Development of sputtered Nb$_3$Sn films on copper substrates for superconducting radiofrequency applications

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Received 5 October 2018, revised 16 November 2018
Accepted for publication 4 December 2018
Published 21 January 2019

Abstract

Superconducting radiofrequency (SRF) cavities that could provide a higher quality factor as well as a higher operational accelerating gradient at an affordable cost are in high demand for the future generation of particle accelerators. This study aims to demonstrate the potential of Nb$_3$Sn as material of choice for such SRF applications. Due to its brittle nature, the only way to produce an Nb$_3$Sn SFR cavity is to synthesise a thin layer inside a cavity made of niobium or copper. In this work, direct current magnetron sputtering using a stoichiometric target of Nb$_3$Sn was employed to produce films on copper samples. Assessment of the morphology, microstructure and superconducting properties were performed in order to ensure that this approach is suitable for SRF applications. The potential of the method is proven by obtaining films, which exhibit a crack-free surface, dense morphology and critical temperatures ($T_c$) up to 16 K. The essential properties of the films have also been investigated with respect to the deposition and annealing conditions. The use of krypton as working gas during deposition increases the atomic percent of Sn in the film compared to argon. However, in contrast to argon, higher krypton pressures reduce the atomic percent of Sn. It was also found that long-lasting high-temperature annealing leads to higher superconducting critical temperatures due to an increased crystallographic order. Particular attention was given to the influence of the copper substrate on the film growth as well as the microstructural and superconducting characteristics. We discuss the main constraints introduced by the copper substrate, such as copper interdiffusion during annealing, lattice mismatch and difference in thermal expansion coefficients and methods to overcome them.

Keywords: Nb$_3$Sn, magnetron sputtering, thin films, copper, cavity, superconducting radiofrequency, SRF

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

1. Introduction

The critical temperature twice as high as niobium and the high superheating field of Nb$_3$Sn motivate the superconducting radiofrequency (SRF) community to develop methods to apply this material to accelerating cavities [1]. Nb$_3$Sn cannot be used as a bulk material for cavity production due to its low thermal conductivity and brittle nature. Nevertheless, a thin film approach, in which a thin superconducting film is synthesised inside a cavity made out of an appropriate material, offers a promising route for exploiting Nb$_3$Sn for SRF use. Two possible candidate substrate materials are niobium and copper. The usual choice of niobium as a substrate for synthesising the Nb$_3$Sn layer is due to the wide spread use of this material for cavity production. The use of copper substrates,
instead, could bring several advantages for operation when considering the higher thermal conductivity of copper at cryogenic temperatures and the cost reduction granted to the lower base material price. Copper has indeed proven its efficiency to stabilise superconducting niobium thin films, preventing thermal runaways [2–5].

Considerable attention has previously been applied to Nb$_3$Sn films synthesis on niobium substrates. Starting from the 1970s, attempts to produce Nb cavities with a Nb$_3$Sn layer grown using Sn vapour-phase diffusion were performed [6]. Similar encouraging results continued from the late 1980s up to the 2000s with several 1.5 GHz Nb and multi-cell cavities being prepared and investigated [7–9]. The main result of this analysis was cavities which had Q-values exceeding those of bulk niobium at 4.5 K and low accelerating fields but which exhibited a strong Q-drop beyond a peak field of 15–30 MV m$^{-1}$. After several years of interruption, different groups resumed the study using modified vapour diffusion methods [10, 11]. Even making use of this new manufacturing method, problems associated with the Q-slope and early quenching remained unsolved. One possible cause for this behaviour was identified as the non-uniform thickness of the Nb$_3$Sn layer and the non-uniform Sn distribution over the cavity surface [12]. However, it should be noted that these Nb$_3$Sn preparation techniques are not applicable when a copper substrate is used, because of the processing temperature that is higher than the melting point of copper.

One of the interesting alternatives that can be used to synthesise Nb$_3$Sn on substrates other than niobium, is electrodeposition from SnCl$_2$ and NbCl$_5$ in 1-butyl-3-methylimidazolium chloride ionic liquids. This approach allows Nb$_3$Sn to be formed on a copper substrate in molecular form without high-temperature treatment [13]. A combination of electrodeposition of Sn with Cu intermediate layers and thermal treatment has also previously been used to deposit Nb$_3$Sn on niobium substrates [14]. The layer elaborated with such technique still needs to be tested under RF field especially to assess the potential effect of undesired crystalline phases, such as NbSn$_2$, observed by x-ray diffraction (XRD) analysis. In addition a recipe should be defined for the removal of the topmost bronze layer.

An alternative manufacturing approach is direct current magnetron sputtering (DCMS), which has been used to produce high-quality Nb$_3$Sn thin films on a range of substrates. Most of this development has been directed towards sapphire substrates, where the potential of this method has been revealed through films exhibiting a critical temperature from 15.3 to 18.3 K [15–18]. The principal advantages of the method are its simplicity and the possibility to produce homogenous layers of Nb$_3$Sn on complex shapes. However, the DCMS synthesis of Nb$_3$Sn films on a copper substrate brings additional challenges. For example, the difference in thermal expansion coefficients between the Nb$_3$Sn film and substrate contributes to disorder and thereby the critical temperature of the film, as shown by Kampwirth et al [17], is lowered by 5 K compared to the equivalent films grown on sapphire substrate.

DCMS has been extensively used to produce niobium thin film coated accelerating cavities for several accelerating facilities as LEP, LHC, HE ISOLDE at CERN, ALPI (INFN, LNL, Italy), and SOLEIL (St-Aubin, Gif-sur-Yvette, France) or ELETTRA and SLS for third harmonic bunch lengthening accelerators [2–5]. These previous successful transitions from test samples to real elliptical and quarter wave cavities justify the continued development of this approach for the production of Nb$_3$Sn films.

In the present work, DCMS has been employed to synthesise Nb$_3$Sn films on copper samples. Two different thermal treatment routes are proposed to produce these films, one in situ and the other which is applied after the coating procedure. We then investigate the effect of prolonged heat treatment on the superconducting critical temperature of the films. The effect of the copper substrate and the impact it has on thin film growth are also discussed. Finally, the substantial impact of the microcrystalline strain on the final superconducting performance of the film is demonstrated.

2. Sample preparation and characterisation

Thin films of Nb$_3$Sn with thicknesses ranging from 1.5 to 2 μm were grown on copper substrates using a planar DCMS system with a niobium–tin stoichiometric target. A schematic image of the sputtering setup is presented on figure 1.

Oxygen-free electronic grade copper substrates were chemically polished for 20 min in a mixture of sulfamic acid (H$_3$NO$_2$S, 5 g l$^{-1}$), hydrogen peroxide (H$_2$O$_2$, 5% vol), n-butanol (5% vol) and ammonium citrate (1 g l$^{-1}$) at 72 °C before being loaded into the deposition chamber. Ar and Kr were used as sputtering gases at coating pressures ranging from 1 × 10$^{-2}$ to 5 × 10$^{-2}$ mbar and from 3 × 10$^{-2}$ to 3 × 10$^{-2}$ mbar, respectively. A power applied during the coatings of 200 W, distance between sample holder and cathode of 100 mm and 1 h coating duration were used for all samples. The gas type and coating pressure were adjusted to control the composition of the Nb$_3$Sn films. This is of primary importance for A15 phase formation, as the production of an A15 lattice structure without the presence of pure niobium, tin or undesired phases of Nb–Sn, (such as Nb$_5$Sn$_2$ and NbSn$_2$) is only possible in the very narrow region of compositions ranging from ~19 to 26 at Sn% [19]. This lead to adopt a stoichiometric target.

To form the A15 superconducting phase, the films must undergo a high-temperature treatment; two different approaches were investigated. The first approach was based on annealing the films after the coating; it will be referred to as ‘reacted AFTER coating’. The annealing temperature applied to the film needs to be sufficiently high to facilitate A15 phase formation, however it cannot exceed the melting temperature of copper or damage the flanges which are brazed at 830 °C. Therefore, the investigated temperature range was selected to...
be from 600 °C to 800 °C, with durations of between 24 and 72 h. Annealing was performed in a vacuum furnace with a base pressure of \(7 \times 10^{-6}\) mbar during the treatment, a temperature uniformity of \(\pm 3^\circ\) C and a ramp rate of 300 °C h^-1. Following the annealing process, the samples were left inside the closed furnace to cool to room temperature.

The second investigated thermal processing route was the use of hot substrates during DCMS, which will be referred to as, ‘reacted DURING coating’. Films were coated using substrate temperatures between 600 °C and 735 °C and the same coating parameters as for ‘reacted AFTER coating’ method. Two thermocouples monitored the temperatures of the sample holder and resistive heater during the entire process. In some cases, samples synthesised using this approach were subsequently annealed either in the high-vacuum furnace described above or directly in situ by keeping the samples at a high temperature without venting to the air the coating system.

The composition of the films was determined using energy-dispersive x-ray spectroscopy (EDS). Measurements were carried out on each sample on five randomly selected sites within a 200 \(\mu m \times 200\) \(\mu m\) area using an accelerating voltage of 10 kV. The aperture has been adjusted to maximise the acquisition rate with minimum acquisition of \(10^6\) total counts per spectra for each sample. For some samples, x-ray photoelectron spectroscopy (XPS) was applied to obtain more detailed elemental quantification of the surface. Analysis was performed at normal emission angle on the surface of the samples and after removing by argon sputtering approximately 2.5 nm of the material using XPS setup with a XR50M M x-ray AlK\(\alpha\) monochromatized source and a PHOIBOS 150 hemispherical energy analyser equipped with nine channel-
3. Results and discussion

3.1. ‘Reacted AFTER coating’ process

As mentioned above, the primary composition adjustment parameter during the coating process was the process pressure. In figure 2 the dependence of the Sn atomic percent in the Nb₃Sn films as-deposited versus coating gas pressures using either Ar (closed circles) or Kr (open circles) is shown. Atomic Sn percentage for the samples reacted AFTER the coating varies from ~21% to ~27%, still in the suitable range according to the Nb–Sn phase diagram [19]. To control Sn distribution along cathode diameter of 150 mm, long copper stripes with length equal to the cathode diameter were coated at four different working pressures and analysed. Composition over the stripes length varies within 1–2 at%.

Deposition using Kr provides an increase of Sn at% up to ~3 at% compared to Ar. While using argon as the process gas, an increase of gas pressure rises atomic Sn percentage in the coated Nb₃Sn films, the use of Kr gas shows an opposite behaviour with a decrease in the atomic tin content in the film for increasing deposition pressures.

Post-coating annealings with various temperatures were applied to the deposited Nb₃Sn films to promote the formation of the superconducting A15 phase.

As it was observed, the annealing process can cause slight Sn evaporation for the films coated at deposition pressures up to 1 × 10⁻² mbar, but such deviation in the composition is not sufficient to interfere with superconducting A15 phase formation during annealing.

Nevertheless, Sn losses become significant for the samples coated at high coating pressures (~5 × 10⁻² mbar and higher), where the annealing process is causing nearly complete Sn evaporation during the heat treatment, and formation of superconducting A15 phase becomes impossible. These samples were excluded from further characterisation.

Witness bulk niobium pieces were annealed together with Nb₃Sn samples and then analysed using XPS analysis to confirm Sn evaporation. Traces of Sn were observed on the surface of the Nb samples after high-temperature treatment. To prevent this problem in future applications annealing could be performed in a closed configuration for cavities.

In figure 3, typical XRD diffractograms for the Nb₃Sn films deposited at an argon pressure of 1 × 10⁻³ mbar as-deposited (a) and after annealing for 24 h at 700 °C under vacuum (b) are shown.

XRD analysis of the samples before and after annealing confirms the formation of superconducting A15 phase only after high-temperature treatment. The expected diffraction pattern lines and their relative intensities are represented in red up-pointing triangles for Cu and blue circles for the Nb₃Sn phase. As-deposited Nb₃Sn films do not exhibit the superconducting A15 phase: the diffractogram (figure 3(a)) suggests a layer either amorphous or nanocrystalline according to the low intensity and broad peak (110) of Nb₃Sn (insert in figure 3(a)). Diffraction peak at 68° could suggest the presence of nanocrystalline Nb or Nb₃O that cannot be disentangled at this stage of the study. The presence of characteristic peaks of Nb₃Sn shown with blue circles (figure 3(b)) confirms the A15 phase formation after high-temperature treatment.

The crystallites’ size of Nb₃Sn films reacted after coating was estimated within the range of ~70–190 nm using Rietveld analysis, which is in good agreement with surface SEM observations.

The coating and annealing conditions strongly influence surface morphologies of the films. As an example, figure 4 displays the SEM surface images of typical Nb₃Sn films as-deposited (a), (c) and after annealing for 24 h at 800 °C (b), (d) coated under Ar (a), (b) and Kr (c), (d) atmospheres with a pressure of 1 × 10⁻³ mbar.

Both films synthesised under Ar and Kr gas pressures are highly disordered as-deposited, while annealing causes the formation of a grainy pattern. A difference in structural properties of the films coated with a different working gas can also be observed after heat treatment. In figure 5 one can see that samples coated under Ar pressure show cracks on the surface after the annealing step while films coated using Kr exhibit a crack-free surface.

Incorporated argon could contribute to the increase of residual stresses in the film and cause cracking during the annealing step. In films produced using the DCMS system some control of discharge gas trapping is needed to fine-tune the properties of the films [25]. The quantity of incorporated gas during deposition strongly depends on the relation between the atomic masses of the discharge process gas, and the cathode sputtered material [26]. The choice of deposition gas could influence the gas content in the produced samples.
As it was shown in [26], Kr content in the film is expected to be lower than Ar. This point is currently under investigation through the measurement of the residual gas content and residual stresses in the different coatings.

3.2. ‘Reacted DURING coating’ process

Nb₃Sn films reacted during the coating were considered as an alternative way to achieve a stable superconducting A15 phase in the samples. As in the case of samples ‘reacted AFTER coating’ the nature of the sputtering gas, as well as its
pressure during coating, were chosen as the mean to optimise the composition and morphology of the synthesised structures. In figure 6 the dependence of tin content versus coating pressure is presented: films coated using Ar as a working gas are shown with black circles, samples coated with Ar and kept at a high temperature for five additional hours are presented with black stars, and open symbols represent samples produced with Kr as sputtering gas.

The tin content in the samples produced using this route varies between ~19 and 25 at%. Most of the samples were synthesised using Ar as a deposition gas, and some trials were made using Kr.

XRD analysis confirms the presence of superconducting A15 phase for all samples reacted during coating at all investigated temperatures.

Typical XRD diffractogram for a sample coated at 710 °C under Ar pressure of 1 × 10⁻³ mbar is presented in figure 7. Nb₃Sn characteristic peaks confirm the formation of the superconducting cubic phase in the film. The crystallite sizes of Nb₃Sn films reacted during coating, calculated using the Rietveld method are significantly larger than for the films reacted after the coating and exceed the limit values of the method (~400 nm).

If the nature of the sputtering gas does not significantly affect the morphology of the films figure 8(a), its absolute pressure seems to have a substantial impact as presented in figure 8(b).

Nb₃Sn films deposited under low deposition pressures tend to produce films with dense, relatively smooth surface morphologies and at the same time exhibit a lower Sn content. On the contrary, high coating pressures increase the Sn content but lead to a rougher surface. During higher gas pressure coatings, sputtered material atoms suffer more collisions with the process gas atoms and lose substantial kinetic energy, resulting in a randomly oriented structure with more voids and rougher structure. We assume that a low coating pressure, close to 1 × 10⁻³ mbar, is the most appropriate one for SRF application as the final surface is smoother and thus less prone in providing field emitting features during cavity operation.

On top of the pressure, the coating temperature strongly affects the final film morphology as shown in figure 8(c). Bigger crystalline structures are obtained at a high temperature thanks to the enhanced adatoms’ surface mobility. Finally keeping the film at a high temperature for several hours once the coating process is ended allows to further modification of the film morphology. The latter results in a clear growth of the crystalline structures suggesting that the coating duration is not long enough to reach a stable phase and that additional time is needed to let the film reaching an equilibrium. More detailed analysis of the surface morphologies using TEM is foreseen in order to assess the nature of the black dots presented at SEM pictures in figure 8(a) which from our first investigation look like to be topological features.

3.3. Superconducting properties

The low temperature magnetic properties of the samples grown on copper substrate were investigated using VSM-SQUID.

A typical dependence of magnetic moment versus temperature is presented in figure 9.

Values of the critical temperature Tc onset were extracted from the transition point of the diamagnetic region with a negative moment (superconductivity) to the paramagnetic region with positive values of magnetic moment (normal conductivity region). We decided to focus on the onset value as it highlights the maximum reachable critical temperature for a given sample. Further optimisation will be carried out in order to minimise the transition width, which is most likely due to unreacted material, uneven composition distribution in the sample, contamination of the film or disorder in the crystalline lattice. It is to be noted that the transition broadness was observed to vary between 0.4 and 5 K with an average value of (1.7 ± 1.4) K.

The best obtained critical temperature for Nb₃Sn samples grown on copper substrate is around 16 K which remains 2 K below the expected theoretical bulk value of 18.3 K.
In figure 10(a) the dependence of the critical temperature onset on tin content in the films as-deposited is shown and compared with bulk Nb₃Sn (black triangles) [27]. Films coated using both routes reacted after and during coating are shown as blue squares and red circles, respectively.
As was discussed in many works, superconducting properties of the A15 material, i.e. critical temperature, critical current density, upper critical field are in correlation with the long-range crystallographic order (LRO) of the lattice [27–29]. Theoretically, bulk Nb3Sn with nominal stoichiometry at low temperature would have the highest possible crystallographic order [30]. In real structures, some disorder may remain in the lattice due to the finite time of cooling down from the melt or applied annealing [30]. In the case of thin films, this is complicated by additional disorder introduced both during deposition and during applied thermal treatment. Deviation from stoichiometry, which causes crystallographic disorder in the A15 phase, leads to lower critical temperature values as can be seen in figures 10(a), (b).

Some previous works on A15 materials have shown that crystallographic order might be destroyed by high-energy neutron radiation and almost completely recovered by annealing [31]. We observed a similar effect of the annealing on critical temperature for the Nb3Sn films reacted during coating. In figure 10(b), two samples marked with red diamonds were coated using the same deposition parameters. The use of post-coating annealing for 57 h to one of the samples allowed increasing its critical temperature by more than 3 K and achieving $T_c$ close to the critical temperature of bulk Nb3Sn with the same composition.

The effect of the high-temperature treatment duration can be considered by comparing films reacted after and during coating with the same composition (see dashed lines in figure 10(a) at Sn at% $\sim$22.9 and Sn at% $\sim$23.7). These points indicate that the critical temperatures for samples reacted during coating are lower compared to samples reacted after annealing. In the case of the films reacted during coating, substrates are heated only during the deposition time, i.e. $\sim$1 h, while post-coating annealing for the films reacted after coating takes $\sim$24 h. We believe that the rise of the annealing time increases the LRO degree and, hence, $T_c$.

### 3.4. Influence of microstrain on the superconducting properties

XRD is a robust method to perform quantitative characterisation of microstructural properties of the films. XRD measurements and details of the Rietveld refinement analysis are described above.

In this study, only samples coated under Ar pressure were taken into consideration. In figure 11 microstrain versus critical temperature dependence calculated for the samples reacted after annealing is shown.

All these films show cracks after annealing, thus uniform residual stress is released, but microstrain, a non-uniform strain that varies on a microscopic scale, remains significant.

This critical temperature sensitivity from microstrain confirms the importance of the microstructural properties for superconducting properties of the Nb3Sn films. In the case of films reacted during coatings no direct correlation could be found between the microstrain and the critical temperature. However it is very likely that in this case the residual stress effect surpasses the microstrain one. Further investigation are now needed to disentangle the effect of stresses and microstrain on the final reachable critical temperature.

### 3.5. Copper substrate influence

The use of copper as substrate requested a detailed study of its influence on the structural and superconducting properties of the sputtered Nb3Sn films. The impact of the copper substrate can be seen in figure 10 where Nb3Sn films grown on copper compared to the reference data of bulk Nb3Sn systematically exhibit lower critical temperature at given composition.

Nb3Sn films were grown both on copper and Al2O3 ceramic substrates using the same deposition parameters to separate the impact of magnetron sputtering film synthesis from that of the copper substrate in disorder degree.

The influence of the substrate choice on critical temperature values can be seen in table 1, which brings into comparison two pairs of films grown on copper and ceramic substrates using identical coating parameters, i.e. argon pressures of $1 \times 10^{-3}$ and $5 \times 10^{-3}$ mbar. After the coating, all samples were annealed at 750 °C for 24 h.

The impact of the copper substrate manifests itself in lower critical temperature values by $\sim$2–3 K.

On top of the critical temperature depression caused by copper substrate the difference in thermal expansion coefficient jointly with the brittleness of Nb3Sn lead to surface cracking after annealing in films reacted after coating. Undesired for RF applications, crack formation could be avoided by a careful optimisation of the deposition process.
parameters, including the substrate choice, and of the annealing parameters.

Another essential problem of the copper substrate associated with long-lasting annealing treatments is copper interdiffusion into the Nb₃Sn layer.

In figure 12(b) EDS mapping of copper on a cross section of Nb₃Sn film reacted after coating is presented. Boundaries of the film as well as copper-rich regions are highlighted.

Elemental mapping was performed on FIB cross section of a Nb₃Sn film. The analysis reveals nano-to-microscale copper inclusions towards the upper surface of the thin film.

In case of high-temperature depositions, copper interdiffusion into the superconducting layer manifest itself differently in films synthesised using Ar or Kr as working gas. Films prepared with Kr reveal copper-rich features (up to ~10 at%) observed on the surface of the produced Nb₃Sn films, as it can be seen in figure 13.

SEM-EDS analysis of the films coated with Ar, did not detect any copper inclusions inside Nb₃Sn layer as in case of reacted after coating films (figure 12) or copper-rich features on the surface as in case of the films grown under Kr pressure, which can be seen (figure 13). However, additional XPS analysis on the surface of the film detected traces of copper up to 7 at%.

For SRF applications, the presence of copper in the superconducting Nb₃Sn layer is undesired, as it creates local non-superconducting spots which would lead to power dissipation during the cavity operation and thus to poor performances. An appropriate intermediate layer between copper and Nb₃Sn could be a good response to the challenges caused by the copper substrate. Choosing a material with similar thermal expansion coefficient could increase the LRO degree, with consequent critical temperature increase, as well as reduction of residual stress in the superconducting film, which could prevent cracking of the films during annealing. Diffusion properties of the material together with the proper thickness of the intermediate layer could help to mitigate copper interdiffusion to the Nb₃Sn film. Several possible materials are currently under investigation.

4. Conclusion

We performed an extensive characterisation of Nb₃Sn thin films grown using a magnetron sputtering system. Thin films with critical temperatures up to 16 K are synthesised onto polycrystalline copper. We established the dependence of the deposition and annealing parameters on the superconducting and microstructural properties of the produced films. Deposition gas type and pressure affect the composition of the films with the latter ranging from 20% to 27% Sn content. High Ar pressure leads to a higher Sn atomic content in the structures, but at the same time causes rough and more porous morphology. The use of Kr as deposition gas also leads to an Sn content increase when compared to Ar, but an increase in Kr pressure decreases the Sn content in the film. Different high-temperature regimes are responsible for the superconducting and microstructural changes. Treatment with higher temperatures and durations increase the crystallographic order degree and hence, critical temperature. When compared to the ceramic substrate the effect of copper is demonstrated in decreasing the actual critical temperature of the film. Based on the present study, we believe that for the moment the most promising method for the future SRF application is high-temperature magnetron sputtering at low argon pressure, which gives dense films morphologies desired.
for SRF. To achieve composition close to stoichiometric, the pressure is chosen to be $5 \times 10^{-3}$ mbar. The lower critical temperature of the samples reacted during coating compared to films reacted after the sputtering can be compensated by the subsequent annealing at $\sim 750^\circ C$ for 24 h or longer. The use of an intermediate layer between copper and Nb$_3$Sn is envisaged to overcome problems due to the use of a copper substrate.

Acknowledgments

The research leading to these results has received funding from the European Commission under the FP7 Research Infrastructures project EuCARD-2, grant agreement no.312453. The research leading to this document is part of the Future Circular Collider Study. We thank the group of applied superconductivity in Geneva University, i.e. C Senatore and M Bonura, who performed the SQUID VSM measurements. The authors also thank the surface treatment team of CERN who performed all the samples chemical preparation prior to coating as well as H Neupert and D Zanin for the XPS analysis.

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References

[1] Belomestnykh S A and Padamsee H S 2015 Applied Superconductivity: Handbook on Devices and Applications ed P Seidel 1st edn (Weinheim: Wiley-VCH) (https://doi.org/10.1002/9783527670635)
[2] Calatroni S 2006 Physica C 441 95
[3] Chiaveri E, Weingarten W et al 1993 Industrial production of superconducting niobium sputter coated copper cavities for LEP Sixth Workshop on RF Superconductivity (Newport News, VA, USA) p 746SRF93f22
[4] Evans L 2007 New J. Phys. 9 335
[5] Sublet A, Aviles Santillana I, Bartova B, Calatroni S, Jecklin N, Mondino I, Therasse M, Venturini Delsolaro W, Zhang P and Cantoni M 2014 Nb coated HIE-ISOLDE QWR superconducting accelerating cavities: from process development to series production Proc. IPAC2014 (Dresden, Germany) p 2571
[6] Hillenbrand B, Martens H, Pfister H, Schnitzke K and Ziegler G 1975 IEEE Trans. Magn. 11 420
[7] Heinrichs H, Grundey T, Minatti N, Muller G, Peiniger M, Piel H, Unterborsch G and Vogel H P 1984 Proc. 2nd Workshop on RF Superconductivity (Geneva)
[8] Muller G et al 1997 Proc. 5th EPAC (Stiges, Barcelona, 10–14 June 1996) ed S Myers (Bristol: IOP Publishing)
[9] Peiniger M and Piel H 1985 IEEE Trans. Nucl. Sci. 32 3610–2
[10] Posen S and Hall D L 2017 Supercond. Sci. Technol. 30 033004
[11] Eremeev G, Clemens B, Macha K, Park H, Williams R S et al 2013 16th Int. Conf. RF Superconductivity (Paris, France) pp 603–6TUP071
[12] Trenkikina Y, Posen S, Romanenko A, Sardela M, Zuo J-M, Hall D L and Lipe M 2018 Supercond. Sci. Technol. 31 015004
[13] Franz S, Barzi E, Turrioni D, Giolonna L and Bestetti M 2015 Mater. Lett. 161 613–5
[14] Barzi E, Bestetti M, Reginato F, Turrioni D and Franz S 2016 Supercond. Sci. Technol. 29 015009
[15] Andreone A, Cassinese A, Di Chiara A, Iavarone M, Palomba F, Rosi A and Vaglio R 1997 J. Appl. Phys. 82 1736
[16] Wu C T, Kampwirth R T and Hafstrom J W 1977 J. Vac. Sci. Technol. 14 134
[17] Kampwirth R T, Hafstromand J W and Wu C T 1977 IEEE Trans. Magn. 13 315–8
[18] Agatsumo K, Tateishi H, Arai K, Saitoh T and Nakagawa M 1996 IEEE Trans. Magn. 32 2925–8
[19] Charlesworth J, MacPhail I and Madsen P 1970 J. Mater. Sci. 5 580
[20] Khereddine A, Larbi F H, Djebala L, Azzeddine H, Alili B and Bradai D 2011 Trans. Nonferr. Met. Soc. Chine 21 482
[21] Lutterotti L 2010 Total pattern fitting for the combined size-strain-stress-texture determination in thin film diffraction Nucl. Instrum. Methods Phys. Res. B 268 334–40
[22] Ozcan S, Can M M and Ceylan A 2010 Mater. Lett. 64 2447
[23] Sahu P and De M 2002 Mater. Sci. Eng. A 333 10–23
[24] Signorini L, Pasquini L, Savini L, Carboni R, Boscherini F, Bonetti E, Giglia A, Pedio M, Mahne N and Nannarone S 2003 Phys. Rev. B 68 195423
[25] Benvenuti C, Calatroni S, Campisi I E, Darriulat P, Peck M A, Russo R and Valente A-M 1999 Physica C 316 153
[26] Amorosi S, Anderle M, Benvenuti C, Calatroni S, Carver J, Chiggiato P, Neupert H and Vollenberg W 2001 Vacuum 60 89–94
[27] Godeke A 2006 Supercond. Sci. Technol. 19 R68–80
[28] Dew-Hughes D 1975 Cryogenics 15 435
[29] Hein R A 1973 The Science and Technology of Superconductivity (New York: Plenum) pp 333
[30] Cardwell D A and Ginley D S 2003 Handbook of Superconducting Materials vol 1 (Bristol: IOP Publishing) p 1075
[31] Sweedler A R, Schweitzer D G and Webb G W 1974 Phys. Rev. Lett. 33 168