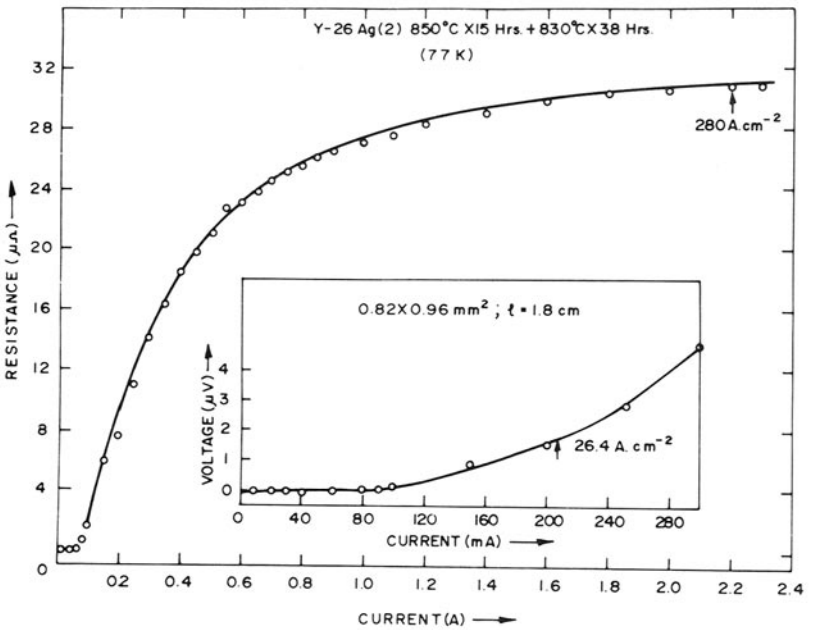

Figure 2. The I-V plot of a Y-26 Ag(2) wire reacted for a longer period shows a $J_c = 26{\cdot}4$ A cm^{-2} (77 K) as per 1 μv/cm voltage criterion (inset). The wire resistance becomes current independent at a current density of 280 A cm^{-2}.

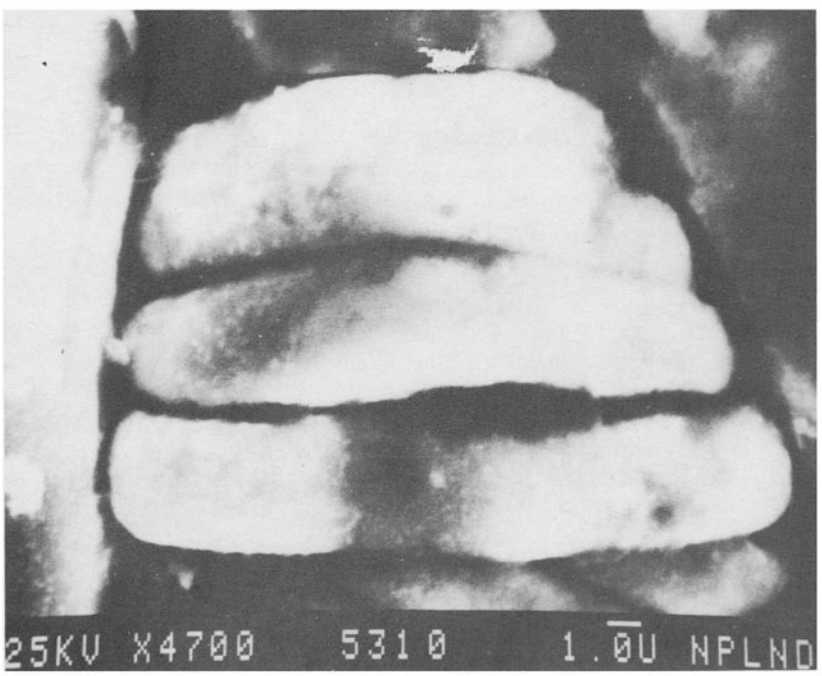

Figure 3. Typical grain structure as revealed by the SEM picture taken on a $Y_1Ba_2Cu_3O_{7-x}$ pellet.

temperature and the magnetic field. The extreme randomness of the grains seen in our samples as seen in figure 3(a) of our earlier paper (Sharma *et al* 1987) will certainly lead to a non-uniformity of the J_c value in the wire specimens. The EDS analysis shows Ba concentration higher than stoichiometric value along the grain boundaries which could be non-superconducting. Reduction in the intergrain separation and a uniform stoichiometric composition throughout the compound will have to be achieved through the adjustment of the process parameters.

One important factor leading to low current density of these wires could be the low sintering temperature used by us. It is known that a sintering temperature of 950° C leads to the highest T_c (for example Takagi *et al* 1987) and perhaps to an ordered Cu-O chains. Yet another reason for low critical current could be the cracks observed in the core of the wire. Figure 4 is a scanning electron microphotograph of the cross section of a Ag(2) wire showing cracks in the superconducting core. Further, the optical microscopic observation along the superconducting core reveals a very significant diffusion of silver into the compound. Silver is found along the grain boundaries and segregating in the voids. This is again bound to bring down the critical current density of the wire. In our systematic studies (Sharma *et al* 1987) on the addition of silver to the $Y_1Ba_2Cu_3O_{7-x}$ compound we have already shown that large addition of silver as much as 60 wt. per cent, does not destroy superconductivity as it does not alter the crystal structure of the superconducting phase. Silver thus appears to be a suitable cladding material but its diffusion into the compound during the reaction has to be prevented through the use of a suitable diffusion barrier.

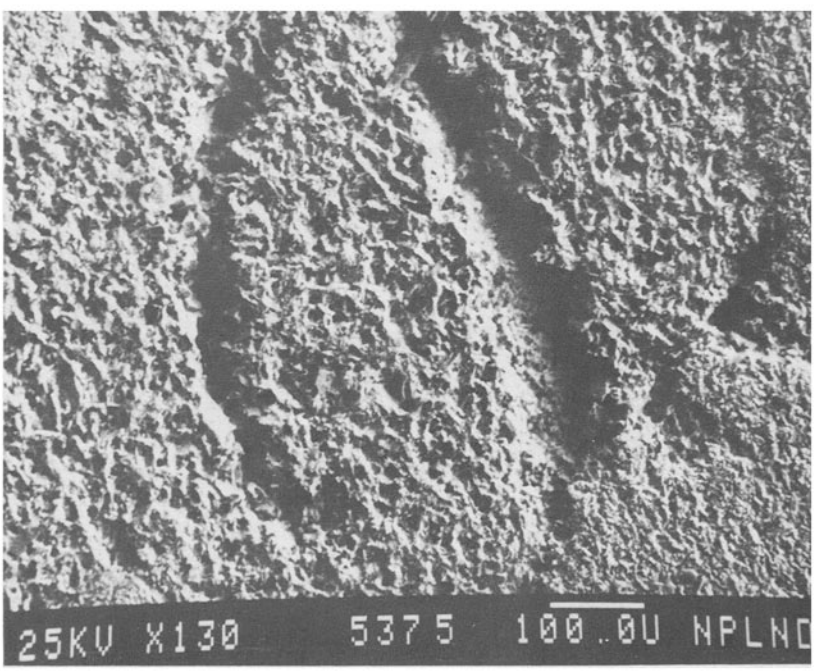

Figure 4. The SEM microphotograph obtained on the cross section of a reacted Y-26 Ag(2) wire. The cracks seen in the picture are believed to be partly responsible for low J_c.

The reacted wires are flexible. The J_c value shows a one time 25 per cent degradation when the wire is bent over a 48 mm dia mendral. Further bendings up to a dia of 30 mm produced degradation less than 2 per cent. When this bent wire was straightened again, no further degradation of J_c was observed. Detailed studies on strain effects will, however, be carried out only after a reasonable J_c is achieved in these wires.

We have thus been able to make silver clad $Y_1Ba_2Cu_3O_{7-x}$ wires which are superconducting with a midpoint T_c of 84 K. The J_c is, however, quite low limited to 26 A cm^{-2} at 77 K. The fact that the wire turns normal only at a current density of 280 A cm^{-2} indicates that it is possible to achieve at least this value of J_c through crystals and more recently in wires by the Bell Laboratory group holds out the hope that metal clad wires will be able to carry larger currents. Once a reasonable J_c is achieved in these wires, improvement through flux pinning enhancement will be possible.

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