Thickness-induced crossover from strong to weak collective pinning in exfoliated FeTe$_{0.6}$Se$_{0.4}$ thin films at 1 T

Ryoya Nakamura,¹ Masashi Tokuda,¹ Mori Watanabe,¹ Masamichi Nakajima,¹ Kensuke Kobayashi,¹,² and Yasuhiro Niimi¹,³

¹Department of Physics, Graduate School of Science, Osaka University, Toyonaka 560-0043, Japan
²Institute for Physics of Intelligence and Department of Physics, Graduate School of Science, The University of Tokyo, Tokyo 113-0033, Japan
³Center for Spintronics Research Network, Osaka University, Toyonaka 560-8531, Japan

We studied flux pinning in exfoliated FeTe$_{0.6}$Se$_{0.4}$ thin-film devices with a thickness $d$ from 30 to 150 nm by measuring the critical current density $J_c$. In bulk FeTe$_{0.6}$Se$_{0.4}$, the flux pinning has been discussed in the framework of weak collective pinning, while there is little knowledge on the pinning mechanism in the thin-film region. From the thickness $d$ dependence of $J_c$ at a fixed magnetic field of 1 T, we found that the strong pinning is dominant below $d \approx 70$ nm, while the weak collective pinning becomes more important above $d \approx 100$ nm. This crossover thickness can be explained by the theoretical model proposed by van der Beek et al [Phys. Rev. B 66, 024523 (2002)].

I. INTRODUCTION

Since 2008, iron-based superconductors (IBSs) have been well studied as a new group of high-temperature superconductors because of the two-dimensional layered structure similar to cuprate superconductors [1]. Their fundamental properties such as the short coherence length and large upper-critical field are useful for superconducting wires [2]. For such applications, it is essential to increase the critical current density $J_c$. In order to achieve this, the pinning mechanism of magnetic vortices needs to be clarified.

Among all kinds of IBSs, the iron-based chalcogenide FeTe$_{1-x}$Se$_x$ with $x = 0.4 \sim 0.5$ (FTS) has the simplest crystal structure with only FeCh ($Ch = Se$ or Te) layers. There have been many reports on bulk [3–12] and thin-film FTS single crystals, which enable a detailed discussion of the pinning mechanism. Although the critical temperature $T_c$ is not so high, the upper critical field is comparable to other IBSs. Thus, FTS has a potential to replace low-temperature superconducting wires such as Nb-Ti superconductors. In addition, FTS is a candidate for topological superconductor. Particularly it has been intensively studied on the detection of zero energy vortex bound state (ZVBS) in magnetic vortices, which is a fingerprint of the Majorana quasiparticles [17,20]. In fact, a recent scanning tunneling microscopy study demonstrated the existence of ZVBS in magnetic vortices [19]. Simultaneously, however, it also revealed that some of the magnetic vortices do not contain the ZVBS. This suggests that the Majorana zero modes gain finite energies via the interaction between magnetic vortices, resulting in an energy splitting of the zero modes. In order to unveil the ZVBS in the superconducting FTS, it is helpful to understand the magnetic vortex lattice and also its pinning mechanism under finite magnetic fields.

In general, there are two pinning mechanisms in superconductors, i.e., weak collective pinning [21,22] and strong pinning [23,24]. In the former case, the pinning stems from the atomic-scale inhomogeneity of the superconducting regions due to impurities or defects and is further categorized into two types. One is $\delta l$ pinning, originating from spatial fluctuations of the mean free path $l$ due to lattice defects [21]. This is related to the derivative of the macroscopic wave function in the Ginzburg-Landau theory. The other is $\delta T_c$ pinning, originating from spatial fluctuations of the critical temperature $T_c$ [21]. This is related to the fact that the coefficient $\alpha$ of the probability density for the macroscopic wave function in the Ginzburg-Landau theory is proportional to $T_c - T$. For FTS single crystals, the pinning mechanisms have been determined by analyzing magnetization measurements with the Bean model [3,5]. In most cases the $\delta l$ pinning is dominant [3,5], while in some cases both the $\delta l$ and $\delta T_c$ pinning coexist [7,8].

Contrary to the weak collective pinning, the strong pinning takes place at large-sized defects comparable to the coherence length ($\approx 10$ nm) and can induce a high critical current density [23,24]. It has been established that the strong pinning is a dominant mechanism for 100–300 nm thick YBa$_2$Cu$_3$O$_x$ (YBCO) films where the critical current density has a large value [24,27]. Although there are some reports on enhanced critical current densities [23,30] as well as strong pinning [31,33] in IBS thin films, the relation between the strong pinning and the thickness of IBSs has not been fully elucidated yet.

In this work, we directly obtained $J_c$ in exfoliated FeTe$_{0.6}$Se$_{0.4}$ thin-film devices with several different thicknesses $d$ by measuring current-voltage properties at a fixed out-of-plane magnetic field of 1 T. For $d \lesssim 70$ nm, $J_c$ linearly increases with increasing $d$. Such a tendency is consistent with the strong pinning reported in YBCO thin-film superconductors. When $d$ exceeds $\approx 100$ nm, $J_c$ decreases with increasing $d$ and eventually approaches the value estimated for bulk FTS. By plotting $J_c$ as a
function of $d$, we observe a crossover behavior from the strong pinning in thin-film regions to the weak collective pinning. This crossover thickness $d' = 70 \sim 100$ nm can be explained by the theoretical model proposed by van der Beek et al [24].

II. EXPERIMENTAL DETAILS

Single crystals of FeTe$_{0.8}$Se$_{0.4}$ were grown from a stoichiometric mixture of Fe, Te, and Se powder. We note that in the present work, FeTe$_{0.8}$Se$_{0.4}$ has been studied, but we use the same abbreviation (FTS) as for a general composition ratio FeTe$_{1-x}$Se$_x$ in this paper. The mixture was loaded into an alumina crucible and sealed in an evacuated quartz tube. The quartz tube was heated at 650°C for 10 h and then 1070°C for 15 h, followed by cooling down to 620°C at a rate of 3°C/h. FTS has a van-der-Waals interaction between the two adjacent Ch layers, which makes it easier to fabricate thin-film devices using the mechanical exfoliation technique. On the other hand, it has been known that excess Fe is present in between the Ch layers, which gives rise to magnetic correlations and suppresses the superconductivity [34]. To remove the excess Fe, we annealed as-grown FTS bulk crystals at 400°C for 40 h under 1% atmosphere of O$_2$ gas [35-37]. To determine $T_c$ of FTS bulk crystals, we measured the dc magnetic susceptibility using Magnetic Property Measurement System (Quantum Design). Figure 1(a) shows the temperature dependence of the magnetic susceptibility $\chi$ for as-grown and annealed FTS. In this work, $T_c^{\text{balk}}$ is defined as a temperature where $\chi$ for zero-field cooling (ZFC) starts to decrease. From the $\chi$ measurements, we determine $T_c^{\text{balk}} = 13.7$ K for the as-grown sample and $T_c^{\text{balk}} = 14.4$ K for the annealed crystal. By annealing the FTS crystal, $T_c^{\text{balk}}$ has been enhanced by 5%, which is consistent with Refs. [35-37]. Hereafter, we mainly focus on the annealed FTS unless otherwise noted, whereas we also use the as-grown FTS for comparison.

To obtain thin-film devices, we adopted the mechanical exfoliation technique using Scotch tapes under ambient conditions. Some of the exfoliated FTS flakes onto the Scotch tape were transferred to a thermally oxidized silicon substrate. We then coated polymethyl-methacrylate resist on the substrate and patterned electrodes with electron beam lithography. After the development of the resist, Ti (5 nm) and Au (150 nm) were deposited. Ti works as an adhesion layer for the Si/SiO$_2$ substrate. Before the deposition of electrodes, we performed Ar milling process for 55 s to remove the residual resist at the surface of FTS. In fact, we have confirmed the following: (1) This process does not give significant damage to FTS; (2) without the Ar milling process, we cannot obtain an Ohmic contact between FTS and the electrodes because of the residual resist at the interface. Figure 1(b) shows an optical microscope image of a typical device. The thickness $d$ of FTS was determined with a commercially available atomic force microscopy after finishing transport measurements. Although we sometimes found not perfectly homogeneous thin-film flakes, we regarded $d$ as a film thickness when surface regions with $d$ exceeds 85%. In the present work, $d$ ranges from 30 to 150 nm. Because of the limitation of the present exfoliation method, we could not obtain FTS devices thicker than 150 nm. We measured the resistivity $\rho$ of thin-film FTS devices with a standard ac Lock-in technique. To determine $J_c$, we performed dc current ($I$)-voltage ($V$) measurements.

III. EXPERIMENTAL RESULTS

First, to determine $T_c$ of FTS thin-film devices, we measured the temperature $T$ dependence of the resistivity $\rho$ in Fig. 2(a). When the film thickness $d$ is larger than 100 nm, $\rho$ decreases with decreasing $T$ for the whole temperature region. Below $d \approx 100$ nm, on the other hand, $\rho$ slightly increases with decreasing $T$, takes a maximum and becomes zero at low temperatures. In particular, a 30 nm thick device shows insulating behavior down to 13 K. Such tendencies are due to the inhomogeneity...
FIG. 2. (a) Temperature $T$ dependence of $\rho$ of FTS thin-film devices. The inset shows the closeup near $T_c$. We define two $T_c$, i.e., $T_{c}^{\text{zero}}$ and $T_{c}^{\text{on}}$, as detailed in the main text. (b) Thickness $d$ dependence of $T_{c}^{\text{zero}}$ and $T_{c}^{\text{on}}$. We also plot $T_{c}^{\text{bulk}}$ of annealed bulk FTS crystal for reference.

of superconducting states in FTS crystals, as reported in Ref. [38]. The superconducting percolation network, which is strongly connected in bulk, gradually weakens with decreasing $d$. This results in weak superconductivity for thinner FTS film devices. Figure 2(b) shows the thickness dependence of $T_c$. There are two ways to define $T_c$: One is to use the initial rise of resistivity from zero ($T_{c}^{\text{zero}}$), and the other is to use the onset of resistivity drop in $\rho$-$T$ curve ($T_{c}^{\text{on}}$) as indicated in Fig. 2(a). Since there is a finite voltage jump even just below $T_{c}^{\text{on}}$ [see Fig. 3(a)], in this work we have adopted $T_{c}^{\text{on}}$ as $T_c$, namely, $T_c \equiv T_{c}^{\text{on}}$. $T_c$ starts to decrease below $d \lesssim 100$ nm and should vanish at $10 \sim 20$ nm. According to Ref. [38], the critical thickness $d_0$ below which $T_c$ vanishes is 12 nm, which is consistent with the present work.

Next we performed $I$-$V$ measurements under the out-of-plane magnetic field $\mu_0 H_{||c} = 1$ T. Figure 3(a) shows $dc$ current-voltage curves for a 95 nm thick FTS device at several different temperatures. We use the current density $J$ instead of $I$ to compare with different $d$ devices. At low temperatures, there is a clear hysteresis that originates from the Joule heating in the FTS device. In the present work, we define $J_c$ as the current density when the measured $dc$ voltage exceeds a threshold value of 1 mV. For example, $J_c$ is 0.78 MA/cm$^2$ at $T = 3$ K. In the vicinity of $T_c$, on the other hand, the hysteresis vanishes and the voltage jump near $J_c$ becomes less clear. We also performed pulse current measurements with a pulse width of 1 ms and an interval of 100 ms in order
to evaluate $J_c$ without the effects of Joule heating. The obtained $J_c$ was almost the same value as in Fig. 3(a). In Fig. 3(b), $J_c$ obtained from Fig. 3(a) is plotted as a function of $T$. $J_c$ monotonically decreases with increasing $T$ and vanishes at $T_c$. We also measured the magnetic field and angle dependences of $J_c$ for the $d = 95$ nm thick FTS device at $T = 3$ K. Figure 4(a) shows $J_c$ as a function of $\mu_0H$ along the $c$ axis and the $ab$ plane. When the magnetic field is applied along the $ab$ plane, $J_c$ is more or less constant. For the perpendicular magnetic field ($\mu_0H \parallel c$), on the other hand, $J_c$ decreases with increasing $\mu_0H$. Figure 4(b) shows $J_c$ at $\mu_0H = 1$ T as a function of the rotation angle $\theta$ from the $c$ axis. $J_c(\theta)$ has a broad maximum at $\theta = 90^\circ$ ($\mu_0H \parallel ab$), and there is no peak at $\theta = 0^\circ$ nor $180^\circ$ ($\mu_0H \parallel c$). If we assume that $J_c(\theta, \mu_0H)$ depends only on the $c$ axis component of the applied magnetic field, i.e., $\mu_0H_{||c} = \mu_0H \cos \theta$, the angle dependence of $J_c$ is expected as shown by the red curve, which is consistent with the experimental data. This clearly shows that $J_c$ depends only on the perpendicular component of the magnetic field. Such magnetic field and angle dependences of $J_c$ are consistent with previous works on FTS thin films [13–16].

Now we move on to the $d$ dependence of $J_c$. In Fig. 5(a), we show the temperature dependence of $J_c(T)$ at $\mu_0H_{||c} = 0$ T (black circles) and 1 T (red triangles) for $d = 30$ and 110 nm. The green dotted line is the best fit for $d = 30$ nm at $\mu_0H_{||c} = 0$ T using Eqs. 2–5. (b) Thickness dependence of $J_{c0}$ for $\mu_0H_{||c} = 0$ T (black circles) and 1 T (red triangles). For comparison, we also plot $J_c$ at $\mu_0H_{||c} = 1$ T for bulk FTS in Ref. [7] (red square). The solid line is the best fit with Eq. 1 for $J_{c0}$ below $d = 50$ nm.
indicates that there are two regions below \( d \approx 70 \) nm and above \( d \approx 100 \) nm; the former is the strong pinning region and the latter is the weak collective pinning region, as discussed in more detail in the next section.

IV. DISCUSSIONS

A. Strong pinning

We first discuss the strong pinning region below \( d \approx 70 \) nm. It is known that FTS is a highly inhomogeneous material. According to electron energy loss spectroscopy measurements, there is an inhomogeneous distribution of Te in FTS crystals; it forms some clusters in FTS and the spatial variation of the cluster is approximately 10 nm. Therefore, the effect of the inhomogeneity should be prominent when \( d \) is of the order of 10 nm, where the strong pinning is more dominant.

Lower magnetic fields are favorable to confirm whether the strong pinning is essential in the thin-film FTS devices. As shown in Fig. 5(a), \( J_c \) for the \( d = 30 \) nm thick device can be fitted with Eqs. (2)–(4). The evaluated \( J_{c0} \) does not depend on the applied magnetic field at least within 1 T and linearly increases with increasing \( d \). In fact, these features cannot be explained by the weak collective pinning theory, as detailed below. This fact also suggests that Eqs. (2)–(4) would be useful even for the strong pinning.

As mentioned above, \( J_{c0} \) at 0 T has the same value as that in our thin-film devices with \( d \geq 100 \) nm. Therefore, we analyze \( J_c(T) \) using the weak collective pinning theory, as discussed in previous works. In the theoretical approach proposed by Griessen et al., \( J_c(T) \) in the moderately low field region, where the strong pinning can be ignored and the motion of single vortex is essential, is given as

\[
J_{c0}^b(t) = J_{c0}^b(1 - t^2)^{5/2} (1 + t^2)^{-1/2}
\]

for \( \delta l \) pinning and

\[
J_{c0}^{bT_c}(t) = J_{c0}^{bT_c}(1 - t^2)^{7/6} (1 + t^2)^{5/6}
\]

for \( \delta T_c \) pinning. \( t \) is the reduced temperature \( (t = T/T_c) \).

B. Weak collective pinning

It has been established that \( J_c \) in bulk FTS can be described by the weak collective pinning theory. As shown in Fig. 3(b), \( J_c \) at 1 T in bulk FTS is very close to that in our thin-film devices with \( d \geq 100 \) nm. Therefore, we analyze \( J_c(T) \) by Eqs. (2)–(4) for the weak collective pinning theory, as discussed in Ref. [24].

\[
J_{c0}^w = \frac{\xi_{ab}}{4\pi\Phi_0}\ln\left(1 + \frac{D_{ab}^2}{2\xi_{ab}}\right)\frac{D_{c}}{(d^* - d_0)^2(d - d_0)},
\]

where \( \xi_0 = (\Phi_0/4\pi\lambda_{ab})^{2}(4\pi/\mu_0) \) is the typical energy scale for single vortex, \( \Phi_0 = h/2e \) is the flux quantum, \( \lambda_{ab} \) is the in-plane magnetic penetration depth at \( T = 0 \), \( \mu_0 = 4\pi \times 10^{-7} \) H/m, \( \xi_{ab} \) is the in-plane coherence length at \( T = 0 \), and \( d_0 \) is the critical thickness where \( J_{c0} \) becomes zero. \( D_{ab} \) and \( D_{c} \) are the typical sizes of defect along the \( ab \)-plane and \( c \)-axis, respectively; \( d^* \) is the crossover film thickness from the very thin-film region to relatively thick film region.

As described in Ref. [24], \( J_c \) in the very thin-film region should be field-independent and linearly increases with increasing \( d \). This is consistent with the \( d \) dependence of \( J_{c0} \) for \( d \geq 70 \) nm in Fig. 4(b), at least in the field range from 0 to 1 T. Furthermore, \( d_0 \) is also consistent with the thickness below which \( T_c \) vanishes in Fig. 4(b). The contribution of the strong pinning is more significant in this thickness range.

As we increase \( d \) further from 70 nm, a significant difference in \( J_{c0} \) is observed between \( \mu_0H = 0 \) and 1 T. The crossover film thickness \( d^* \) for both magnetic fields and suddenly decreases with increasing \( d \). This thickness range will be discussed in the next subsection.
The total critical current density at $T_{c}$ is shown in Fig. 6. (a) Thickness dependence of $\eta^{\delta l}$ (blue circles) and $\eta^{\delta T_c}$ (red circles) for annealed FTS samples. We also plot $\eta^{\delta l}$ (blue triangles) and $\eta^{\delta T_c}$ (red triangles) for as-grown FTS samples. For comparison, we also plot $\eta$ for bulk FTS in Ref. [7] (square). (b) Temperature dependence of $J_c(T)$ for $d = 105$ nm thick as-grown device under the out-of-plane magnetic field $\mu_0 H_{\perp} = 1$ T. The green dotted line is the best fit with Eqs. (2)–(4), while the red and blue dashed lines are the contributions from $\delta T_c$ and $\delta l$ pinning, respectively.

$J_{c0} = J_c^{\delta l} + J_c^{\delta T_c}$ is the total critical current density at $T = 0$. As mentioned in the previous subsection, the fitting with Eqs. (2)–(4) works even for the strong pinning region. In the present $d$ range, $\eta^{\delta T_c}$ is much larger than $\eta^{\delta l}$, indicating that the $\delta T_c$ pinning is more dominant. With increasing $d$, the proportion of $\eta^{\delta l}$ becomes larger and exceeds that of $\eta^{\delta T_c}$ in bulk FTS. This fact suggests that a crossover from the strong pinning to the weak collective pinning occurs at $d \approx 100$ nm.

Let us discuss the reason why the $\delta T_c$ pinning is dominant in our thin-film devices. One possibility is the existence of excess Fe atoms. In most previous works on bulk FTS [4,5], where as-grown samples were used, the $\delta l$ pinning is dominant. In such a case, a small amount of Fe impurities would be in between the two chalcogen layers. To remove the Fe impurities, Sun et al. performed oxygen annealing for FTS and found that both the $\delta l$ and the $\delta T_c$ pinnings coexist [7]. As mentioned in Sec. II, excess Fe (i.e., atomic-scale) impurities would be segregated at the surface of FTS by annealing it under an oxygen flow. This results in the reduction of the $\delta l$ pinning contribution. To confirm the above scenario, we fabricated as-grown FTS thin-film devices and measured the temperature dependence of the critical current density in the same way as for annealed FTS devices. Figure 6(b) shows the temperature dependence of $J_c(T)$ for a 105 nm thick as-grown FTS device and also the result of fitting with Eqs. (2)–(4). When we measured $I-V$ curves for as-grown devices, the critical current changed with every measurement. Therefore, the data points in Fig. 6(b) are scattered compared with those for the annealed FTS device shown in Fig. 6(a). The contribution of $\delta l$ pinning is comparable to that of $\delta T_c$ pinning, even at $d \approx 100$ nm where the latter is much more dominant than the former in the annealed devices. This clearly shows that the excess Fe plays an important role in the pinning mechanism in FTS. We note that even in as-grown FTS thin-film devices, the contribution of $\delta T_c$ pinning is larger than that of the annealed bulk FTS sample.

Another possibility is the similarity between $\delta T_c$ pinning and strong pinning. In the case of strong pinning at nanometer-sized defects, the $\delta l$ pinning is less important than the $\delta T_c$ pinning, as pointed out in Ref. [24]. This is because the scattering cross section of quasiparticles at atomic-scale defects is negligibly small compared with that at large-sized defects. In thin-film FTS devices [42], the inhomogeneous distribution of Te would be more essential. Therefore, the contribution of the $\delta T_c$ pinning is much more dominant in annealed FTS thin-film devices and is still large even in as-grown FTS thin-film devices which contain many atomic-scale defects. In bulk FTS, on the other hand, quasiparticles may scatter at atomic-scale defects and thus the $\delta l$ pinning can be dominant because the path of the superconducting state is robust and the inhomogeneous distribution is not crucial [38].

V. CONCLUSION

We have studied the critical current density $J_c$ in FeTe$_{0.6}$Se$_{0.4}$ (FTS) devices with thicknesses $d$ from 30 to 150 nm by means of electric transport measurements. Below $d \approx 70$ nm, $J_c$ does not depend on the applied magnetic field and linearly increases with increasing $d$. This result is consistent with the strong pinning theory. Above $d \approx 100$ nm, on the other hand, $J_c$ decreases with increasing $d$ and becomes comparable to that for bulk FTS, suggesting that the thicker film region corresponds to the weak collective pinning region. From these experimental results, a crossover from the strong pinning to the weak collective pinning is realized at $d^* = 70 \sim 100$ nm at a fixed magnetic field of 1 T. This value can be explained by the theoretical model where a defect size is comparable to the inhomogeneity of Te ($\approx 10$ nm) in FTS.
Moreover, the $\delta T$, pinning is much more dominant than the $\delta l$ pinning in our thin-film devices. This is partly due to the fact that our FTS have been annealed under an oxygen flow before the exfoliation, resulting in a reduction of excess Fe (i.e., atomic-size impurities) in between the two chalcogen layers.

ACKNOWLEDGMENTS

This work was supported by JSPS KAKENHI (Grants No. JP16H05964, No. JP17K18756, No. JP19K21850, No. JP19H00656, No. JP19H05826, No. JP20H02557, No. JP20J02299, and No. JP21J20477), the Mazda Foundation, the Shimadzu Science Foundation, the Yazaki Memorial Foundation for Science and Technology, the SCAT Foundation, the Murata Science Foundation, Toyota Riken Scholar, the Kato Foundation for Promotion of Science, and the Asahi Glass foundation.

[1] Y. Kamihara, T. Watanabe, M. Hirano, and H. Hosono, J. Am. Chem. Soc. 130, 3296 (2008).
[2] H. Hosono, A. Yamamoto, H. Hiramatsu, and Y. Ma, Mater. Today 21, 278 (2018).
[3] B. C. Sales, A. S. Sefat, M. A. McGuire, R. Y. Jin, D. Mandrus, and Y. Mozharivskyj, Phys. Rev. B 79, 094521 (2009).
[4] P. Das, A. D. Thakur, A. K. Yadav, C. V. Tomy, M. R. Lees, G. Balakrishnan, S. Ramakrishnan, and A. K. Grover, Phys. Rev. B 84, 214526 (2011).
[5] M. Bonura, E. Gianinni, R. Viennois, and C. Senatore, Phys. Rev. B 85, 134532 (2012).
[6] A. Galluzzi, K. Buchkov, V. Tomov, E. Nazarova, A. Leo, G. Grimaldi, S. Pace, and M. Polichetti, Supercond. Sci. Technol. 33, 094006 (2020).
[7] Y. Sun, T. Taen, Y. Tsuchiya, S. Pyon, Z. Shi, and T. Tamegai, Europhys. Lett. 103, 57013 (2013).
[8] T. Tamegai, Y. Sun, T. Yamada, and S. Pyon, IEEE-Transactions on Applied Superconductivity 26, 1 (2016).
[9] C. Yadav and P. Paulose, Solid State Commun. 151, 216 (2011).
[10] D. Miu, T. Noji, T. Adachi, Y. Koike, and L. Miu, Supercond. Sci. Technol. 25, 115009 (2012).
[11] M. Shahbaz, X. L. Wang, S. X. Dou, H. Fang, and C. T. Li, J. Appl. Phys. 113, 17E115 (2013).
[12] Z. F. Wu, Z. H. Wang, J. Tao, L. Qu, S. G. Yang, and H. H. Wen, Supercond. Sci. Technol. 29, 035006 (2016).
[13] K. Iida, J. Hänisch, E. Reich, F. Kurth, R. Hühne, L. Schultz, B. Holzapfel, A. Ichinose, M. Hanawa, I. Tsukada, M. Schulze, S. Aswartham, S. Wurmehl, and B. Büchner, Phys. Rev. B 87, 104510 (2013).
[14] V. Braccini, S. Kawale, E. Reich, E. Bellingeri, L. Pellegrino, A. Sala, M. Putti, K. Higashikawa, T. Kiss, B. Holzapfel, and C. Ferdeghini, Appl. Phys. Lett. 103, 176201 (2013).
[15] P. Yuan, Z. Xu, Y. Ma, Y. Sun, and T. Tamegai, Supercond. Sci. Technol. 29, 035013 (2016).
[16] H. Bryja, R. Hühne, K. Iida, S. Molatta, A. Sala, M. Putti, L. Schultz, K. Nielsch, and J. Hänisch, Supercond. Sci. Technol. 30, 115005 (2017).
[17] P. Zhang, K. Yaji, T. Hashimoto, Y. Ota, T. Kondo, K. Okazaki, Z. Wang, J. Wen, G. D. Gu, H. Ding, and S. Shin, Science 360, 182 (2018).
[18] D. Wang, L. Kong, P. Fan, H. Chen, S. Zhu, W. Liu, L. Cao, Y. Sun, S. Du, J. Schneeloch, R. Zhong, G. Gu, L. Fu, H. Ding, and H. J. Gao, Science 362, 333 (2018).
[19] T. Machida, Y. Sun, S. Pyon, S. Takeda, Y. Kohsaka, T. Hanaguri, T. Sasagawa, and T. Tamegai, Nat. Mater. 18, 811 (2019).
[20] S. Zhu, L. Kong, L. Cao, H. Chen, M. Papaj, S. Du, Y. Xing, W. Liu, D. Wang, C. Shen, F. Yang, J. Schneeloch, R. Zhong, G. Gu, L. Fu, Y. Y. Zhang, H. Ding, and H. J. Gao, Science 367, 189 (2020).
[21] G. Blatter, M. V. Feigel’man, V. B. Geshkenbein, A. I. Larkin, and V. M. Vinokur, Rev. Mod. Phys. 66, 1125 (1994).
[22] R. Griessen, W. Hailu, A. J. J. van Dalen, B. Dam, J. Rector, H. G. Schmack, S. Libbrecht, E. Osquiguil, and Y. Bruynseraede, Phys. Rev. Lett. 72, 1910 (1994).
[23] Y. N. Ovchinnikov and B. I. Ivlev, Phys. Rev. B 43, 8024 (1991).
[24] C. J. van der Beek, M. Konczykowski, A. Abal’oshev, I. Abal’osheva, P. Gierlowski, S. J. Lewandowski, M. V. Indenbom, and S. Barbanera, Phys. Rev. B 66, 024523 (2002).
[25] W. K. Yeoh, J. Horvat, S. X. Dou, and V. Keast, Supercond. Sci. Technol. 17, S572 (2004).
[26] Z. Chen, F. Kametani, Y. Chen, Y. Xie, V. Selvamanickam, and D. C. Larbalestier, Supercond. Sci. Technol. 22, 55013 (2009).
[27] A. O. Ijaduola, J. R. Thompson, R. Feenstra, D. K. Mandrus, and G. Grimaldi, Supercond. Sci. Technol. 17, 84021 (2012).
[28] P. Mele, K. Matsumoto, K. Fujita, Y. Yoshida, T. Kiss, A. Ichinose, and M. Mukaida, Supercond. Sci. Technol. 25, 85006 (2012).
[29] W. Si, S. J. Han, X. Shi, S. N. Ehrlich, J. Jaroszynski, A. Goyal, and Q. Li, Nat. Commun. 4, 1347 (2013).
[30] J. Lee, J. Jiang, F. Kametani, M. J. Oh, J. D. Weiss, Y. Collantes, S. Seo, S. Yoon, C. Tarantini, Y. J. Jo, E. E. Hellstrom, and S. Lee, Supercond. Sci. Technol. 30, 85006 (2017).
[31] K. Iida, J. Hänisch, S. Trommler, V. Matias, S. Haindl, F. Kurth, I. L. del Pozo, R. Hühne, M. Kidszun, J. Engelmann, L. Schultz, and B. Holzapfel, Appl. Phys. Express 4, 013103 (2010).
[32] Y. Zhang, C. T. Nelson, S. Lee, J. Jiang, C. W. Bark, J. D. Weiss, C. Tarantini, C. M. Folkman, S.-H. Baek, E. E. Hellstrom, D. C. Larbalestier, C.-B. Eom, and X. Pan, Appl. Phys. Lett. 98, 042509 (2011).
[33] E. Bellingeri, S. Kawale, I. Pallecchi, A. Gerbi, R. Buzio, V. Braccini, A. Palenzona, M. Putti, M. Adamo, E. Sar-
nelli, and C. Ferdeghini, Appl. Phys. Lett. 100, 082601 (2012).

[34] M. Bendele, S. Weyeneth, R. Puzniak, A. Maisuradze, E. Pomjakushina, K. Conder, V. Pomjakushin, H. Luetkens, S. Katrych, A. Wisniewski, R. Khasanov, and H. Keller, Phys. Rev. B 81, 224520 (2010).

[35] Y Sun, T Taen, Y Tsuchiya, Z X Shi, and T Tamegai, Supercond. Sci. Technol. 26, 015015 (2013).

[36] Y. Sun, Y. Tsuchiya, T. Taen, T. Yamada, S. Pyon, A. Sugimoto, T. Ekino, Z. Shi, and T. Tamegai, Sci. Rep. 4, 4585 (2014).

[37] Y. Sun, Z. Shi, and T. Tamegai, Supercond. Sci. Technol. 32, 103001 (2019).

[38] C. Yue, J. Hu, X. Liu, A. M. Sanchez, Z. Mao, and J. Wei, ACS Nano 10, 429 (2016).

[39] B. Joseph, A. Iadecola, A. Puri, L. Simonelli, Y. Mizuguchi, Y. Takano, and N. L. Saini, Phys. Rev. B 82, 20502(R) (2010).

[40] H. Hu, J.-M. Zuo, J. Wen, Z. Xu, Z. Lin, Q. Li, G. Gu, W. K. Park, and L. H. Greene, New J. Phys. 13, 53031 (2011).

[41] U. R. Singh, S. C. White, S. Schmaus, V. Tsurkan, A. Loidl, J. Deisenhofer, and P. Wahl, Phys. Rev. B 88, 155124 (2013).

[42] Y. Zhu, L. Chen, J. Ciston, and H. Wang, J. Phys. Chem. C 117, 7170 (2013).

[43] C. J. van der Beek, G. Rizza, M. Konczykowski, P. Fertey, I. Monnet, T. Klein, R. Okazaki, M. Ishikado, H. Kito, A. Iyo, H. Eisaki, S. Shamoto, M. E. Tillman, S. L. Bud’ko, P. C. Canfield, T. Shibauchi, and Y. Matsuda, Phys. Rev. B 81, 174517 (2010).

[44] E. Sheriff, R. Prozorov, Y. Yeshurun, A. Shaulov, G. Koren, and C. Chabaud-Villard, J. Appl. Phys. 82, 4417 (1997).

[45] X. Wang and J. Z. Wu, Phys. Rev. B. 76, 184508 (2007).

[46] Y. V. Cherpak, V. L. Svetchnikov, A. V. Semenov, V. O. Moskaliuk, C. G. Tretiatchenko, V. S. Flis, and V. M. Pan, J. Phys.: Conf. Ser. 97, 012259 (2008).