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Improvement of the critical temperature of NbTiN films on III-nitride substrates

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Abstract

In this paper, we study the impact of using III-nitride semiconductors (GaN, AlN) as substrates for ultrathin (11 nm) superconducting films of NbTiN deposited by reactive magnetron sputtering. The resulting NbTiN layers are (111)-oriented, fully relaxed, and they keep an epitaxial relation with the substrate. The higher critical superconducting temperature ($T_c = 11.8$ K) was obtained on AlN-on-sapphire, which was the substrate with smaller lattice mismatch with NbTiN. We attribute this improvement to a reduction of the NbTiN roughness, which appears associated with the relaxation of the lattice misfit with the substrate. On AlN-on-sapphire, superconducting nanowire single photon detectors were fabricated and tested, obtaining external quantum efficiencies that are in excellent agreement with theoretical calculations.

Keywords: NbTiN, superconductor, GaN, AlN, single photon detector

(Some figures may appear in colour only in the online journal)

1. Introduction

Superconducting ultrathin (few-nm size) films are the key-stone of devices such as superconducting hot electron bolometer mixers [1–4] or superconducting nanowire single photon detectors (SNSPDs) [5–9]. Furthermore, the integration of such superconducting layers in semiconductor structures has been explored as an approach to improve the performance and add new features to high electron mobility transistors [10], and could grant access to interesting physical phenomena involving Majorana fermions [11] and topological insulators [12]. Niobium titanium nitride (NbTiN) is widely used in applications requiring nm-size films due to its physical and chemical stability. In comparison with NbN, NbTiN devices have demonstrated higher optical quantum efficiency, smaller recovery time, and reduced noise [13, 14]. The origin of these improvements remains partially unclear, though it is generally attributed to a better crystalline quality, lower atomic concentration of oxygen, and higher metallic

conductivity, which results in lower kinetic inductance [13, 15].

The deposition of crystalline Nb(Ti)N materials by molecular-beam epitaxy [10] or chemical vapor deposition [16] requires relatively high substrate temperatures (≈ 800 °C and 900 °C–1300 °C, respectively). Therefore, compatibility with device processing imposes the deposition by sputtering, where polycrystalline ultrathin films with excellent morphology can be obtained in the temperature range of 25 °C–300 °C [17–19]. However, the superconducting performance of NbTiN layers, namely the critical temperature at which the phase transition from resistive to superconducting state occurs, depends on the layer thickness and its structural quality.

The most stable crystalline configuration of NbTiN is the rock-salt cubic structure, with a lattice parameter between $a_{\text{NbN}} = 4.39$ Å [17] and $a_{\text{TiN}} = 4.24$ Å [20]. The use of III-nitride semiconductors (GaN, AlN) as substrates for NbTiN is interesting because the equivalent lattice constant of NbN in

the (111) plane $a_{\text{NbN}}^{(111)} = 3.104 \text{ \AA}$ is closely lattice matched to the in-plane lattice constant of wurtzite AlN ($a_{\text{AlN}} = 3.112 \text{ \AA}$, mismatch $< 1\%$) and GaN ($a_{\text{GaN}} = 3.189 \text{ \AA}$, mismatch $\approx 3.5\%$). These III-N materials present the additional interest of being optically transparent in a large spectral range, from their band gap (200 nm and 365 nm for AlN and GaN, respectively) to the far infrared, with the exception of the 9.6–19 and 7.3–9 μm regions that correspond to the first and second harmonics of the reststrahlen band [21, 22].

It has been reported that the use of crystalline GaN, AlGaN or AlN as substrates results in an improvement of the crystalline quality and critical superconducting temperature of Nb(Ti)N thin films, which grow epitaxially on (0001)-oriented III-N [16–18, 23]. However, there is almost no information on the performance of SNSPDs fabricated on III-nitrides. In a conference contribution, Zhu *et al* showed NbN SNSPDs on crystalline AlN-on-sapphire with reduced jitter but poor detection efficiency ($< 4\%$) [24]. On the other hand, the use of sputtered (polycrystalline) AlN as a buffer layer for the deposition of NbN on GaAs substrates resulted in improved performance of SNSPDs in comparison with detectors fabricated on bare GaAs [25]. However, the efficiency of such devices was only 5% at 900 nm. It increased up to 31% at 385 nm, but the substrate absorption at that wavelength would prevent the implementation of the microcavity or waveguide based designs required to get close-to-unity efficiencies [26].

In this paper, we study the impact of the substrate on the crystalline quality and superconducting critical temperature of ultrathin (11 nm) films of NbTiN deposited on almost-lattice-matched III-nitride semiconductors by reactive magnetron sputtering. Various materials are considered as substrates, namely a 1 μm thick AlN-on-sapphire template, a GaN/AlN (650 nm/1 μm) heterostructure on sapphire, and 350 μm thick GaN substrates, either non-intentionally doped (n-type conductive) or Fe-doped (semi-insulating).

2. Methods

Thin ($\approx 11 \text{ nm}$) NbTiN films were deposited by reactive magnetron co-sputtering at room temperature in an ultra-high vacuum chamber (background pressure between 5×10^{-9} Torr and 1×10^{-8} Torr). After solvent cleaning (acetone, methanol), the substrates were placed on a 4" wafer chuck and held in place using aluminum pins. During the deposition, the substrate holder was rotating in order to improve the radial uniformity of the samples. The substrates were held at a distance of about 30 cm from the targets. The Nb target was placed directly below the substrate holder, while the Ti target was placed in a confocal geometry, the target pointing towards the substrate holder. Both targets were 3" in diameter and 0.250" thick. The applied power on the Nb target was 200 W (direct current), while the applied power on the Ti target was 200 W (radiofrequency), which corresponds to a power density of 4.39 W cm^{-2} . The sputtering pressure was set to 3 mTorr with a gas mixture of 100 sccm of Ar and 10 sccm of N_2 . A quartz crystal monitor was calibrated by deposition of a thick film and subsequent measurement of the

deposited thickness with a profilometer. From measurements on previous samples deposited in the same conditions, we estimate the thickness variation in $1 \times 1 \text{ cm}$, to be less than 1%. The estimated deposition rate was 0.1 nm s^{-1} . The samples were deposited two at a time during subsequent runs, and under similar conditions, namely same pressure, gas partial pressures, applied powers, plasma voltage and current. The nominal Ti mole fraction in the $\text{Nb}_{1-x}\text{Ti}_x\text{N}$ layer was $x = 0.2$, but chemical measurements by electron energy loss spectroscopy and energy-dispersive x-ray spectroscopy of reference specimens, performed in a transmission electron microscope, point to $x = 0.15 \pm 0.03$. Substrates consisted of:

- (i) a commercial template (AlN-on-sapphire template) consisting of 1 μm thick (0001)-oriented AlN deposited by metalorganic vapor phase epitaxy (MOVPE) on c-plane sapphire (sample SA);
- (ii) a 650 nm-thick (0001)-oriented GaN layer deposited by plasma-assisted molecular-beam epitaxy (PAMBE) on a commercial template consisting of 1 μm thick AlN deposited by MOVPE on c-plane sapphire (sample SAG);
- (iii) a 350 μm thick (0001)-oriented GaN substrate (sample SG) diced from thick GaN boules grown by hydride vapor phase epitaxy on c-plane sapphire; and
- (iv) a 350 μm thick (0001)-oriented semi-insulating Fe-doped GaN substrate (sample SGF) diced from thick Fe-doped GaN boules grown by hydride vapor phase epitaxy on c-plane sapphire.

Prior to the deposition, the surface was treated *in situ* with a 25 W radiofrequency Ar plasma for 10 s. The samples under study are listed in table 1.

X-ray reflectivity and x-ray diffraction measurements (XRD) were performed in a Panalytical Empyrean diffractometer equipped with a cobalt anode beam tube ($\text{Co } \lambda_{\text{K}\alpha} = 1.789 \text{ \AA}$), a Göbel mirror and a two-dimensional pixel detector used in both zero-dimensional and one-dimensional modes.

The microstructure of the samples was analyzed by high-resolution transmission electron microscopy (HRTEM). The samples were prepared in cross-section by either mechanical polishing followed by ion milling or by focused ion beam milling and measured in a FEI-Tecnai microscope operated at 200 kV. Finally, the surface morphology was measured by atomic force microscopy (AFM) in a Dimension 3100 system operated in the tapping mode.

3. Results and discussion

3.1. Material characterization

The thickness of the NbTiN layer has been determined precisely using x-ray reflectivity. Examples of these measurements corresponding to samples SA and SGF are illustrated in figure 1, together with fits using the GenX software [27]. For

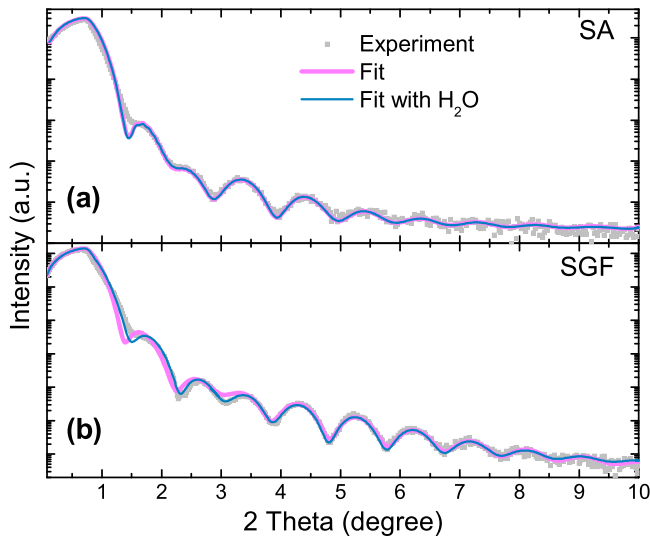


Figure 1. X-ray reflectivity of (a) the NbTiN film on AlN-on-sapphire (sample SA), and (b) the NbTiN film on GaN:Fe (sample SGF). Grey squares are experimental measurements and the solid lines are fits generated with the GenX software. Note that in the case of SA, fits with and without H₂O wetting layer are almost superimposed.

all the samples, the fit confirms total layer thickness around 11 nm (see precise values and error bars in table 1). Getting a good agreement with the experimental results requires assuming a multilayer structure consisting of a bottom Nb_{0.8}Ti_{0.2}N layer ($\approx 8.45 \pm 0.45$ nm) capped by an Nb_{0.8}Ti_{0.2}N_{0.9}O_{0.1} layer ($\approx 2.1 \pm 0.2$ nm) and by a mixture of Nb₂O₅ and TiO₂ (less than one atomic layer, with (Nb₂O₅)_{0.7}(TiO₂)_{0.3} composition). Post-growth oxidation and Nb₂O₅ and TiO₂ demixing are known to occur in NbTiN thin films, but experiments as a function of time show a saturation so that it remains a surface phenomenon [28]. Furthermore, due to the hydrophilic nature of the material, a further improvement of the fit is observed when assuming the presence of a layer of H₂O at the surface (see figure 1). The values of root-mean-square (RMS) roughness of the topmost layer extracted from the calculations are also summarized in table 1, considering the fits with and without water termination.

To identify the crystalline orientation of the NbTiN film and its epitaxial relationship with the substrates, structural studies were performed by HRTEM along both [11–20] and [10–10] zone axis of the III-nitride substrates. Figure 2(a) show sample SAG viewed along the [11–20] zone axis of GaN. The superconducting film grows along the [111] axis, and it contains twin domains that are rotated by 60°. The images confirm the nominal crystallographic orientation of the substrates, and the establishment of NbTiN(111) [2–1–1] || (Al,Ga)N (0001) [10–10] and NbTiN(111) [2–1–1] || (Al,Ga)N(0001) [01–10] epitaxial relationships, as previously observed in the case of NbN layers [10, 18, 23]. The presence of dislocations at the interface is consistent with the relaxation of the misfit (36 GaN planes correspond to 37 NbTiN planes, as illustrated in figure 2(b)). In the case of SGF, observed in the same orientation, the thin film presents the same structural

characteristics, but a strong contrast is observed at the NbTiN/GaN:Fe interface (arrow in figure 2(c)). This contrast has been assigned to the segregation of Fe [29, 30], which is used as a dopant to render the GaN insulating. Finally, figure 2(d) depicts the NbTiN/AlN interface in sample SA, viewed along the [10–10] zone axis of AlN. The interface presents an epitaxial relationship as observed for NbTiN on GaN, but with lower density of dislocations, which appear separated by more than 100 AlN m planes. This is consistent with the NbTiN lattice being more closely matched to AlN than to GaN.

The variation of the resistance with temperature for the samples under study is presented in figure 3. The superconducting critical temperature moves in the range from $T_c = 10.6$ to 11.8 K, depending on the substrate, with a superconducting transition range (over which the resistance varies from 90% to 10% of its maximum value) $\Delta T = 0.83$ – 0.86 K. Precise values of T_c and ΔT for the samples under study are summarized in table 1, and they represent a significant improvement in comparison with our typical results of similar NbTiN layers deposited on dry thermal SiO₂ on Si, whose T_c is in the range of 9.8 ± 0.3 K with $\Delta T \approx 0.87$. A similar enhancement of T_c was observed by Shiino *et al* when inserting a sputtered AlN buffer layer between the NbTiN thin film and the quartz substrate [17]. This improvement is attributed to the good lattice matching between (111)-oriented rock-salt NbTiN and (0001)-oriented wurtzite AlN.

The superconducting properties of NbTiN might depend on its strain state [31]. In this sense, Makise *et al* showed an increase of T_c in NbTiN ultrathin films with a moderate tensile strain (out-of-the-plane lattice constant a_{\perp} around 4.35–4.39 Å in layers with a relatively high Ti content, around 30%) grown on various substrates. Krause *et al* report a degradation of the superconducting properties of NbN when increasing the Al content of AlGa_{1-x}N substrates from GaN towards AlN [17], which implies reducing the lattice mismatch with the substrate. However, the possible correlation of this trend with the strain in the layers was not analyzed. In our case, if we look at the in-plane lattice parameter of the substrate listed in table 1 (results extracted from XRD measurements of the (0002) reflection of AlN or GaN, assuming biaxial strain), we see that it evolves from a_{AlN} (SA) to a_{GaN} (SG, SGF), passing through an intermediate state with $a = 3.170$ Å due to the partial relaxation of the thin GaN layer on AlN for sample SAG. This slow relaxation is characteristic of III-nitride films grown by PAMBE under metal-rich conditions [32]. If we compare the evolution of T_c with the lattice parameter of the substrate (both listed in table 1), we observe an increase of T_c for decreasing substrate lattice parameter, i.e. a systematic improvement of the superconducting properties as we reduce the lattice mismatch between NbTiN and the substrate.

To get a precise assessment of the strain state of the superconducting layer, a detailed XRD study was conducted. The (111) inter-planar distance of the NbTiN layers along the [111] growth axis (d_{111}) was extracted from the angular position of the (111) and (222) reflections in ω - 2θ scans,

Table 1. Structural and electrical properties of the samples under study: nature of the substrate; substrate lattice parameter (a_{Subs}); thickness of NbTiN and total thin film thickness extracted from x-ray reflectivity (XRR), together with the root-mean-square (RMS) topmost surface roughness assumed for the fit, without and with water termination; out-of-the-plane and in-plane lattice parameters of NbTiN measured from XRD ($d_{[111]}$ and $d_{[2-1-1]}$, respectively) together with the cubic lattice parameters calculated from $d_{[111]}$ and $d_{[2-1-1]}$ assuming a regular cubic lattice (a_{\perp} and a_{\parallel} , respectively); average peak height and RMS roughness extracted from AFM measurements of the NbTiN samples and of the substrates; from resistance versus temperature measurements (R(T)), critical superconducting temperature (T_c) and temperature range of superconducting switching (ΔT) taken as the difference between the temperatures corresponding to a resistance drop of 10% and 90% from the maximum value.

Sample	SA	SAG	SG	SGF
Substrate	Sapphire/AlN	Sapphire/AlN/GaN	GaN	GaN:Fe
a_{Subs} (Å)	3.112	3.170	3.189	3.189
XRR NbTiN thickness (nm)	8.7 ± 0.1	8.3 ± 0.1	8.9 ± 0.6	7.9 ± 1.0
Total thickness (nm)	11.0 ± 1.4	10.9 ± 0.3	11.3 ± 0.5	10.8 ± 0.5
RMS roughness (nm)	3.5	1.5	0.8	2.0
RMS roughness with H ₂ O (nm)	0.5	0.6	1.1	1.8
XRD $d_{[111]}$ (Å)	2.538 ± 0.003	2.536 ± 0.003	2.538 ± 0.003	2.536 ± 0.004
$d_{[2-1-1]}$ (Å)	1.797 ± 0.010	1.791 ± 0.005	1.798 ± 0.004	1.791 ± 0.005
a_{\perp} (Å)	4.396 ± 0.006	4.393 ± 0.006	4.396 ± 0.006	4.393 ± 0.008
a_{\parallel} (Å)	4.402 ± 0.010	4.386 ± 0.008	4.403 ± 0.006	4.386 ± 0.008
AFM average peak height sample (nm)	1.9	2.3	3.8	4.7
RMS roughness sample (nm)	0.85	0.85	1.5	1.6
average peak height substrate (nm)	1.3	2.0	1.0	1.0
RMS roughness substrate (nm)	0.15	0.5	0.11	0.11
R(T) T_c (K)	11.8	11.3	11.0	10.6
ΔT (K)	0.83	0.85	0.86	0.85

whereas the inter-planar distance along the in-plane [2–1–1] axis ($d_{[2-1-1]}$) was extracted from the angular position of the (311), (133), (224), and (511) reflections. Average values of $d_{[111]}$ and $d_{[2-1-1]}$ are presented in table 1. From such values, we can calculate the equivalent out-of-the-plane and in-plane lattice parameters, $a_{\perp} = \sqrt{3} d_{[111]}$ and $a_{\parallel} = \sqrt{6} d_{[2-1-1]}$, respectively, also presented in the table. For all the samples, a_{\perp} and a_{\parallel} are identical, within the error bars of the measurements, which means that the samples are fully relaxed.

Figure 4 compares the surface morphology of the NbTiN layers under study, measured by AFM in the tapping mode. Typical AFM images shown in figures 4(a) and (b) (describing samples SA and SGF, respectively) show thin grainy layers. The surface roughness presents significant differences from sample to sample, as summarized in figure 4(c). Comparing samples SA and SGF, the average height increases from 1.9 to 4.7 nm, while the RMS roughness evolves from 0.85 to 1.6 nm, respectively. Note that these values are higher than the surface roughness of the substrates (shown in table 1), and there is no correlation between the NbTiN roughness and the substrate roughness. On the contrary, the variation of the NbTiN roughness from sample to sample correlates well with the lattice mismatch between the superconducting film and the substrate, which confirms that the strain relaxation plays a role in the surface roughness development. Let us remind that for low misfits and ultrathin layers, relaxation can be achieved by elastic undulation of the surface prior to the formation of dislocations [33].

From the structural analysis of the NbTiN layers we conclude that there is an enhancement of T_c in epitaxial layers grown on III-nitrides with respect to layers deposited on

amorphous materials, like SiO₂. A further improvement is observed as the substrate gets more closely lattice matched to the NbTiN epilayer, i.e. the T_c is higher for material deposited on AlN with respect to deposition on GaN. This trend cannot be related to the strain state, since the samples are fully relaxed. We therefore attribute it to the reduced roughness of the NbTiN film. This result might also explain the results observed by Krause *et al* when depositing NbN on various Al_xGa_{1-x}N, buffer layers [23], who showed a decrease of T_c for increasing Al mole fraction. In absence of studies of the strain state of the superconductor, an enhancement of the roughness in the ternary alloys could explain the degradation of the superconducting performance.

3.2. Fabrication and characterization of SNSPDs

In order to assess the suitability of such NbTiN films to the foreseen application, we have fabricated SNSPDs from sample SA, which displayed the best superconducting properties. The devices were designed for operation under normal incidence illumination. The NbTiN film was patterned into a meander structure as described in [8], using e-beam lithography and inductively-coupled plasma etching. The diameter of the detector area was 15 μm . The nanowire width was 70 nm, with a pitch of 150 nm, as illustrated by the scanning electron microscopy (SEM) image in the insert of figure 5(a).

We used a dip-stick system to immerse the detector in liquid helium. In our system, the detectors and the internal electronics are immersed within liquid helium, ensuring a stable operation temperature of 4.2 K. Continuous-wave laser diodes emitting at wavelengths $\lambda = 1.31$ and 0.65 μm were used for optical excitation. The laser power was attenuated in

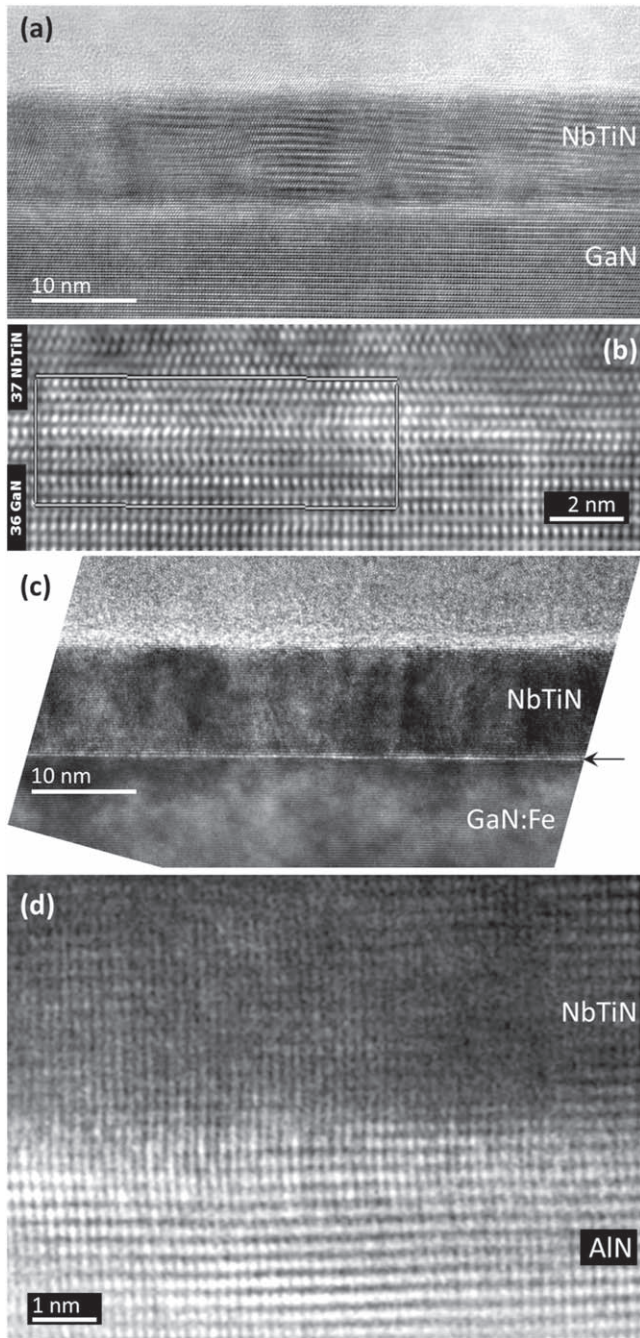


Figure 2. (a), (b) HRTEM images of the NbTiN layer and the NbTiN/GaN interface of sample SAG viewed along the [11–20] zone axis of GaN. (c) HRTEM image of SGF viewed along the [11–20] zone axis of GaN. The arrow marks the contrast at the NbTiN/GaN:Fe interface. (d) HRTEM image of the NbTiN/AlN interface in sample SA, viewed along the [10–10] zone axis of AlN.

order to obtain a photon flux around 10^6 photons s^{-1} . Figures 5(a) and (b) show the evolution of the detection efficiency at $\lambda = 1.31 \mu m$ and $0.65 \mu m$, respectively, as a function of the bias current/critical current ratio (I_b/I_{cr}). The dark count rate is superimposed on figure 5(a). As expected [34], the efficiency is maximum for transverse-electric (TE) polarization, where the electric field oscillates along the nanowire length, and drops down for transverse-magnetic

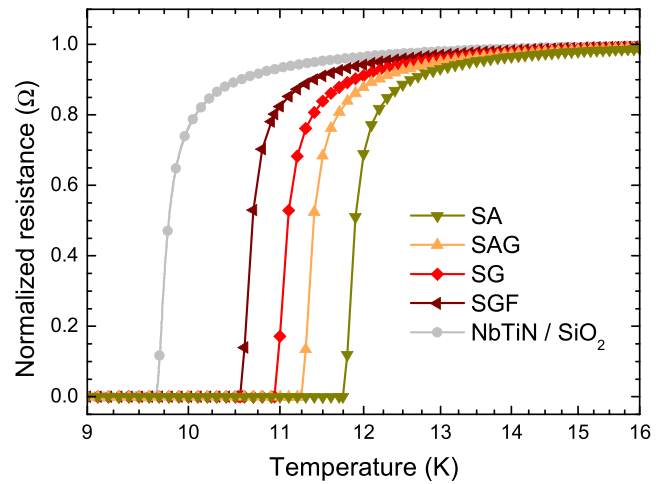


Figure 3. Variation of the resistance of the NbTiN layers under study as a function of temperature. The curve for an NbTiN deposited on dry thermal SiO₂ on Si under identical conditions was added for comparison. The resistance of the various samples was normalized to its maximum value.

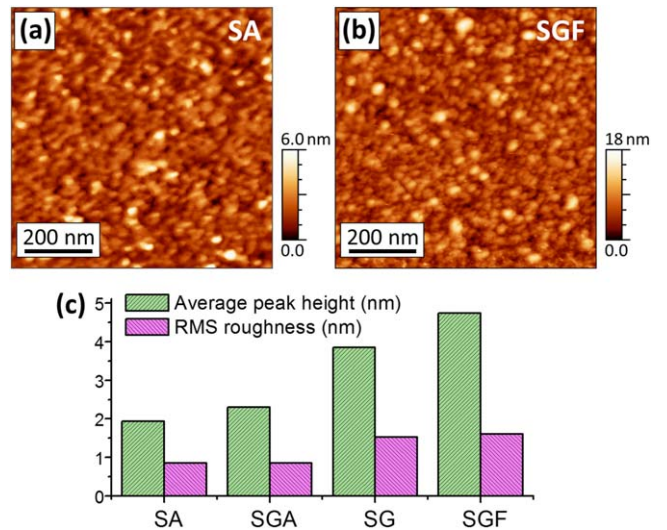


Figure 4. Atomic force microscopy images of samples (a) SA and (b) SGF. (c) Representation of the average height and RMS roughness extracted from $800 \times 800 \text{ nm}^2$ AFM images of samples SA, SAG, SG and SGF.

(TM) polarized light, where the electric field oscillates along the nanowire width. At $\lambda = 1.31 \mu m$, the highest efficiency recorded was 21% for TE polarization, but the response is not fully saturated. On the contrary, at $\lambda = 0.65 \mu m$, the higher photon energy results in full saturation for $I_b/I_{cr} \approx 0.8$ at 4.2 K.

Note that the devices present a very low level of dark counts, about ten times lower than typical measurements in SNSPDs on SiO₂/Si substrates. The low dark count rate is explained by the high critical current ($35 \pm 3 \mu A$, which corresponds to a critical current density in the meander of $\approx 5.9 \text{ MA cm}^{-2}$). Therefore, the voltage trigger level is well above the noise level of the readout system, which reduces the probability of dark counts. The enhancement of the critical current (and reduction of the noise level) is consistent with the

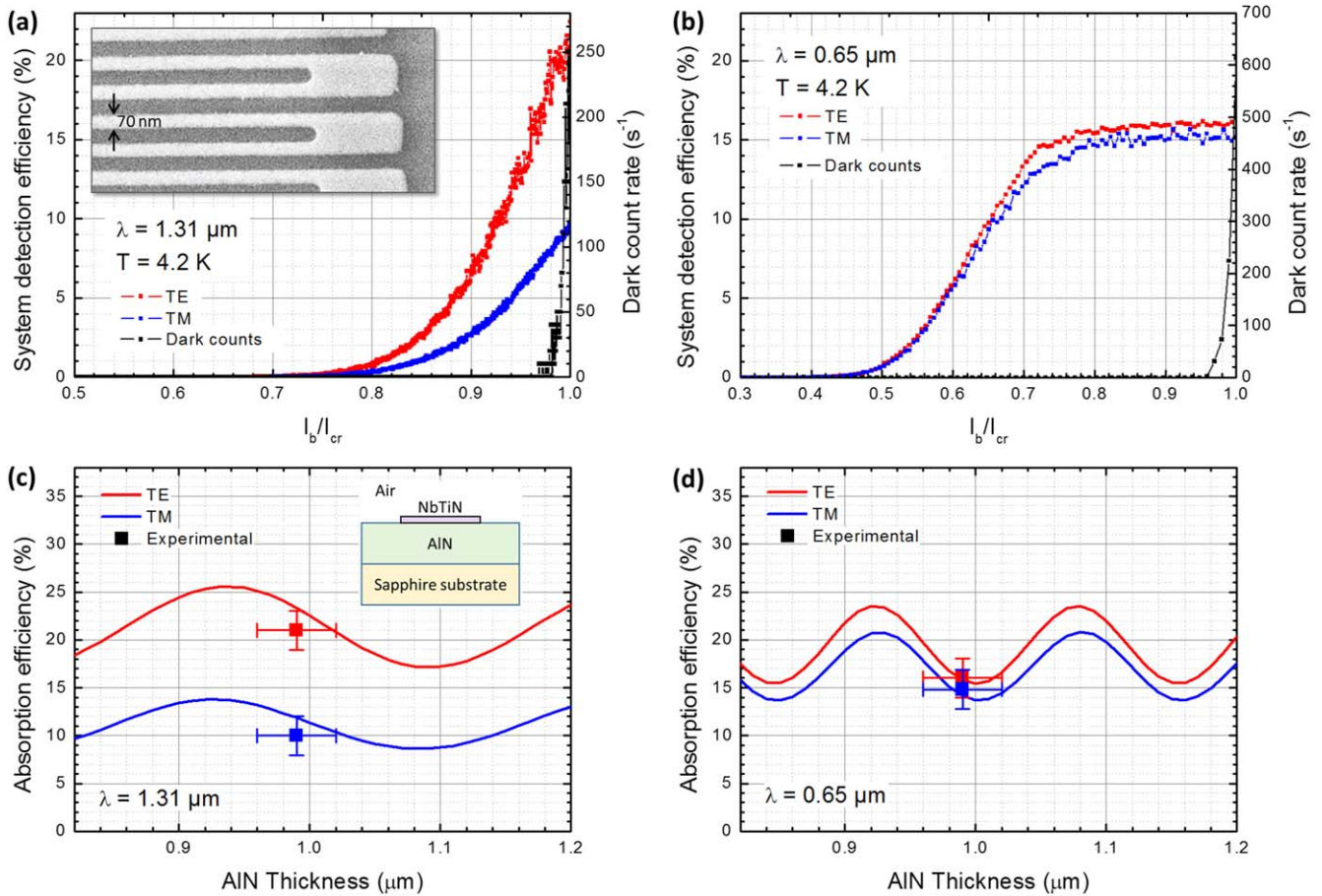


Figure 5. (a) Detection efficiency as a function of the bias current/critical current ratio (I_b/I_{cr}), measured for an SNSPD device fabricated on sample SA and measured at $1.31 \mu\text{m}$ and at 4.2 K (liquid He immersion). The dark count rate is also represented. Inset: SEM image of a detail of the superconducting nanowire meander. (b) Detection efficiency as a function of I_b/I_{cr} , measured at $0.65 \mu\text{m}$ and at 4.2 K . The dark count rate is also represented. (c) Calculated dependence of TE and TM absorption efficiency at $1.31 \mu\text{m}$ as a function of the thickness of the AlN layer. Experimental results are presented as squares with their associated error bars. Inset: scheme of the structure for which simulations have been performed. (d) Calculated dependence of TE and TM absorption efficiency at $0.65 \mu\text{m}$ as a function of the thickness of the AlN layer. Experimental results are presented as squares with their associated error bars.

improvement observed by Schmidt *et al* when inserting a sputtered AlN buffer layer before the deposition of NbN on GaAs [25].

The device performance is modeled using finite-difference time-domain (FDTD) calculations with the commercial RSoft FullWave software [26, 34]. The meander is modeled as a single grating period with in-plane boundary periodic conditions, as described in the inset of figure 5(c). Calculations assume an 8.5 nm -thick NbTiN active wire with a width of 70 nm and a pitch of 150 nm . The absorption efficiency is calculated from the difference between the input power and the reflected plus transmitted power, once the steady state is reached. The refractive indices used as input parameters were $4.17 + i5.63$ and $2.08 + i2.95$ for NbTiN at $1.31 \mu\text{m}$ and $0.65 \mu\text{m}$, respectively, 1.74 for sapphire, and 2.15 for AlN.

Figures 5(c) and (d) present the simulated absorption efficiency of the NbTiN wire at $\lambda = 1.31$ and $0.65 \mu\text{m}$, for TE and TM polarization, as a function of the thickness of the AlN template. The oscillatory patterns are due to Fabry–Perot interference in the AlN layer. The experimental values are represented with error bars that illustrate the uncertainty on

the thickness of the AlN layer and the dispersion of efficiency values for several devices. At $\lambda = 1.31 \mu\text{m}$, the measurements approach the theoretical results, in spite of the fact that the experimental efficiency curves are not fully saturated. On the contrary, at $\lambda = 0.65 \mu\text{m}$, full saturation is reached, leading to an excellent fit between the experimental results and the simulations, which implies that the internal quantum efficiency is close to one. Therefore, these experiments can be considered as a validation of the approach from material point of view. Then, an enhancement of the external efficiency can be easily obtained by modifying the nitride layer thickness to create a microcavity operating at the desired wavelength, or by implementing a GaN waveguide, if the aim is broadband operation (note that the SAG sample has already the adequate layer sequence for waveguide fabrication) [26].

4. Conclusion

In conclusion, we studied superconducting and structural properties of ultrathin layers of NbTiN sputtered on III-nitride

semiconductor substrates. The NbTiN films grow along the [111] axis, and they keep an epitaxial relation with the wurtzite substrate. The superconducting critical temperature is in the range of $T_c = 10.6\text{--}11.8$ K, depending on the substrate, with a sharp superconducting transition ($\Delta T = 0.83\text{--}0.86$ K). This represents an improvement with respect to layers deposited on amorphous substrates, like dry thermal SiO₂ ($T_c = 9.8 \pm 0.3$ K, $\Delta T \approx 0.87$). Higher T_c with smaller ΔT is obtained for deposition on AlN, whose crystalline lattice is better matched to NbTiN than that of GaN. The improvement with respect to GaN-based substrates is not related to the strain state, since all the NbTiN layers are fully relaxed. Instead, we attribute the higher T_c to a reduction of the NbTiN roughness when deposited on a lattice matched substrate. Based on these results, we have demonstrated the first superconducting nanowire single photon detectors on crystalline AlN, showing external quantum efficiencies that fit well with theoretical calculations.

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