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Transport properties of ultrathin YBa$_2$Cu$_3$O$_{7-\delta}$ nanowires: A route to single-photon detection

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We report on the growth and characterization of ultrathin YBa$_2$Cu$_3$O$_{7-\delta}$ (YBCO) films on MgO (110) substrates, which exhibit superconducting properties at thicknesses down to 3 nm. YBCO nanowires, with thicknesses down to 10 nm and widths down to 65 nm, have also been successfully fabricated. The nanowires protected by an Au capping layer show superconducting properties close to the as-grown films and critical current densities, which are limited by only vortex dynamics. The 10-nm-thick YBCO nanowires without the Au capping present hysteretic current-voltage characteristics, characterized by a voltage switch which drives the nanowires directly from the superconducting to the normal state. We associate such bistability to the presence of localized normal domains within the superconductor. The presence of the voltage switch in ultrathin YBCO nanostructures, characterized by high sheet resistance values and high critical current values, makes our nanowires very attractive devices to engineer single-photon detectors.

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I. INTRODUCTION

Single-photon detectors based on superconducting nanowires (SNSPDs) [1] employ thin films with a thickness of a few atomic layers, in which the superconducting properties are well preserved after nanopatterning [2–4].

At present, only conventional superconductors like NbN, NbTiN, Nb, and WSi, in which reliable nanofabrication routines are well established, have been employed to realize such devices. YBa$_2$Cu$_3$O$_{7-\delta}$ (YBCO) and, more generally, high-critical-temperature superconductors (HTSs) are promising materials to use as SNSPDs since they meet all the crucial requirements for detection, such as a short coherence length $\xi \approx h v_F/\pi \Delta$ (where $v_F$ is the Fermi velocity and $\Delta$ is the superconducting gap) and, simultaneously, the highest possible transition temperature $T_C$ [5–8]. However, the serious degradation of the superconducting properties, which typically occurs in HTS thin films when the thickness approaches the nanometer scale, is a crucial issue for the realization of nanoscale devices.

The growth of ultrathin YBCO films has been the object of intense research since the discovery of cuprate superconductors. Initially, the focus was on YBCO films a few (ideally one) unit cells thick, embedded between layers of PrBa$_2$Cu$_3$O$_{7-\delta}$ (PBCO), and the main systems under investigation, precursors of two-dimensional (2D) superconductors [9], were YBCO/PBCO superlattices [10,11] and PBCO/YBCO/PBCO trilayers structures [12,13].

The choice of using PBCO in combination with YBCO was initially made because of the peculiar PBCO properties, neither metallic nor superconducting, with a crystallographic structure and lattice parameters very similar to those of YBCO. PBCO has often been used as a protective layer since bare ultrathin YBCO films are characterized by degraded superconducting properties as a consequence of the surface and chemical instability of the compound [12]. However, the deposition at high temperature of a PBCO layer may also lead to some complications, such as the interdiffusion at the interface between PBCO and YBCO, which causes the substitution of Pr atoms in Y sites, with percentages as high as 30% [14]. It is known that a 45% atomic interdiffusion into the first YBCO layer results in a $T_C$ of ~30 K for the resulting alloy [9]. Hence, few-layer YBCO thin films are characterized by a reduced $T_C$.

In this paper we show the growth and characterization of ultrathin YBCO films on MgO (110) substrates, which exhibit superconductivity at thicknesses down to 3 nm. High-quality superconducting properties have been obtained using a Au film as a capping layer. Nanowires with widths down to 65 nm have also been realized on bare 10-nm-thick films and on films capped with Au. The YBCO nanowires without Au capping show a voltage switch in the current-voltage characteristics (IVCs), similar to that of NbN nanowires, which brings the system directly from the superconducting to the normal state. This feature, occurring in structures with high sheet resistance and high critical current density, makes our ultrathin (~10 nm) nanowires ideal candidates for ultrafast single-photon detection.

II. HTSs AS MATERIALS FOR SNSPDs

YBCO and HTSs in general are of great interest in the quest for new materials for SNSPDs. The high critical temperature, $T_C^{YBCO} \approx 90$ K, could broaden the ultralow-temperature range of operation (1.7–5 K) characterizing traditional SNSPDs made of low-critical-temperature superconductors (LTSSs), such as NbN, NbTiN, and Nb. The coherence length, $\xi^{YBCO} \approx 2$ nm, which is comparable to that of traditional SNSPD...
materials, $\delta R_{\text{NbN}} \approx 5 \text{ nm}$ [15], offers the possibility to squeeze the dimensions of the detecting elements down to several nanometers without entering in a regime dominated by phase fluctuations, where dark counts would play a big role [16,17]. HTS SNSPDs can also be interesting from a more fundamental physics point of view, offering new scenarios for intrinsic mechanisms involved in the formation of the photoresponse signal. In LTS SNSPDs, models based on vortex-assisted photon counting are nowadays the most commonly accepted to describe the detection process (although none of the models proposed so far has been able to describe all the experimental data) [4,18]. Superconductivity in HTS materials and nanostructures is, however, quite different from conventional superconductors: the role of vortices could, for example, be different as a consequence of the peculiar vortex dynamics characterizing this class of materials [19].

Several works have been devoted to the realization of YBCO nanowires for their application in single-photon detectors [20–22]: despite all this technological effort, a HTS SNSPD has still not been realized. The reason for this could lie in the different material properties of YBCO with respect to LTS materials, commonly used in SNSPDs.

LTS nanowires are commonly characterized by high critical current densities, hysteretic IVCs, and high sheet resistances. High critical current densities are required to improve the signal-to-noise ratios; the amplitude of the output signals is proportional to the superconductive critical current, which is limited by the small cross-sectional area of the wire. Hysteretic IVCs ensure the switching mechanism of a SNSPD as a consequence of the photon event. The presence of the hysteresis is very common in NbN nanostructures already at thicknesses $t = 50 \text{ nm}$, much larger than those requested for their employment in SNSPDs (see Fig. 1). The hysteresis is a distinctive feature of the formation and development of a self-stabilizing hot spot in the superconducting wire [25,26], and it is a consequence of the low value of thermal conductivity in this material ($\kappa \approx 10^{-2} - 10^{-1} \text{ W m}^{-1} \text{K}^{-1}$ in the normal state). Finally, high sheet resistance values make the nanowire resistance much higher than the external load resistance $R_L$ (typically 50 $\Omega$) in parallel with the device, thus favoring the redistribution of all the bias current into the readout line, where the signal appears. In NbN, this is commonly achieved using microcrystalline, granular, ultrathin films: the normal resistance $R_N$ increases without reducing the homogeneity of the films, and nanodefects, acting as pinning centers, may even enhance the critical current density of the devices [27,28].

For an ideal YBCO-based SNSPD, the critical current density $J_C$ should be as close as possible to the depairing/vortex entry limit, which equals $J_c \approx 8 \times 10^7 \text{ A/cm}^2$ at $T = 4.2 \text{ K}$ [29,30] (see Sec. V for more details). Until now, this condition has been fulfilled only in nanowires with a thickness of several tens of nanometers [7,31]. A first limitation for the employment of YBCO nanowires with such thicknesses as a consequence of phase slips or of entry of Abrikosov vortices [33,34]. The high value of the thermal conductivity in YBCO, which is two orders of magnitude higher than in NbN [35], prevents self-heating effects that could bring the whole device from the superconducting state directly to the normal state. Indeed, the average differential resistance $\delta R$ in the measured finite-voltage range of the IVCs is much lower than the resistance $R_N$ of the nanowire, measured at the onset of the superconducting transition. Therefore, a photon event, independently of the detection mechanism which takes place, cannot drive the whole nanowire in the normal state since $\delta R \ll R_L$, preventing the proper redistribution of the bias current from the wire into the load. Another drawback in the employment of YBCO in SNSPDs comes from the impossibility to increase the sheet resistance of the nanowires by increasing the granularity of the films. This is because grain boundaries significantly reduce the critical current instead of increasing it (which happens for NbN): granular sub-100-nm-wide YBCO nanowires present $J_c$ no higher than $2 \times 10^6 \text{ A/cm}^2$ [36].

This complex scenario has been confirmed by our previous works [7,37], in which we have shown that 50-nm-thick YBCO nanowires, despite critical current densities $J_C$ close to the depairing limit, are characterized by a typical flux-flow-like behavior in the full temperature range. They have shown a purely bolometric response under irradiation.

To overcome these limitations, one needs to employ ultrathin YBCO films. Indeed, in films with thickness $t \lesssim 15 \text{ nm}$ the YBCO resistivity increases [38–40]; therefore, the thermal conductivity, via the Wiedemann-Franz law, decreases. Nanowires patterned on these films can be characterized by a physics close to that describing NbN nanowires, giving rise to hysteretic IVCs. The issue is therefore to reduce the thickness of YBCO nanowires, keeping the superconducting properties as close as possible to the bulk material. Indeed, as a consequence of the chemical instability of this material,

![Fig. 1. Current-voltage characteristics at $T = 4.2 \text{ K}$ of a 50-nm-thick, 75-nm-wide, and 120-nm-long NbN nanowire (blue) and of a bare YBCO nanowire (orange) having the same dimensions. Both the nanowires have been realized following the nanopatterning procedure described in Refs. [23,24]. In NbN, a large-voltage switch, occurring just above the critical current, brings the whole nanowire into a state characterized by a differential resistance corresponding to the normal resistance $R_N$. In YBCO, a flux-flow behavior, characterized by differential resistances $\ll R_N$, dominates at currents higher than the critical one.](image-url)
used as reference structures for the maximal 50-nm-thick nanowires [31], can preserve the superconducting properties already successfully used on nanowires, with thickness down to 10 nm, to show that our nanopatterning procedure, with this thickness, we present the structural and transport characterization of ultrathin YBCO nanowires have led to critical currents for the sub-100-nm-wide bridges more than one order of magnitude lower than \( J_c \). This effect can be due to the reduced Joule heating in the nanowires and therefore the probability to get hysteretic IVCs. Previous attempts to realize ultrathin YBCO nanowires have led to critical currents for the sub-100-nm-wide bridges more than one order of magnitude lower than \( J_c \): as a consequence, despite the small cross sections, the IVCs are still flux-flow-like [20], or they show voltage switches only in the resistive state [21].

The structure of this paper is as follows. In Secs. III and IV we present the structural and transport characterization of ultrathin YBCO films. We show that, despite the lack of PBCO encapsulation, commonly used to protect thin YBCO, the film properties are intrinsic of the material, confined at thicknesses of a few nanometers. In Sec. V we study Au-capped YBCO nanowires, with thickness down to 10 nm, to show that our nanopatterning procedure, already successfully used on 50-nm-thick nanowires [31], can preserve the superconducting properties also in ultrathin structures. These nanowires are used as reference structures for the maximal \( J_c \) value one can achieve at this thickness. Finally, in Sec. VI we focus on bare 10-nm-thick YBCO nanowires to show their potentialities as SNSPDs.

### III. ULTRATHIN FILMS: GROWTH AND STRUCTURAL CHARACTERIZATION

We deposited YBCO films, with thicknesses from 50 down to 3 nm, on MgO (110) substrates by pulsed laser deposition (heater temperature of 760 °C, oxygen pressure of 0.7 mbar, energy density of 2.1/cm²). After the deposition, the films were slowly cooled down (cooling rate of 5 °C/min) at an oxygen pressure of 900 mbar to promote full oxidation of the YBCO chains. With this procedure, we have previously shown that 50-nm-thick YBCO films are slightly overdoped [41]. For each thickness, we have deposited two YBCO films: one with a 50-nm-thick Au capping layer, in situ sputtered after the YBCO deposition, and another without the protective layer.

A morphological and structural characterization was carried out using atomic force microscopy (AFM) and scanning electron microscopy (SEM) to establish the quality of the YBCO films. The films present smooth surfaces, characterized by the typical \( c \)-axis domains with three-dimensional (3D) spirals and an average roughness which is of the order of one atomic cell (see Supplemental Material, Sec. S1 [42]).

The structural properties were determined by x-ray diffraction (XRD) analysis with a four-circle Panalytical X’pert diffractometer with Cu \( K_\alpha \) radiation using a hybrid Ge(220) monochromator and a PIXcel 3D detector matrix. Symmetric 20-20 scans show that the ultrathin films are highly crystalline and \( c \)-axis oriented [see Fig. 2(a)]. The crystallographic quality of the films is confirmed by the presence of the interference fringes which are visible on both sides of the lowest-diffraction-order (00n) YBCO Bragg peaks [see Fig. 2(b)] in films with thicknesses down to 20 nm. The thickness \( t \) was calculated, with an error of less than one atomic layer (~1 nm), by considering couples of adjacent maxima and using the relation \( t = \frac{\lambda_{Cu}}{2\sin(\omega_n - \sin(\omega_{n-1}))} \) (where \( \omega_n = 2\theta_n / 2 \) are the angles corresponding to the maxima of the interference fringes and \( \lambda_{Cu} = 1.540598 \) Å is the wavelength of the incident x-ray wave).

From the angular position of the (00n) peaks we can estimate the \( c \)-axis length. On the basis of the bulk lattice...
parameters of YBCO and MgO (110), a large in-plane lattice mismatch $\delta^m$ (defined as $|a_{sub} - a_{film}|/a_{sub}$) is expected between the films and the substrate ($\delta^m$ is $\sim 9\%$ and $\sim 35\%$ along the [001] and [1 $-$ 1 0] MgO directions, respectively) [43]. In ultrathin YBCO films, especially when grown on substrates with a large mismatch, a c-axis expansion has usually been reported [44] and attributed to a large number of misfit dislocations caused by the extreme strain induced by the substrate on the whole film thickness. However, in our films, in spite of the large mismatch between the films and the substrates, no c-axis expansion is measured. In more detail, our films show a c-axis length of $\sim 11.715$ Å, which is sizably smaller than those we have measured in optimally doped films grown on MgO (110) substrates, corresponding to $\sim 11.74$ Å [see Fig. 2(c) and the inset therein]. Interestingly, films below 30-nm thickness have a slightly shorter c axis (i.e., 11.705 Å) than the reference slightly overdoped 50-nm-thick films [see Fig. 2(c) and the inset therein]. Such a small difference, compatible with a higher oxygen content, might be related to the efficiency of the postannealing in incorporating oxygen in films of only a few unit cells.

Cross-sectional high-resolution transmission electron microscopy (TEM) and high-angle annular dark-field scanning TEM (HAADF-STEM) investigations performed on 10-nm-thick films unveiled a complex nanostructure of the films characterized by a high density of Y$_2$Ba$_4$Cu$_8$O$_{16}$ (Y248) intergrowths intercalating within the YBCO matrix. Figure 3(a) shows a high-resolution Z-contrast image of the YBCO film in which the presence of Y248 planar faults is observed (indicated by arrows). In particular, as highlighted in the enlarged image [see Fig. 3(b)], the Y248 intergrowths are sometimes accompanied by a rigid shift of half the unit cell along the b axis of YBCO; hence, one can easily identify the [010] and [100] zone axes of the Y248 structure as a consequence of YBCO twinning. As a result of the presence of a high concentration of intergrowths, a waving of the (001) planes is observed with a consequent break in the vertical coherence which extends over the whole film thickness. The same nanostructural assessment is observed in the case of the 50-nm-thick films (not shown here). The presence of Y248 planar faults in YBCO thin films has been previously reported [45].

The Y248 intergrowths introduce additional CuO chains into the structure, making our thin films Cu rich and overdoped at all thicknesses. This latter characteristic has been additionally confirmed by a transport characterization of the films, as will be discussed in the next section.

IV. ULTRATHIN FILMS: TRANSPORT CHARACTERIZATION

Transport properties of the ultrathin YBCO films have been studied using resistance vs temperature $R(T)$ measurements. For ultrathin YBCO films grown on substrates with a large mismatch (without a PBCO seed layer), a strong reduction of the zero-resistance temperature $T^0_C$ has been reported [46]. Here, we show instead that the $T^0_C$ of our films is higher than the liquid-nitrogen temperature down to thicknesses of 10 nm. From the onset of the superconducting transition and the normal-state properties of the ultrathin films we are able to extract information about the oxygen doping of nanometer-thick YBCO films.

A. Determination of the oxygen doping

In Figs. 4(a) and 4(b) the $R(T)$ measurements of the bare 10- and 20-nm-thick films are compared with those of the 50-nm-thick film (also without Au capping).

The onset temperature of the superconducting transition $T^{\text{onset}}_C$ slightly changes with the film thickness [see inset table in Fig. 4(a)], while $T^0_C$ substantially decreases with reducing the thickness [due to the significant broadening of the superconducting transition; see Fig. 4(a)].

In principle a lower $T^{\text{onset}}_C$ could be associated with both an increase and a reduction of the oxygen doping within the film. In our case, we can attribute the $T^{\text{onset}}_C$ reduction to better oxygenation of the ultrathin films. This is clear from the normal-state $R(T)$ dependence of these films [see Fig. 4(b)] and from the comparison with $R(T)$ of an optimally doped 50-nm-thick film [see inset of Fig. 4(a)].

The optimally doped films are characterized by a linear behavior of the resistance as a function of the temperature in the full temperature range down to $T^{\text{onset}}_C$, as expected at this oxygen doping [47] [see inset of Fig. 4(a)]. The $R(T)$ measurements of the other three films are instead linear only at high temperature, while they exhibit a power-law dependence, with the same exponent $n = 1.2$, below a temperature $T \approx 200$ K. This behavior is typical of fully oxygenated YBCO films lying in the overdoped side of the phase diagram. The overdoped region of HTS materials is characterized by the so-called coherence temperature $T_{coh}$ (a crossover temperature from a coherent and an incoherent metal state [48]), which increases as the oxygen doping increases: below $T_{coh}$ the dependence of the resistance with the temperature is not linear.
any longer, and it becomes power-law-like, \(R(T) = a + bT^n\). The exponent \(n\) varies from 1 at the optimal doping to 2 at the oxygen doping corresponding to the disappearance of the superconducting phase [47,49,50]. As shown in the inset table of Fig. 4(b), \(T_{coh}\) slightly increases by 10 K when the film thickness is reduced to 10 nm, signifying an oxygen content at least comparable to that of 50-nm-thick films, if not slightly higher than that.

The \(R(T)\) of the films was measured with the Van der Pauw method [51] to extract the sheet resistance \(R_{\parallel}\) and the resistivity \(\rho = R_{\parallel}/t\) and to determine their dependence on the film thickness. The resistivity is thickness independent in films down to 20 nm, with a value \(\rho \approx 60 \mu\Omega\cdot\text{cm}\) at \(T = 100\) K, which is comparable to that obtained from YBCO single crystals [52]. In 10-nm-thick films \(\rho(100\) K) is instead \(\sim 130 \mu\Omega\cdot\text{cm}\): such an increase is expected since in the literature \(R_{\parallel}\) is reported to scale faster than \(1/t\) in sub-15-nm-thick films [38-40], where a significant drop in \(T_c^0\) occurs (see the next section).

### B. Broadening of the superconducting transition

In Fig. 5 we plot (red circles) the \(T_c^0\) extracted by our YBCO films (bare films above \(t = 10\) nm and Au-capped films below \(t = 10\) nm) as a function of the thickness. As discussed earlier, \(T_c^0\) slightly decreases down to 5 nm, which corresponds approximately to four unit cells. For lower thicknesses, it abruptly drops. Films with thickness of 3 nm, where only two complete atomic layers are present, are still superconducting with \(T_c^0 = 14.5\) K. At even lower thicknesses the films are not superconducting down to 4.2 K.

We compare our results with those published in the literature starting from the beginning of the 1990s up to now. As previously mentioned, most of the works have focused on YBCO films sandwiched between two PBCO layers [12-14,39,53,54]. The \(T_c^0\) values as a function of the YBCO thickness, collected from these papers, are shown as open symbols in Fig. 5. In all the previous works, \(T_c^0\) gradually decreases down to four-unit-cell-thick (\(t = 5\) nm) YBCO films while presenting a sudden drop at lower thicknesses. The same behavior can also be seen in superlattices when the YBCO layers are interchanged with either PBCO layers [55] or SrTiO\(_3\) layers [56]. Below 5-nm-thick films, the spread of \(T_c^0\) values which have been measured in different works becomes wider. The origin of this spread can have several possible reasons, related to an error in the determination of the thickness and/or to small differences between different samples (in the interdiffusion rate between Pr and Y atoms at the interfaces or in the PBCO stoichiometry and/or the presence of crystalline defects) [57].
Our films (solid circles in Fig. 5), capped at most with a Au film, are characterized by $T_C^n$ values which are comparable, down to 5 nm, to the best results of PbCO/YBCO/PbCO multilayers (open triangles), while $T_C^n$ of the very thin films are the best reported values for bare YBCO structures.

The wide broadening of the superconducting transition (defined as $T_C^{\text{onset}} - T_C^n$), significantly suppressing $T_C^n$ of nanometer-size thin films, has been commonly attributed to an intrinsic size effect of the material and explained in terms of the Kosterlitz-Thouless (KT) transition related to vortex-antivortex pair dissociation in two-dimensional systems [53,58,59]. Even though such an interpretation has not achieved a general consensus [13,60], other mechanisms implicating transfer of charge carriers from the YBCO layer to with Au for thicknesses below 10 nm).

$k_F \ell$

$k_F \ell$ not achieved a general consensus [13,60], other mechanisms have been associated with a moderate disorder where a weakening of the pairing interactions is present but the role of phase fluctuation is still very limited [63]. The disorder becomes much stronger, and the phase fluctuations acquire a dominant role in films with thickness $t < 10$ nm. Here, we cannot make a quantitative study since the measured resistivities are much stronger, and the phase fluctuations acquire a dominant role of disorder in our YBCO films as a function of the thickness, and the increase in resistivity exhibited by the ultrathin films are expected to drop even further, and the films characterized by dimensions $w, t \gg 4.44 \xi_0$ (where $\xi_0 \approx 2$ nm for YBCO is the superconducting coherence length at zero temperature), the maximum $J_C$ is given by the entry of Abrikosov vortices, driven by the Lorentz force, overcoming the bridge edge barrier. In this regime $J_C$ (critical current density due to vortex entry) can approximately be written as [29,30]

$$J_C \approx 0.826 J_d,$$

where $J_d$ is the depairing critical current density, predicted by the Ginzburg-Landau theory, whose expression [66]

$$J_d = \frac{\Phi_0}{3\pi^{3/2} \mu_0 \lambda_0^2 \xi_0}$$

equals to $1-3 \times 10^8$ A/cm$^2$ for YBCO nanowires [23,24,31]. In Eq. (2), $\Phi_0$ is the flux quantum, $\mu_0$ is the vacuum permeability, and $\lambda_0$ is the London penetration depth at zero temperature. The maximum $J_C$ value of our nanostructures is therefore close to the theoretical limit, showing that superconducting properties close to the as-grown films are preserved even at ultrathin thicknesses.

The resistance as a function of the temperature close to the transition is shown in Fig. 6(c) for three nanowires, with thicknesses $t = 10, 20$, and 50 nm. For each device, the first transition in $R(T)$, at higher temperature, is related to the wide electrodes connected to the nanowire; the temperature range of this transition corresponds to that of the unpatterned film [seen in Fig. 4(a)], although the onset $T_C^{\text{onset}}$ is lower because of the shunt effect of the Au layer [23]. The second, broader transition in $R(T)$, at lower temperature, is instead associated with the nanowire: what is more remarkable about this transition is that the onset temperature $T_C^{\text{onset}}$ is typically only 1–2 K lower than that of the wide electrodes $T_C^{\text{onset}}$, independent of the thickness.

The broadening of the superconducting transition of the thinnest nanowires is not much wider than that of the 50-nm-thick nanowires. However, this broadening cannot be ascribed only to the dissipation induced by Abrikosov vortices crossing the nanowires, shown for the 50-nm-thick nanodevices in other
VI. ULTRATHIN BARE NANOWIRES: A POSSIBLE PLATFORM FOR YBCO SINGLE-PHOTON DETECTORS

For most applications requiring ultrathin superconducting devices, a Au capping layer is not ideal. For single-photon detection applications, for example, the high conductivity of Au leads to a high reflection of the light. At the same time, the Au capping layer works as an electrical shunt for the photoreponse and could also represent a thermal shunt for the Au capping layer works as an electrical shunt for the photoreponse and could also represent a thermal shunt for the Au capping layer. This may lead to possible deoxygenation from the nanowire sidewalls during the nanopatterning and the occurrence of some randomly distributed inhomogeneities.

Transport measurements were done on 20-nm-thick bare YBCO nanowires, characterized by resistivity values very close to those of 50-nm-thick wires. From the IVCs at $T = 4.2$ K of the nanostructures, we have extracted $J_c$ up to $5 \times 10^7$ A/cm$^2$, the average critical current density of the reference Au-capped nanowires (see Sec. V). However, a voltage switch from the superconducting to the normal state has never been observed.

In the following we show that bare nanowires, with thickness $t \approx 10$ nm, are instead characterized, similar to NbN nanowires, by hysteretic IVCs, with a voltage switch driving the structures abruptly from the superconducting to the normal state as soon as the critical current is exceeded. Current-voltage characteristics and resistance versus temperature measurements were carried out on 38 nanowires with widths in the range 65–120 nm and lengths in the range 100–500 nm and patterned on several 10-nm-thick films.

From the IVCs, measured at $T = 4.2$ K in the current bias mode, we have inferred critical current densities as high as $J_c \approx 3.5 \times 10^7$ A/cm$^2$, quite close to the average critical current density of the Au-capped nanowires (which are reference nanowires). As a consequence of the choice of carbon as a capping layer, we have obtained only a slightly lower $\bar{J}_c$, but with a larger spread of $J_c$ values (see Supplemental Material, Sec. S2 [42]), and broader resistive transitions compared with nanowires capped with Au [see Fig. 7(b)].

The IVCs of the nanowires with $J_c$ lower than $\sim 5 \times 10^6$ A/cm$^2$ are flux-flow-like (see Supplemental Material, Sec. S2 [42]). The shape of the IVCs instead abruptly changes for nanowires with $J_c$ higher than this threshold. A large voltage switch, of the order of several millivolts, appears above $I_c$, which drives the devices to a state characterized by a constant differential resistance, $dV/dI \approx 1–4 \times 10^2$ $\Omega$, up to $V \approx 20$ mV when applying a bias current $I_b > I_c$ [see Fig. 7(a)]. The difference between the switching ($I_c$) and re-trapping ($I_r$) currents is on the order of several microamperes. The fact that we observe the voltage switch only in nanowires with the highest $J_c$, approaching the maximum current density
10-nm-thick wires might be related to the increase in the film diffusion. The occurrence of the switch in the IVCs of NbN [26,75,76] and cuprate nanowires (see inset). The IVC presents an excess current $I_{exc}$, possibly deriving from Andreev reflections at the interface between the superconductor and the normal region originating from a hot spot [70]. (b) $R(T)$ measurement of the same nanowire. The value of the normal resistance, corresponding to $T_C^{exc}$ of the nanowire, is $R_N \approx 100 \ \Omega$, very close to the $dV/dI$ value measured in the IVC.

FIG. 7. (a) The IVC measured at $T = 4.2$ K of a 75-nm-wide and 100-nm-long bare nanowire ($J_C = 1.8 \times 10^7$ A/cm$^2$) presents a 2-mV-wide voltage switch above $I_C$, which drives the system to the normal state, corresponding to a differential resistance $dV/dI = 112 \ \Omega$ (see inset). The IVC presents an excess current $I_{exc}$, possibly deriving from Andreev reflections at the interface between the superconductor and the normal region originating from a hot spot [70]. (b) $R(T)$ measurement of the same nanowire. The value of the normal resistance, corresponding to $T_C^{exc}$ of the nanowire, is $R_N \approx 100 \ \Omega$, very close to the $dV/dI$ value measured in the IVC.

a YBCO nanowire can carry, is an indication that the switching is not caused by the presence of local grain boundaries or defects.

The normal resistance $R_N$ (the value of the resistance at the onset of the superconducting transition) of the nanowires is very close to the differential resistance $dV/dI$ measured in the IVCs at 4.2 K in all the investigated samples [see Fig. 7(b)], in agreement with what is observed in NbN nanowires (see Fig. 1). This occurrence supports the hypothesis that the voltage switch above $I_C$ is a sign of self-heating in the nanowires: a normal/conducting domain, induced inside the ultrathin wire, drives the system directly from the superconducting to the normal state. This hypothesis is further supported by the presence of excess current in the hysteretic IVCs of the nanowires, again in accordance with results for NbN nanowires (see Fig. 1). The excess current, corresponding to the extrapolation to zero voltage of the normal branch at high bias, derives from Andreev reflections at the two interfaces between the superconductor and the normal region originating from a self-stabilizing hot spot [70].

It is worth mentioning that the self-heating hot-spot model is simplistic, and it is not meant to describe the detection mechanism in these nanowires: in type-II superconductors, like NbN and YBCO, magnetic vortices play an important role (as already highlighted in Sec. V) [29]; in addition to this, the physics of ultrathin films and nanowires is characterized by thermal and quantum phase slips [71,72]. Nevertheless, NbN nanowires still work as SNSPDs whether their physics is dominated by phase slips or vortices [4,73]. The understanding of the processes responsible for the formation and development of the normal/conducting domain within our nanowires upon irradiation goes beyond the purpose of the present paper.

Qualitatively, within the simplistic framework of the self-heating hot-spot model [74], which recently has even been used to describe the IVCs of NbN [26,75,76] and cuprate HTS [77,78] nanowires, the central part of the wire becomes normal due to Joule heating, producing a power equal to $IV$. A cooling mechanism follows via phonons and electron diffusion. The occurrence of the switch in the IVCs of 10-nm-thick wires might be related to the increase in the film resistivity in YBCO below 15 nm [38–40]. As mentioned in Sec. IV, in our 10-nm-thick nanowires we have measured at $T_C$ a resistivity $\rho$ which is more than a factor of 2 higher than the $\rho$ of our 50- and 20-nm-thick nanowires. Indeed, an increase in resistivity would result in an increase of Joule heating, while at the same time the thermal conductivity $\kappa$ is decreased via the Wiedemann-Franz law, $\kappa(T) = \mathcal{L}T/\rho$, where $\mathcal{L} = \pi^2 k_B^2/3e^2 \approx 2.45 \times 10^{-8} \ \text{W} \Omega/\text{K}^2$ is the Lorentz number. Both factors would support the stabilization of a hot spot in our ultrathin nanowires. It is worth mentioning that in NbN wires, the value of the retraction current $I_r$, which within the self-heating hot-spot model depends on the $\rho$ of the material, on the thickness, and on the thermal coupling between the wire and the substrate, has been experimentally found to decrease with decreasing thickness [76]. This implies that the stabilization of a hot spot is favored in ultrathin wires.

Finally, we have measured the IVCs of the 10-nm-thick bare YBCO nanowires as a function of the temperature up to $T_C$: the switching current $I_C$ decreases more rapidly than the retraction current $I_r$ when increasing the temperature, leading to a gradual narrowing of the hysterisis (see Fig. 8).

FIG. 8. Experimental critical current $I_C$ (circles) and retraction current $I_r$ (squares) as a function of the temperature for the nanowire shown in Fig. 7. The solid line is a guide for the eye.
particular, the hysteresis is almost negligible if \( T > 0.4T_c \). In NbN nanowires the hysteresis, which also decreases with increasing temperature [79], can instead persist up to temperatures close to \( T_c \). In our case, the narrower \( T_c / T_c^0 \) range in which \( I_c > I_r \) might be due to the smaller value of \( \kappa \) of YBCO with respect to that of NbN. This is also evident from the shape of the IVCs at \( T = 4.2 \) K, presenting hysteresis, i.e., values of the retraction voltage and difference between \( I_c \) and \( I_r \) much smaller than those of NbN nanowires (see Fig. 1).

VII. SUMMARY AND CONCLUSIONS

We have shown the realization of YBCO films which exhibit superconductivity down to 3 nm. The films show a \( c \) axis, close to the reference 50-nm-thick films, still corresponding to a slightly overdoped regime and signifying a high crystallinity down to two to three unit cells. Moreover, the \( T_c^0 \) of the thinnest films, obtained by using Au capping, is comparable, down to four unit cells, to the best value obtained by sandwiching YBCO films with PBCO. Since the presence of the PBCO capping and buffering induces substitutions and/or chemical modifications, our results can be considered a first step toward the realization of few-unit-cell structures with properties representative of the YBCO bulk.

We have also fabricated YBCO nanowires with thicknesses down to 10 nm and widths down to 65 nm, both with and without the protection of a Au capping layer. The ultrathin Au-capped YBCO nanowires \( (t = 10 \) nm) behave similarly to the 50-nm-thick nanowires and are characterized by superconducting properties not affected by the nanopatterning procedure up to their \( T_c^0 \), a value close to the \( T_c^0 \) of the unpatterned films. The IVCs are nonhysteretic. The ultrathin 10-nm-thick bare YBCO nanowires are instead characterized by hysteretic current-voltage characteristics, with a voltage switch of the order of a few millivolts from the superconducting to the normal state. In addition to this bistability of the IVCs around \( I_c \), in our bare nanowires coexist (1) very small cross sections with sub-100-nm widths and thicknesses not far from the superconducting coherence length; (2) high critical current densities, approaching the critical depairing value, limited by vortex entry; and (3) enhanced sheet resistances, expected in sub-15-nm-thick films and nanostructures, which make it possible to engineer sub-micrometer-long wires with \( \kappa \Omega \) resistance values which are much larger than the 50 \( \Omega \) resistance used in parallel with the device in high-speed SNSPD readout circuits. The coexistence of these four features makes our ultrathin YBCO nanowires very attractive for SNSPDs.

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