Grain boundary characteristics of oxypnictide NdFeAs(O,F) superconductors

Kazumasa Iida¹,², Taito Omura², Takuya Matsumoto¹, Takafumi Hatano¹,² and Hiroshi Ikuta¹,²

¹Department of Materials Physics, Nagoya University, Furo-cho, Chikusa-ku, Nagoya 464-8603, Japan
²Department of Crystalline Materials Science, Nagoya University, Furo-cho, Chikusa-ku, Nagoya 464-8603, Japan

E-mail: iida@mp.pse.nagoya-u.ac.jp

Received 28 December 2018, revised 31 March 2019
Accepted for publication 5 April 2019
Published 23 May 2019

Abstract

We have systematically investigated the grain boundary angle (θgb) dependence of transport properties for NdFeAs(O,F) fabricated on [001]-tilt symmetric MgO bicrystal substrates. In our previous study, NdFeAs(O,F) bicrystal films showed a weak-link behaviour even at a θgb of 6°. However, this was caused by an extrinsic effect originating from the damage to both NdFeAs(O,F) and MgO substrate by excess F-diffusion along the grain boundary. To investigate the intrinsic nature of grain boundaries, we minimized the damage to NdFeAs(O,F) and MgO by reducing the deposition temperature of the NdOF over-layer needed for F-doping. The resultant NdFeAs(O,F) bicrystal films have a critical angle of 8.5°, above which Jc starts to decrease exponentially. This critical angle is almost the same as those of other Fe-based superconductors.

Supplementary material for this article is available online

Keywords: Fe-based superconductors, thin films, grain boundary, critical angle

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

1. Introduction

The discovery of iron-based superconductors (FBS) triggered a huge excitement for the community of fundamental and applied superconductivity research. Since then many physical quantities of FBS have been revealed, e.g. the coherence length is short, which results from the small Fermi velocity and low carrier density [1]. The in-plane coherence length ξ∥(0) at zero kelvin obtained from upper critical field measurements is 2–4 nm for FBS (e.g. the respective ξ∥(0) for FeSe(Se,Te), BaFe2(As0.7Sb0.3)2, (Ba,K)Fe2As2, NdFeAsO0.8F0.2, and SmFeAs(O,F) were 1.5 nm [2], 2.1 nm [3], 1.6 nm [4], 3.9 nm [5], and 1.4 nm [6]), which is similar to that for the cuprates. This short coherence length arouses concern that the superconductivity of FBS at grain boundaries may be depressed by crystalline disorder. However, the symmetry of the superconducting order parameter of FBS is reported to be an extended s-wave [7–9], which differs from the d-wave of the cuprates [10]. Hence, the critical current density (Jc) of FBS across the GB is expected to be not as severely reduced as cuprates. In fact, the critical angle (θc), above which Jc starts to decrease exponentially, is around 9° for Co-doped BaFe2As2 (Ba-122) [11] and Fe(Se,Te) [12, 13], much larger than YBa2Cu3Ox (θc = 3°–5°) [14]. Additionally, the grain boundaries for Co- and P-doped Ba-122 [11, 15], and Fe(Se, Te) [12, 13] are of metallic nature. Both of these features are favourable for high-field conductor applications.

Among the various FBS LuFeAs(O,F) (Lu: lanthanoid) has the highest superconducting transition temperature (Tc), which yields a large margin between Tc and the temperature where a cryocooler can reach. Hence, SmFeAs(O,F) wires [16] and NdFeAs(O,F) tapes [17] have been demonstrated as proof-of-principle studies for conductor applications. However, a large Jc gap between those conductors and the counterpart of single crystals or epitaxial thin films was recognised [18]. To understand the reason for such performance gap, GB characteristics of LuFeAs(O,F) should be clarified.

Previously, we have investigated the grain boundary angle (θgb) dependence of inter-grain Jc (Jc(θ)) for NdFeAs(O,F) grown on [001]-tilt symmetric MgO bicrystal substrates.
As a result, \( J_{c}^{\text{GB}} \) for \( \theta_{\text{GB}} = 6^\circ \) was reduced by nearly 30% compared to the intra-grain \( J_{c} (J_{c}^{\text{Grain}}) \). Microstructural investigation by transmission electron microscope (TEM) revealed that fluorine preferentially diffused through GB and thereby eroded NdFeAs(O,F) and the substrate. Hence, suppressing the excess diffusion of F is the key to exploring the intrinsic grain boundary properties. In the aforementioned study, F-doping has been conducted by a NdOF over-layer method [20], where parent NdFeAsO was deposited at 800°C, followed by the deposition of NdOF at the same temperature. Here, a growth temperature of 800°C was necessary for achieving high crystallinity of NdFeAsO [21]. On the other hand, the deposition temperature (\( T_{\text{dep}} \)) of NdOF would not affect the crystalline quality of NdFeAs(O,F). Hence, it is possible to suppress the excess F diffusion by lowering \( T_{\text{dep}} \) of NdOF. However, the lack of F leads to low-\( T_{c} \) or even non-superconducting NdFeAs(O,F) films. Therefore, NdFeAs(O,F) bicrystal films should be fabricated by employing a \( T_{\text{dep}} \) of NdOF as low as possible without compromising the superconducting properties.

In this paper, we firstly determine the lowest \( T_{\text{dep}} \) for the NdOF over-layer that results into a NdFeAs(O,F) film with a sufficiently high \( T_{c} \) by mapping the diagram of \( T_{c} \) versus \( T_{\text{dep}} \). Then NdFeAs(O,F) bicrystal films are fabricated by employing the thus determined \( T_{\text{dep}} \) for NdOF, followed by structural characterisations as well as transport measurements.

2. Experiment

NdFeAs(O,F) thin films were grown on MgO(001) single crystalline substrates by molecular beam epitaxy (MBE) using solid sources of Fe, Fe2O3, As, NdF3 and Ga. Here, Ga was used as a F-getter to adjust the amount of fluorine [22]. A two-step process was employed to grow superconducting NdFeAs(O,F) thin films: parent NdFeAsO films of 40 nm thickness were grown at 800°C, followed by the deposition of 12 nm thick NdOF in the temperature range 600°C \( \leq \ T_{\text{dep}} \leq \ 800°C \). For identifying the lowest deposition temperature of NdOF, \( T_{\text{c}} \) of the resultant NdFeAs(O,F) films was plotted as a function of \( T_{\text{dep}} \) (see figure 2). After determination of the lowest \( T_{\text{dep}} \) of NdOF, NdFeAs(O,F) thin films were fabricated on [001]-tilt symmetric MgO bicrystal substrates (\( \theta_{\text{GB}} = 6^\circ, 9^\circ, 12^\circ, 24^\circ \)) by the two-step process, whereas the respective thicknesses of NdFeAsO and NdOF were 160 nm and 50 nm to maintain the thickness ratio of NdOF to NdFeAsO as around 0.3.

Phase purity and c-axis texture were examined by x-ray diffraction (XRD) in Bragg-Brentano geometry using Cu-Kα radiation. In-plane orientation of NdFeAs(O,F) was investigated by the \( \phi \)-scan of the 102 peak. After structural characterisation by XRD, the NdOF over-layer was removed by Ar-ion milling. Microstructural analyses on the NdFeAs(O,F) bicrystal film with \( \theta_{\text{GB}} = 6^\circ \) were performed by TEM. From the TEM image of cross-sectional view away from the GB region, the thickness of NdFeAs(O,F) was found to be around 130 nm after Ar-ion milling.

For transport measurements of both inter- and intra-grain \( J_{c} \), two micro-bridges were fabricated on the same film. The thin films were patterned by a photolithography method and then etched by Ar-ion milling to form micro-bridges. The bridges were 20-40 μm wide and 1 mm long across the grain boundary and 0.25 mm long within the grain away from the grain boundary. Transport properties using the resultant bridges were measured by a four-probe method. The onset \( T_{c} \) was determined as the intersection between the fit to the normal state resistivity and the steepest slope of resistivity.

An electric field criterion of \( E = 1 \ \mu \text{V cm}^{-1} \) was applied for evaluating intra-grain \( J_{c} (J_{c}^{\text{Grain}}) \) and inter-grain \( J_{c} (J_{c}^{\text{GB}}) \) for \( \theta_{\text{GB}} = 6^\circ \) and \( 9^\circ \). \( J_{c}^{\text{GB}} \) for \( \theta_{\text{GB}} = 12^\circ \) was defined as the intersection between \( E = 0 \) and a linear fit to the non-ohmic linear differential (NOLD) region (i.e. the region where \( E \) increased linearly with the current density \( J \)).

3. Results and discussion

Figure 1(a) shows the XRD patterns of the NdFeAs(O,F) thin films for various deposition temperatures of the NdOF over-layer. All NdFeAs(O,F) films were grown c-axis oriented. The diffraction peak of NdAs, which is a by-product of chemical reaction between NdOF and NdFeAsO, appeared above \( T_{\text{dep}} = 700°C \). This is a natural consequence that higher deposition temperatures stimulate the chemical reaction. Another feature is a shift of the 00l reflections to higher angles with increasing \( T_{\text{dep}} \), indicative of shortening the c-axis length (figure 1(b)). It was reported that the c-axis lattice parameter of LaFeAsO1-xFx [23] and CeFeAsO1-xFx [24] monotonously decreased with increasing F content. Hence, the observed change indicates that F diffuses into NdFeAsO and the amount of F in NdFeAs(O,F) increased with increasing the deposition temperature of the NdOF over-layer.

The temperature dependence of the normalised resistivity \( (\rho/\rho_{300K}) \) varies with \( T_{\text{dep}} \) (figure 1(c)). Below \( T_{\text{dep}} = 650°C \), the normalised resistivity curves of the films were almost similar to those of the parent NdFeAsO. The upturn of \( \rho/\rho_{300K} \) at around 150 K corresponds to the structural transition from a high temperature tetragonal phase to a low temperature orthorhombic one. An onset \( T_{c} \) of 38 K was recorded for the film with \( T_{\text{dep}} = 700°C \). Further increase of the \( T_{\text{dep}} \) leads to an improvement of \( T_{c} \).

Figure 2 summarises the c-axis lattice parameter and \( T_{c} \) for the NdFeAs(O,F) films as a function of deposition temperature of NdOF. The c-axis lattice parameter decreased almost linearly with increasing \( T_{\text{dep}} \). Superconductivity can be induced by the deposition of NdOF at \( T_{\text{dep}} > 650°C \). From these results, the lowest \( T_{\text{dep}} \) with keeping superconductivity is determined to be 700°C. In the following results, the NdOF over-layer was deposited at 700°C.

Structural characterisation of the NdFeAs(O,F) films grown on [001]-tilt symmetric MgO bicrystal substrates is summarised in figure 3. The \( \theta/2\theta \)-scans showed that the NdFeAs(O,F) films were c-axis oriented. The azimuthal \( \phi \)-scan of the off-axis 102 reflection of NdFeAs(O,F) showed
clearly eight peaks from two adjacent NdFeAs(O,F) grains separated by \( \theta_{\text{GB}} \) (figure 3(b)). Hence, the NdFeAs(O,F) films were grown epitaxially on [001]-tilt symmetric MgO bicrystal substrates.

In figure 4(a) the resistivity curves for the inter-grain bridges (bridge width \( w = 30 \, \mu\text{m} \)) with various misorientation angles are shown. For comparison the data of the intra-grain bridge (\( \theta_{\text{GB}} = 0^\circ \)) fabricated from the NdFeAs(O,F) film with \( \theta_{\text{GB}} = 6^\circ \) are also plotted. As can be seen, the superconducting transition temperature did not systematically change with the grain boundary angle, which differs from our previous study (i.e. \( T_c \) decreased with increasing \( \theta_{\text{GB}} \)) [19]. The inter-grain bridges with \( \theta_{\text{GB}} = 6^\circ \) and \( 12^\circ \) had an onset \( T_c \) (\( T_{\text{onset}} \)) of 45 K, which is almost the same as for the intra-grain bridges. For \( \theta_{\text{GB}} \geq 9^\circ \) a wide transition width of 5 K was observed. As shown in supplementary figure S1 available online at stacks.iop.org/SUST/32/074003/mmedia, almost no difference in \( T_c \) as well as the superconducting transition width were found with changing the bridge width.

A finite resistivity was observed for the inter-grain bridge with \( \theta_{\text{GB}} = 24^\circ \) below \( T_c \), whereas the other bridges showed a resistivity well below the instrumental limitation. Hence, the grain boundary region for \( \theta_{\text{GB}} = 24^\circ \) was destroyed by F even at \( T_{\text{dep}} = 700^\circ \text{C} \), and accordingly, \( J_c \) across the GB is zero.

Figure 4(b) compares the \( E-J \) curves for the inter-grain bridges with various \( \theta_{\text{GB}} \) measured at 4.2 K without a magnetic field. The curve for the intra-grain bridge (\( \theta_{\text{GB}} = 0^\circ \)) is also included, which exhibited a power-law behaviour described by \( E \sim J^a \), representing flux creep effects. The same behaviour was also seen for the inter-grain bridges with \( \theta_{\text{GB}} = 6^\circ \) and \( 9^\circ \), which indicates that the two adjacent grains were strongly coupled. On the other hand, the inter-grain curve for \( \theta_{\text{GB}} = 12^\circ \) showed a NOLD behaviour (i.e. \( E \) is linearly changing with \( J \)) originating from viscous flux flow along the grain boundaries [25], indicative of \( J_c \) limitation by grain boundaries [26]. The resistance area product for \( \theta_{\text{GB}} = 12^\circ \) was estimated to be \( 8.5 \times 10^{-11} \, \Omega \, \text{cm}^2 \), which is of the same order of magnitude for Co-doped Ba-122 with \( \theta_{\text{GB}} = 45^\circ \) (\( 5 \times 10^{-10} \, \Omega \, \text{cm}^2 \)) [11], P-doped Ba-122 with \( \theta_{\text{GB}} = 24^\circ \) (\( 1.3 \times 10^{-10} \, \Omega \, \text{cm}^2 \)) [15], and Fe(Se,Te)
with $\theta_{\text{GB}} = 24^\circ$ ($7 \times 10^{-10} \Omega \text{cm}^2$) [12]. Hence, the grain boundary of NdFeAs(O,F) is of metallic nature, which is similar to other FBS.

The field dependence of $J_c$ normalised to the self-field $J_c (J_{c,\text{sf}})$ plotted in figure 4(c) also shows a typical behaviour of GB transport observed for FBS. For the bridge with $\theta_{\text{GB}} = 12^\circ$, inter-grain $J_c$ was decreased by 70% even at a small applied magnetic field of 0.5 T, whereas the corresponding reduction for $\theta_{\text{GB}} = 0^\circ$ (i.e. intra-grain) and $\theta_{\text{GB}} = 9^\circ$ were 23% and 36%, respectively. Such a sharp drop of inter-grain $J_c$ for large $\theta_{\text{GB}}$ by small magnetic fields is due to the weak coupling nature between adjacent grains.

To further characterise the coupling between the grains with $\theta_{\text{GB}} = 6^\circ$, microstructural observation by TEM was conducted (figure 5(a)). For comparison, the microstructures away from the GB region are also shown in figure 5(b). The elemental mappings revealed a uniform distribution of all elements except for O and F (figure 5(b)). The observed high concentration of O and F at the surface region may be caused by heating due to Ar-ion milling to remove the NdOF over-layer. Uniform distribution of Nd, Fe, As, O and F within NdFeAs(O,F) was also observed in a different bridge (see supplementary figure S2).

Figure 3. Summary of the structural characterisation for the NdFeAs(O,F) thin films fabricated on [001]-tilt symmetric MgO bicrystal substrates having various $\theta_{\text{GB}}$. (a) The $\theta$/2$\theta$-scans and (b) the azimuthal $\phi$-scans of the off-axis 102 reflection of NdFeAs(O,F).

Figure 4. (a) Semi-logarithmic plot of the resistivity curves for the inter-grain bridges having various $\theta_{\text{GB}}$ and the intra-grain bridge ($\theta_{\text{GB}} = 0^\circ$). The intra-grain bridge was fabricated from the NdFeAs(O,F) film with $\theta_{\text{GB}} = 6^\circ$. The measurements were conducted using 30 $\mu$m wide bridges. (b) $E$-$J$ curves of the inter-grain bridges of 30 $\mu$m wide for various $\theta_{\text{GB}}$ and the intra-grain bridge ($\theta_{\text{GB}} = 0^\circ$) measured at 4.2 K without magnetic field. The intra-grain bridge was fabricated from the NdFeAs(O,F) film with $\theta_{\text{GB}} = 12^\circ$. (c) Comparative field dependence of $J_c$ normalised to the self-field $J_c (J_{c,\text{sf}})$ measured at 12 K for $H||c$. The bridges are the same as those presented in (b).
was strongly inhomogeneous due to the damage to NdFeAs (O,F) by F even though the deposition temperature of the NdOF over-layer was lowered (figure 5(a)). This is very similar to what was observed in our previous investigation [19]. However, the bottom region (about half of the thickness from the substrate) was quite uniform and free from damage by F. Hence, the grains are still coupled strongly in this region.

The temperature dependence of self-field \( J_c \) for various misorientation angles are summarised in figures 6(a)–(c). The intra-grain \( J_c^{\text{Grain}} \) of each film is also shown for comparison. Here, the measurements were conducted using 30 \( \mu \)m wide bridges. The results obtained from various bridges are summarised in supplementary figure S3. The self-field \( J_c \) of the intra-grain at low temperatures is in the order of \( 10^5 \) A cm\(^{-2} \) for all three samples, which is one order of magnitude lower than our NdFeAs(O,F) films grown by employing \( T_{\text{dep}} = 700 \) °C and 800 °C. 

\[ J_c (A/cm^2) \]

\[ T (K) \]

Figure 5. (a) TEM micrograph of a cross section of the GB region having \( \theta_{\text{GB}} = 6^\circ \) and the elemental mappings at the same area. The location of the GB is indicated by a dashed yellow line in the micrograph. (b) The microstructures away from the GB region. The thickness of NdFeAs(O,F) was around 130 nm.

Figure 6. (a)–(c) Temperature dependence of the self-field \( J_c \) for the intra- and inter-grain bridges (bridge width \( w = 30 \mu \)m) having various \( \theta_{\text{GB}} \). (d) Self-field \( J_c^{\text{Grain}} \) as a function of reduced temperature for NdFeAs(O,F) films grown by employing \( T_{\text{dep}} = 700 \) °C and 800 °C.
than that of a film with $T_{dep} = 800 \, ^\circ$C (figure 6(d)). This is due to low carrier concentration. Hall effect measurements on our NdFeAs(O,F) films grown on single crystalline MgO revealed that the carrier concentration decreased from $1.98 \times 10^{21} \text{cm}^{-3}$ for $T_{dep} = 800 \, ^\circ$C to $1.34 \times 10^{21} \text{cm}^{-3}$ for $T_{dep} = 750 \, ^\circ$C at 50 K [28]. Although a precise evaluation of the carrier concentration is not easy due to the multi-band nature of the current superconductor. The carrier concentration of NdFeAs(O,F) for $T_{dep} = 700 \, ^\circ$C is expected to be even smaller than that for $T_{dep} = 750 \, ^\circ$C, which probably explains the smaller $J_c$.

The distinct feature of the above results is that the respective inter-grain critical current densities ($J_{cGB}$) for $\theta_{GB} = 6^\circ$ and $9^\circ$ are almost comparable to their intra-grain $J_c$ ($J_{cGrain}$) at all temperatures (figures 6(a) and (b)), indicating the absence of weak-links for both bicrystal films. For $\theta_{GB} = 12^\circ$, $J_{cGB}$ at 4.2 K decreased by only 28% compared to $J_{cGrain}$. For $\theta_{GB} = 12^\circ$, $J_{cGB}$ below 32 K was certainly lower than $J_{cGrain}$. In Cu–Ag alloys it has been observed that the GBs are completely wetted by Ag when the misorientation angle is large [29]. It might be possible that F-rich phase had covered GBs and caused the transition from strong-link to weak-link behaviour. Nevertheless, $J_{cGB}$ at 4.2 K decreased by only 28% compared to $J_{cGrain}$ in stark contrast to our previous study [19].

Note that the effective cross-sectional area in the vicinity of the GB with $\theta_{GB} = 6^\circ$ was smaller than the intra-grain bridge as a consequence of the damage caused by F diffusion (see figure 5(a)). This is probably the reason why the plotted $J_{cGB}$ values are higher than $J_{cGrain}$ in figure 6(a), since $J_c$ was calculated by assuming that both the inter- and intra-grain bridges have the identical thickness.

Figure 7 shows the $\theta_{GB}$ dependence of the ratio of inter-to intra-grain $J_c$ ($J_{cGB}/J_{cGrain}$) for the NdFeAs(O,F) bicrystal films at 4.2 K. For comparison the data of Co-doped Ba-122 at 4 K [11], Fe(Se,Te) at 4.2 K [12, 13], and our previous work on NdFeAs(O,F) at 4.2 K [19] are also plotted. The present work on NdFeAs(O,F) bicrystal films showed that $J_{cGB}/J_{cGrain}$ was unity up to 8.5° (i.e. $\theta_c = 8.5^\circ$), followed by a decrease with $\theta_{GB}$. On the other hand, $\theta_c$ was less than 6° in our previous investigation. These results indicate that the grain boundary properties were improved by lowering $T_{dep}$ of NdOF. It is interesting that all FBS have a similar critical angle of around 9°, suggesting that the symmetry of the superconducting order parameter for NdFeAs(O,F) may be the same as Co-doped Ba-122 and Fe(Se,Te).

$\theta_c = 8.5^\circ$ was determined for NdFeAs(O,F), it is still ambiguous how the inter-grain $J_c$ decreases with $\theta_{GB}$ beyond 12°. Additionally, it is not ruled out completely that the damage to the GB region by F had influenced the obtained results. To answer these issues and explore further the intrinsic GB characteristics, studying F-free oxypnictides, e.g. Ln(Fe$_{1-x}$Co$_x$)AsO [30, 31], would be interesting. Albeit the highest $T_c$ is around 25K for Nd(Fe$_{1-x}$Co$_x$)AsO [31], there would be no damage to the GB region by F. The misorientation angle dependence of $J_{cGB}$ for Nd(Fe$_{1-x}$Co$_x$)AsO bicrystal films would be the direction of our future studies.

4. Summary

NdFeAs(O,F) epitaxial thin films were fabricated by MBE using a two-step process, where the deposition temperature ($T_{dep}$) of NdOF was varied in the temperature range $600 \, ^\circ$C $\leq T_{dep} \leq 800 \, ^\circ$C. The c-axis lattice parameter and $T_c$ of NdFeAs(O,F) changed systematically with $T_{dep}$ of NdOF, and the F content in NdFeAs(O,F) increased with increasing $T_{dep}$. The lowest $T_{dep}$ of NdOF without compromising $T_c$ was determined as 700°C. By using $T_{dep} = 700 \, ^\circ$C epitaxial NdFeAs(O,F) thin films were grown on [001]-tilt symmetric MgO bicrystal substrates with $\theta_{GB} = 6^\circ$, $9^\circ$, $12^\circ$ and $24^\circ$. Even for this reduced $T_{dep}$, however, the inter-grain bridge with $\theta_{GB} = 24^\circ$ showed a finite resistance and hence $J_c$ was zero due to the erosion of NdFeAs(O,F) by F. Additionally, $J_c$ of the inter-grain bridges with $\theta_{GB} \leq \theta_c$ and the intra-grain bridge were around $10^7 \, \text{A cm}^{-2}$ at low temperatures, which is almost one order of magnitude lower than NdFeAs(O,F) with $T_{dep} = 800 \, ^\circ$C due probably to low carrier concentration. Nevertheless, the GB properties of the inter-grain bridges with lower $\theta_{GB}$ were improved, resulting in a critical angle of $\theta_c = 8.5^\circ$. This critical angle is identical to Co-doped Ba-122 and Fe(Se,Te).

Acknowledgments

This work was supported by the JSPS Grant-in-Aid for Scientific Research (B) Grant Number 16H04646 as well as JST CREST Grant Number JPMJCR18J4.
ORCID iDs

Kazumasa Iida https://orcid.org/0000-0003-1038-9630

References

[1] Putti M et al 2010 Supercond. Sci. Technol. 23 034003
[2] Klein T et al 2010 Phys. Rev. B 82 184506
[3] Chaparro C, Fang L, Rydh A, Crabtree G W, Stantev V, Kwok W K and Welp U 2012 Phys. Rev. B 85 184525
[4] Kacmarcik J, Marcenat C, Klein T, Pribulova Z, van der Beek C J, Konczykowski M, Budko S L, Tillman M, Ni N and Canfield P C 2019 Phys. Rev. B 85 184525
[5] Adamski A, Krellner C and Abdel-Haq M 2017 Phys. Rev. B 96 100503
[6] Welp U, Chaparro C, Koshelev A E, Kwok W K, Rydh A, Zhigadlo N D, Karpinski J and Weyeneth S 2011 Phys. Rev. B 83 100513
[7] Mazin I L, Singh D J, Johannes M D and Du M H 2008 Phys. Rev. Lett. 101 057003
[8] Kuroki K, Onari S, Arita R, Usui H, Tanaka Y, Kontani H and Aoki H 2009 Phys. Rev. Lett. 101 087004
[9] Hanaguri T, Niitaka S, Kuroki K and Takagi H 2010 Science 328 474
[10] Tsuei C C and Kirtley J R 2000 Rev. Mod. Phys. 72 969
[11] Katsane T, Ishimaru Y, Tsukamoto A, Hiramatsu H, Kamiya T, Tanabe K and Hosono H 2011 Nat. Commun. 2 409
[12] Si W, Zhang C, Shi X, Ozaki T, Jaroszynski J and Li Q 2015 Appl. Phys. Lett. 106 032602
[13] Sarnele E, Nappi C, Camerlingo C, Enrico E, Bellingeri E, Kawale S, Braccini V, Leveratto A and Ferdeghini C 2017 IEEE Trans. Appl. Supercond. 27 7400104
[14] Hilgenkamp H and Mannhart J 2002 Rev. Mod. Phys. 74 485
[15] Sakagami A, Kagawuchi T, Tabuchi M, Ujihara T, Takeda Y and Ikuta H 2013 Physica C 494 181
[16] Zhang Q, Zhang X, Yao C, Huang H, Wang D, Dong C, Ma Y, Ogino H and Awaji S 2017 Supercond. Sci. Technol. 30 065004
[17] Iida K et al 2014 Appl. Phys. Lett. 105 172602
[18] Iida K, Hänisch J and Tarantini C 2018 Appl. Phys. Rev. 5 031304
[19] Omura T, Matsumoto T, Hatano T, Iida K and Ikuta H 2018 J. Phys.: Conf. Ser. 1054 012024
[20] Kagawuchi T, Uemura H, Ohno T, Tabuchi M, Ujihara T, Takenaka K, Takeda Y and Ikuta H 2010 Appl. Phys. Lett. 97 042509
[21] Chihara M, Sumiyama N, Ariki K, Ichinose A, Tsukada I, Hatano T, Iida K and Ikuta H 2015 Physica C 518 69
[22] Kagawuchi T, Uemura H, Ohno T, Tabuchi M, Ujihara T, Takeda Y and Ikuta H 2011 Appl. Phys. Express 4 083102
[23] Huang Q, Zhao J, Lynn J W, Chen G F, Luo J L, Wang N L and Dai P 2008 Phys. Rev. B 78 054529
[24] Zhao J et al 2008 Nat. Mater. 7 953
[25] Díaz A, Mechin L, Berghuis P and Evetts J E 1998 Phys. Rev. B 58 R2960
[26] Verebelyi D T et al 2000 Appl. Phys. Lett. 76 1755
[27] Chiara T et al 2016 Sci. Rep. 6 36047
[28] Matsumoto T, Kondo K, Hatano T, Iida K and Ikuta H in preparation
[29] Straumal B B, Bokshtein B S, Straumal A B and Petelin A L 2008 JETP Lett. 88 537
[30] Wang C et al 2009 Phys. Rev. B 79 054521
[31] Kim S K, Tillman M E, Kim H, Kracher A, Bud’ko S L, Prozorov R and Canfield P C 2010 Supercond. Sci. Technol. 23 054008