CRITICAL PARAMETERS OF Ag-BSCCO RIBBONS

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At present many investigations have been carried out with the aim to develop technologies to produce the high-current HTSC materials. Efforts in this direction permitted to produce ribbons, wires, cables and other products for practical applications, for which the critical current $j_c \sim 10^4$ A/cm^2 at helium temperatures proved to be stable at high ($H > 20$ T) magnetic fields [1]. But for Bi-ceramics at nitrogen temperatures the state of affairs is worse, so the high enough current density of 10^4 A/cm^2 is reached in the zero magnetic field only. However, at $H = 1$ kOe the order in the current density value is lost, and at $H > 2 \div 4$ kOe for the conventional Bi-ceramics the so called "nitrogen catastrophe" takes place, characterized by an exponential rapid decrease of the critical current for H perpendicular to *ab*-plane.

As is known, ceramic-producing technologies lead to the granular products. In the conventional granular superconductors (for example the thin films) the interfaces between the granules do not limit the critical current considerably as usually the size of granules $a_0 << \xi$, where ξ is the coherency length. Therefore, the structure of the magnetic flux is determined by Abrikosov vortices even then, when intergranular layers are weak couplings. In high-temperature metaloxides the condition ($a_0 << \xi$) is valid only in the vicinity of T_c. On the another hand, many studies have shown [2] that any defect in the metalceramics destroying its uniformity, even at the atomic level, is sufficient for giving birth to a weakly coupled interface. Even quite perfect low-angle interfaces show their character of being Josephson weakly coupled. Therefore, by the traditional ceramic technologies one has no real chances to produce samples free of defects and weak couplings.

However, a possibility exists to considerably improve current properties of samples without abrupt change of the microstructure of the contacts. It is based on usage of an anisotropic growth of almost single-crystal blocks of HTSC. The technique of melt texturing growth leads to considerable improvement of the metalceramics current characteristics [3]. For such samples is typical the structure of oblong, nearly single-crystal blocks tightly connected in c direction. Intergranular interface both along *ab*-plane and normal to it possesses many defects and faults of stoichiometry in metal oxides as in the usual ceramics technology. But whence do the considerable improvements of the current characteristics of well-aligned samples follow? We presume that it is due to changes in the character in which the supercurrent flows through the samples.

In the well-aligned ceramics with the size of granules along *ab*-plane of $l_{ab} >> d_c$, where d_c is the thickness of the grain, the transport current mainly flows through the Josephson junctions in the *c*-axis direction. Namely, the coupling with neighbor granules A and B, positioned in *ab*-planes, is through the granule C connecting granules A and B. This is also called "brick-wall" geometry [4,5], or the "cable structure", where the alignment grains are closely interwoven with each other (see the insert in Fig. 1). It should be noted that the flat cut of the "cable structure" may not have the typical "brick-wall" geometry, but the current characteristics if these structures are the same.

The role of junctions, positioned in the *ab*-plane, in the two models is insignificant. The main flow of a current from *A* to *B* is along the *A*–*C*–*B*-line. Josephson junctions, arranged along this way, are formed by low-angle boundaries (normal to the *c*-axis), which corresponds to the optimum condition for the maximum Josephson critical current density j_{cj} through the boundary between ceramics blocks. Due to flow of the current along *A*–*C*–*B*-line the density of transport current J_c is increased as L_{ab}/d_c, so $J_c \approx j_{cj}(L_{ab}/d_c)$. (The ratio L_{ab}/d_c corresponds to the relation of the granule *ab*-surface area to it cross-section *S*.) As a result, the effective density of the alignment critical current J_c is increased by the coefficient $L_{ab}/d_c \sim 100$ relative to irregular ceramics. Thus, the main *ab* direction concerning the development of HTSC products with high critical currents depends on the improvement of the ceramic texture.

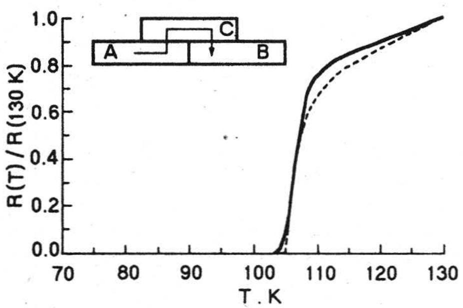
We investigated characteristics of Bi ceramic ribbons in silver sheath. The initial ceramics was prepared by the conventional method of solid-phase synthesis by annealing pressed tablets of the composition $Bi_{1.7}Pb_{0.3}Sr_2Ca_2Cu_3O_x$ at temperature 860 °C during twenty hours. After the first anneal the tablets were crushed and again pressed under pressure (~ 5 kbar). We repeated such cycle of anneal-crushing-pressing three times. After this processing the tablets contained almost 95 % of 2223-phase, as is shown by the *X*-ray diffraction.

Further, the obtained tablets were again crushed to the fine-grained powder and from the powder semi-fabricates were obtained for production of the ribbons. At this stage the main was to obtain a high degree of texture even for initial semi-fabricates. Pressing of the samples was carried out at 30–40 kbar on the anvil with a support. The annealed cooper wires were placed on a steel plate and fixed by a glue. The powder was poured between the wires. All this was covered by the second plate and then pressed. Produced by this method, samples were uniform over the cross-section and possessed sufficient mechanical strength. Depending on the final length of the ribbon ceramics, the samples were from 0.1 up to 1 mm thick with the length from 15 up to 100 mm. They were annealed at 855 °C for sixteen hours. Tests showed $J_c = (0.6 \div 1.5) \cdot 10^3$ A/cm^2 (E = 1 μV/cm, $H = 0$) and high dependence of the critical current on the magnetic field *H*. The *X*-ray studies showed irregularities in the lattice.

It is known that mechanical deformation enhances the oriented growth of granules in ceramic [3,4,6]. Therefore, we repeated pressing of samples by the method described above. The repeated pressing and annealing resulted in considerable increase of the critical current up to $J_c = (6 \div 8) \cdot 10^3$ A/cm^2 ($T = 77$ K) and the *X*-ray studies showed that at diffractograms bright reflections where seen from the (111)-planes. Simultaneously, the samples became more stable to the magnetic field.

Then the samples were enveloped into silver foils 0.1–0.3 mm thick and rolled between the rollers to reduce their thickness down to 40–50 μm. The obtained ribbon was annealed at the same temperatures as ceramic semi-fabricates. Tests of annealed ribbons showed that the samples had low densities of the critical current or, in general, had a residual resistance. We presume that it can be explained by the fact that nonuniform deformation at rolling produces a network of microcracks, which was not removed by the anneal. To cure them a hydrostatic treatment or one-axis deformation of ribbons between anvils is needed.

The best results for the critical current density were obtained for the samples treated by one-axis deformation. In Fig. 1 resistive transition to superconductivity is shown for the segment of ribbon of $30 \times 1.5 \times 0.05$ mm^3 size, the thickness of the superconducting layer is equal to 0.02 mm. This transition, in fact, coincides with the transition of a well-textured initial semifabricate to superconductivity. The value of $T_c = 110$ K corresponds to T_c of the bulk Bi-ceramic samples of 2223-phase. The results of measurements of the critical current in the magnetic field at $T = 77$ K are given in Fig. 2. The critical current density J_c for this sample at zero magnetic field was $J_c = 1.2 \cdot 10^4$ A/cm^2.



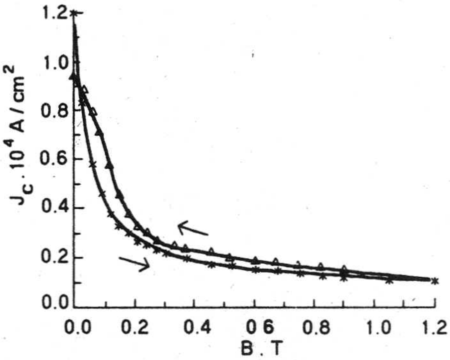


Fig. 1. Characteristic curves for two batches of samples: Bi(2223)/Ag composite ribbons and Bi(2223) ceramics (dashed curve)

Fig. 2. Hysteresis curves for Bi(2223)/Ag composite ribbon at 77 K

As is seen in Fig. 2, the J_c vs H-dependence shows a hysteresis the nature of which can be explained by availability of Josephson junctions in the current conducting cluster. It is well known [7], that the field H suppresses the Josephson current if full phase difference $\Delta\varphi$ along the contacts width W reaches value of the order of π, i.e. $\Delta\varphi \cong W(d\varphi/dx) \approx \pi$. If there are no Abrikosov vortices in granules, then derivative

$$d\varphi/dx = (4\pi\lambda^2\mu_0/\Phi_0)H/\lambda\ ,$$

so the Josephson current is small for $H >> H_0 \approx \Phi_0/(\mu_0^2\lambda W)$ (here λ is the London penetration depth into the granules). But in the presence of the Abrikosov vortices in the granules, the derivative $d\varphi/dx$ is not simple proportional to the external field H value, [8–10]

$$\Delta\varphi \approx d\varphi/dx = (2\pi\mu_0/\Phi_0)\cdot\{2\lambda^2\cdot j_{sf}(H) + tH\}\ , \qquad (1)$$

where direction X is perpendicular to the external field $H = H_z$, t is the distance between electrodes and j_{sf} is the supercurrent density on the granules surface. If there are no Abrikosov vortices in the granules, then j_s is given by the external field H, thus $j_{sf} \equiv H/\lambda$ and $\Delta\varphi \sim (2\lambda + t)H$. After the Abrikosov vortices have entered granules, the j_{sf} value is not proportional to the external field H. At high fields $H >> H_{c1}$, where H_{c1} is the lower critical magnetic field of the grains, the magnitude of the surface screening current j_{sf} in granules has two components: the equilibrium parts j_{sm} , connected with the reversible granule magnetization M_{eq} , $j_{sm} \cong - M_{eq}/\lambda$ and irreversible one, j_{cq} , determined by the magnetic flux pinning in the vicinity of the granule surface, $j_{sf} = j_{sm} \pm j_{cq}$. Here we used the boundary condition that, at the granule surface ($x = 0$) the value of induction $B(0)$ and external field H are in an equilibrium $B(0)/\mu_0 = H + M_{eq}$, M_{eq} is the equilibrium (Abrikosov) diamagnetic magnetization. For $H > H_{c1}$ the M_{eq} value is almost constant vs field H, $M_{eq} \approx H_{c1}$ and $j_{sf} \approx$ constant vs H.

Thus, the Josephson junction critical current I_c vs H depends on H very feebly, mainly on the magnetic flux penetrating into the small intergranular surface $S^* = W \cdot t$.

The sign of granular current j_{cq} depends on the magnetic history of the sample $j_{sf} = j_{sm} \pm j_{cq}$. Here the sign (+) corresponds to increase of the external magnetic field, and the sign (-) corresponds to its decrease. The variation of the surface current value with the sign of the magnetic field increment results in the abnormal hysteresis of the critical current of Josephson media [8-10]:

$$J_c(H^{\downarrow\uparrow}) = J_c\lambda\,(j_{sm} \pm j_{cq}) + (t/2\lambda)H\,, \quad \text{so} \quad J_c(H^{\uparrow}) < J_c(H^{\downarrow})\,. \qquad (2)$$

The important characteristic for understanding of the nature of the maximum critical current is the hysteresis amplitude. If, in the result of increase of temperature or the magnetic field, melting of the Abrikosov lattice in granules takes place, this immediately leads to change of the pinning component j_{cq} .

By some technological methods the pinning in the grain might be increased and, thus, the critical current hysteresis can be increased too. In fact, at decrease of temperature one observes the growth of the critical current as well as the growth of the abnormal hysteresis (compare the current hysteresis at $T = 77$ K (Fig. 2) and at $T = 4.2$ K (Fig. 3), here $J_c = 1.2 \cdot 10^5 \mathrm{A/cm^2}$).

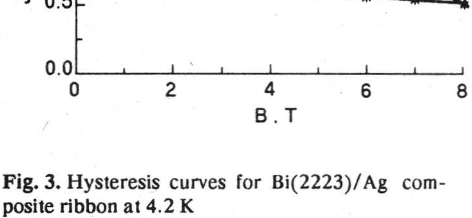


Fig. 3. Hysteresis curves for Bi(2223)/Ag composite ribbon at 4.2 K

The existence of the abnormal hysteresis indicates that transport current in the sample is determined by the Josephson weak links. But, in principle, there may be a situation when $j_{cj}(L_{ab}/d_c) > j_{cq}$, thus the transport value $J_s \approx j_{cq}$. It should be noted that the stability of Josephson critical current I_c to the high magnetic fields can be sufficiently increased if the junction had a great spatial uniformity appearing as a result of structural fluctuations, when the corresponding effective contacts dimension $r << W$ and the J_c vs H plato width is broader,

$J_c(H) \approx$ constant for external magnetic fields H in the interval:

$$B_{c1} << \mu_0 H \leq \Phi_0/(2\pi rt) + (2\lambda/t)\{B_{c1} + \lambda\mu_0 j_{cq}\}.$$

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