Corrigendum: Influence of carbon-ion irradiation on the superconducting critical properties of MgB$_2$ thin films (2019 Supercond. Sci. Technol. 32 025006)

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We have found that there was an error in the calculation of the atomic density of Mg and B in MgB$_2$. The number of MgB$_2$ molecules per cm$^3$ ($\rho_{\text{MgB}_2}$) is around $3.4 \times 10^{22}$, so the number of Mg and B atoms per cm$^3$ must be equal to $\rho_{\text{MgB}_2}$ and twice that of $\rho_{\text{MgB}_2}$, respectively. However, the number of Mg and B atoms per cm$^3$ presented in page 7 correspond to 1/3 and 2/3 of $\rho_{\text{MgB}_2}$, respectively.

In order to estimate the atomic density, the mass density of bulk value is much more useful than that of strained sample such as a thin film. Since MgB$_2$ has the mass density of 2.57 g cm$^{-3}$ and the molar mass of 45.93 g mol$^{-1}$, and Avogadro’s number $N_A \approx 6.02 \times 10^{23}$ mol$^{-1}$, the $\rho_{\text{MgB}_2}$ is approximately $3.37 \times 10^{22}$ molecules cm$^{-3}$. Therefore, the number of Mg and B atoms per cm$^3$ in MgB$_2$ must be $\sim 3.37 \times 10^{22}$ atoms cm$^{-3}$ and $6.74 \times 10^{22}$ atoms cm$^{-3}$, respectively, and the atomic density of MgB$_2$ is around $10.11 \times 10^{22}$ atoms cm$^{-3}$.

We believe that the change in atomic density does not lead to any significant issues for all the results and conclusions drawn in the paper.

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Influence of carbon-ion irradiation on the superconducting critical properties of MgB2 thin films

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Abstract
We investigate the influence of carbon-ion irradiation on the superconducting (SC) critical properties of MgB2 thin films. MgB2 films of two thicknesses, 400 nm (MB400nm) and 800 nm (MB800nm), were irradiated by 350 keV C ions having a wide range of fluence, $1 \times 10^{12} – 1 \times 10^{15}$ C atoms cm$^{-2}$. The mean projected range ($R_p$) of 350 keV C ions in MgB2 is 560 nm, thus the energetic C ions will pass through the MB400nm, whereas the ions will remain into the MB800nm. The SC transition temperature ($T_c$), upper critical field ($H_{c2}$), c-axis lattice parameter, and corrected residual resistivity ($\rho_{corr}$) of both the films showed similar trends with the variation of fluence. However, a disparate behavior in the SC phase transition was observed in the MB800nm when the fluence was larger than $1 \times 10^{14}$ C atoms cm$^{-2}$ because of the different $T_c$s between the irradiated and non-irradiated parts of the film. Interestingly, the SC critical properties, such as $T_c$, $H_{c2}$, and critical current density ($J_c$), of the irradiated MgB2 films, as well as the lattice parameter, were almost restored to those in the pristine state after a thermal annealing procedure. These results demonstrate that the atomic lattice distortion induced by C-ion irradiation is the main reason for the change in the SC properties of MgB2 films.

Keywords: carbon-ion irradiation, upper critical field, thermal annealing, MgB2 thin films

(Some figures may appear in colour only in the online journal)
materials produce columnar defects along the ion tracks, which act as a strong flux-pinning source to improve the in-field $J_c$ of the high-$T_c$ cuprate superconductors [20, 21]. However, heavy-ion irradiations in MgB$_2$ seem to rapidly suppress its superconductivity rather than improve its critical SC parameters, such as $H_{c2}$ and $J_c$ [18, 22]. For these reasons, irradiation studies of MgB$_2$ have been mostly focused on neutron irradiation.

Irradiation is a unique technique that makes it possible to examine the effects of radiation and additional defects in an identical sample [13, 23, 24]. The study of radiation effects on a superconductor is important to test their application in a radiation environment, such as space, fusion reactions, and so on [13, 23]. In addition, ion irradiations can create defects in target materials mainly by nuclear stopping of energetic ions, although the exact mechanism for the formation of defects is still debatable [23, 25]. In superconductors, defects are normally beneficial for the enhancement of SC critical properties, especially in-field $J_c$, because defects in type-II superconductors can interrupt vortex motion. Therefore, understanding the effects of defects on SC critical properties is a key issue for technological applications of superconductors as well as for scientific advancement [26, 27].

In this study, we report the influence of carbon-ion irradiation on the SC critical properties of the $c$-axis-oriented MgB$_2$ thin films with thicknesses of 400 and 800 nm. The C was selected as the irradiating ion source because C has been known as one of the most effective elements to enhance $H_{c2}$ and in-field $J_c$ of MgB$_2$. Carbon ions of various fluences ranging between $1 \times 10^{13}$ and $1 \times 10^{15}$ C atoms cm$^{-2}$ with an incident energy of 350 keV (mean projected range $R_{p}$, $\sim 560$ nm) were irradiated into the prepared MgB$_2$ films. The SC properties after irradiations were slightly different depending on whether the irradiated C ions penetrated fully the film (400 nm) or were implanted into the film (800 nm). However, the degraded superconductivity of both the irradiated MgB$_2$ films was almost recovered to those in the pristine state after a thermal annealing procedure. These results indicate that the effect of the implanted C ions on the superconductivity of MgB$_2$ films is insignificant, whereas the displacement damage by irradiation is crucial to the change in $T_c$, $H_{c2}$, and $J_c$ of MgB$_2$.

2. Experimental

High-quality MgB$_2$ thin films of thicknesses 400 nm (MB400nm) and 800 nm (MB800nm) were fabricated on c-cut Al$_2$O$_3$ substrates for C-ion irradiation studies by using a hybrid physical-chemical vapor deposition method, the details of the growth technique were described earlier [28, 29]. C-ion irradiation was carried out using Cockcroft-Walton type 400 kV ion beam accelerator at Korea Institute of Science and Technology (KIST), Seoul. Various fluences of $1 \times 10^{13}$ (1E13), $5 \times 10^{13}$ (5E13), $1 \times 10^{14}$ (1E14), $2 \times 10^{14}$ (2E14), $5 \times 10^{14}$ (5E14), and $1 \times 10^{15}$ (1E15) C atoms cm$^{-2}$ with a tilted angle of 7° to avoid channeling effects during irradiation were irradiated into the prepared MgB$_2$ films at room temperature. The mean projected range ($R_{p}$) of the irradiated C ions, with an incident beam energy of 350 keV, was estimated to be around 560 nm from the Stopping and Range of Ions in Matter (SRIM), a Monte Carlo simulation program$^5$. Therefore, the irradiated C ions were remained into the MB800nm due to the $R_{p}$ being greater than the film thickness whereas most of the ions penetrated through the MB400nm.

The thickness of the MgB$_2$ thin films was measured by using a scanning electron microscope, and the crystallinity of the films before and after irradiation was checked by using an x-ray diffractometer. The electrical resistivity ($\rho$) was measured by using the standard four-probe technique with gold (Au) coating on the four-point contact regions to achieve good ohmic contact. The upper critical field ($H_{c2}$) for a magnetic field applied perpendicular to the $ab$ plane of the surface of the film was measured by the physical property measurement system (9 T, Quantum Design) and the temperature dependence of magnetization, $M(T)$, was measured by using the magnetic property measurement system (5 T, Quantum Design) before and after C-ion irradiation.

Thermal annealing was carried out for the MB400nm and MB800nm irradiated with the highest fluence of $1 \times 10^{15}$ atoms cm$^{-2}$ (1E15). In addition, a non-irradiated MB400nm was also prepared to check thermal annealing effects on the pristine sample. Three films were wrapped in tantalum (Ta) foil to minimize the Mg decomposition during thermal annealing, sealed inside an evacuated quartz tube, and then the ampoules were annealed at temperatures of 200 °C, 300 °C, 400 °C, and 500 °C for 30 min in a box furnace. The same irradiated films with the 1E15 were consecutively annealed at each temperature, and the non-irradiated MB400nm was annealed at 500 °C. The annealing experiments were conducted at temperatures up to 500 °C as large Mg deficiency can be induced at higher temperatures [30]. $M(T)$ and x-ray diffraction patterns of the irradiated films were examined after the annealing procedure for each annealing temperature ($T_A$). In addition, $\rho$, $H_{c2}$, and magnetization hysteresis ($M$-$H$) loops for the films annealed at $T_A = 500$ °C were investigated by using the same techniques mentioned above.

3. Results and discussion

Figure 1 describes the mean projected range ($R_{p}$) of the irradiated C ions with a beam energy of 350 keV into MgB$_2$, simulated by using SRIM software$^5$, presenting that most of the C ions are stopped at around $R_{p} \sim 560$ nm after undergoing a series of collisions with the Mg and B atoms. In view of the magnitude of the $R_{p}$, two MgB$_2$ thin films having thicknesses of 400 nm (MB400nm) and 800 nm (MB800nm) were irradiated with the 350 keV C ions, and while most of the energetic C ions penetrated through MB400nm, as shown in figure 1, they remained in MB800nm as defects or were substituted into B sites.

Figures 2(a) and (b) show the temperature dependence of the normalized electrical resistivity, $\rho(T)/\rho(41$ K), near the

$^5$ The mean projected range of C ions in the MgB$_2$ was calculated using the SRIM software (www.srim.org).
Figure 1. SRIM simulation for the range of irradiated carbon (C) ions with incident energy of 350 keV in MgB$_2$, where the density of MgB$_2$ is 2.57 g cm$^{-3}$ and the estimated mean projected range ($R_p$) is around 560 nm. Most of the irradiated C ions pass through MB400nm, but for the MB800nm, the ions are distributed around 560 nm below the MgB$_2$ surface.

Figure 2. Superconducting transitions measured by (a) and (b) electrical resistivity ($\rho$) and (c) and (d) magnetization ($M$) for MB400nm and MB800nm at different damage levels of irradiation. Kinks in the $M(T)$ of MB800 nm result from a different $T_c$ between the irradiated and non-irradiated layers of MgB$_2$.
SC transition regions for pristine (pri.) and irradiated MgB$_2$ films of MB400nm and MB800nm, respectively, whereby C-ion irradiation into MgB$_2$ thin films resulted in suppression of the SC transition temperature ($T_c$). In this case, $T_c$ was determined from 50% of the normal state resistivity value at SC onset temperature, as indicated by arrows in figures 2(a) and (b). For the films irradiated with higher fluences than $1 \times 10^{14}$ atoms cm$^{-2}$ (1E14), the SC transition width of MB800nm is much broader than that of MB400nm, while $T_c$ of MB800nm is higher than that of MB400nm. As shown in figures 2(c) and (d), the zero-field-cooled (ZFC) and field-cooled (FC) dc magnetization ($M$) data for MB400nm and MB800nm also showed similar results to the electrical resistivity. Here, $M(T)$ values were normalized by the absolute ZFC $M$ value at 5 K for comparison. The onset temperature for the diamagnetic signal gradually reduces with increasing fluence for both MB400nm and MB800nm. A kink in the ZFC $M(T)$ data, however, is apparent in the SC transition region for MB800nm irradiated with large fluences of $5 \times 10^{14}$ (5E14) and $1 \times 10^{15}$ atoms cm$^{-2}$ (1E15), while it is absent for MB400nm. Since the thickness of MB800nm is larger than $R_p$ ($\sim 560$ nm), the kink observed in the MB800nm may be ascribed to two different $T_c$s between the irradiated and non-irradiated parts of the MgB$_2$ films.

The temperature dependence of electrical resistivity ($\rho$) is comparatively plotted in figures 3(a) and (b) for MB400nm and MB800nm before and after irradiation, respectively. The resistivity increases over the whole temperature range with increasing fluence of the C-ion radiation. Figures 3(c) and (d) present a plot of $T_c$ as a function of the residual resistivity, $\rho_{41K} = \rho(41 \text{ K})$, and corrected residual resistivity ($\rho_{\text{corr}}$) for MB400nm and MB800nm, respectively, where $\rho_{\text{corr}}$ was introduced to exclude the effects of grain boundaries on the resistivity and is defined as $\rho_{\text{corr}} = \rho_{41K} \times \Delta \rho_{\text{ideal}}/\Delta \rho$, where $\Delta \rho = \rho(300 \text{ K}) - \rho(41 \text{ K})$ and $\Delta \rho_{\text{ideal}}$ is $\Delta \rho$ of MgB$_2$ with full connectivity [12, 15, 31]. Here, we used $\Delta \rho_{\text{ideal}} = 4.3 \mu\Omega$ cm determined from a MgB$_2$ single crystal because $\Delta \rho$ of the pristine MB400nm and MB800nm is 6.4 $\mu\Omega$ cm and 5.9 $\mu\Omega$ cm, respectively [31]. For fluences higher than 1E14, the $\rho_{41K}$ values of MB800nm is much larger than those of MB400nm, while the reduction rate of $T_c$ for MB800nm is weaker than that for MB400nm. When $\rho_{41K}$ is converted to
ρcorr, however, $T_c$ of MB400nm and MB800nm are well scaled on a single curve, as shown in figure 3(d) [13, 15]. These results indicate that, despite a larger residual resistivity than MB400nm, the higher $T_c$ in MB800nm results from the proximity effect between the irradiated and non-irradiated parts in it. In addition, the decrease in $T_c$ of both the films can be attributed to enhanced intra-grain scattering by C-ion irradiation [13, 31], whereas the effect of implanted C ions on $T_c$ seems negligible in the MB800nm films.

Figures 4(a) and (b) show the x-ray diffraction (XRD) patterns of a $\theta$–$2\theta$ scan for pristine and irradiated MB400nm and MB800nm, respectively, showing a shift of (00l) peak positions of MgB$_2$ due to C-ion irradiation. Figures 4(c) and (d) are the magnified views of the (002) peak of MB400nm and MB800nm, respectively. With increasing fluence, the peak position is progressively shifted towards a lower angle and the peak width is broadened [12, 14, 15, 32], reflecting an expansion of the c-axis lattice parameter and degradation of crystallinity of the MgB$_2$ films. The (002) peak is considerably broadened for the irradiated MB800nm than for the MB400nm. In addition, MB800nm irradiated with 1E15 show a double-peak like behavior due to the implanted C ions around the $R_p$, as shown in figure 1.

The dependence on C-ion fluence of the c-axis lattice constant, $\rho_{41K}$, and $T_c$ for MB400nm and MB800nm are plotted in figures 5(a)–(c), respectively. In comparison with the pristine films, the c-axis lattice parameter of MB400nm and MB800nm irradiated with 1E15 fluence is increased by 0.75% and 1.22%, respectively. The expanded lattice parameter in irradiated films underlines the fact that displacement damage by nuclear stopping (elastic scattering between energetic C ions and Mg/B atoms) is more prominent than the damage by electronic stopping (interaction between the energetic C ions and electrons of MgB$_2$) for an increase of $\rho_{41K}$ and suppression of $T_c$ [23]. A sudden increase in $\rho_{41K}$ of MB800nm at $2 \times 10^{14}$ atoms cm$^{-2}$ (2E14) implies that the implanted C ions create substantial defects due to large damage events around $R_p$, while the $T_c$ suppression is smeared out due to the proximity effects between the irradiated and non-irradiated parts of the MgB$_2$ films.

In order to study the effect of C-ion irradiation on the upper critical field ($H_{c2}$) of MB400nm and MB800nm, the electrical resistivity was measured as a function of temperature for magnetic fields applied perpendicular to the ab plane. Figures 6(a) and (b) display the $\rho(T)$ curves in a magnetic field for the pristine and C-ion irradiated MB400nm
and reaches a maximum of 11.09 T for MB400nm and 9.78 T for MB800nm at 2E14, and decreases slowly with further increasing fluence. Both the films show the largest enhancement of $H_{c2}(0)$ at the same fluence of 2E14. Even though there occurs a large improvement in $H_{c2}(0)$ by C-ion irradiation, the $T_c$ suppression in MgB$_2$ films by C-ion irradiation, as shown in figure 7(b), is relatively small and comparable with the results of irradiation of MgB$_2$ by other particles, such as neutrons, $\alpha$ particles, oxygen ions, and so on [12, 15, 17, 32]. In addition, the $T_c$ of MB800nm is insensitive to the irradiation compared to that of MB400nm, suggesting that undamaged SC layer is helpful to improve SC critical properties, such as $H_{c2}$ and critical current density ($J_c$), while minimizing $T_c$ reduction.

Figure 8 shows the evolution of XRD patterns around the (002) peak for the pristine and C-ion irradiated MB400nm and MB800nm with the fluence of 1E15, when the irradiated films were subsequently annealed for 30 min at $T_A = 200$ °C, 300 °C, 400 °C, and 500 °C. The (002) peak position of the irradiated film, which is shifted towards a lower angle than that for the pristine film, moves to a higher angle with increasing annealing temperature and is close to that of the pristine film for $T_A = 500$ °C. Figure 9 shows a plot of the $c$-axis lattice parameter of the irradiated MgB$_2$ films as a function of the annealing temperature. Unlike MB400nm, the $c$-axis lattice parameter of MB800nm does not fully regain its value to that in the pristine state after thermal annealing at 500 °C, indicating that the C ions implanted into MB800nm remain in the films as interstitial defects. These results indicate that lattice distortion by Frenkel disorder, such as interstitials and vacancies, is mainly produced by C ions traveling through the MgB$_2$ because thermal annealing for a short period (30 min) was sufficient enough to restore the expanded lattice parameter of the irradiated MgB$_2$ films to that in the pristine state [13, 34, 35].

Figures 10(a) and (b) show the temperature dependence of ZFC and FC dc magnetization ($M$) for the annealed MB400nm and MB800nm films, respectively. The SC regions of both the films degraded by irradiation are gradually recovered as the annealing temperature ($T_A$) increases. Furthermore, the kink observed in MB800nm irradiated with a fluence of 1E15 disappears by thermal annealing, underlining that it originated due to the $T_c$ difference between the irradiated and non-irradiated layers in MB800nm, as discussed in figure 2(d). It was noted that the magnetization of MB400nm did not fully recover to that of the pristine state after annealing at 500 °C, even though its $c$-axis lattice parameter was completely restored, as shown in figures 8(a) and 9. The ZFC transition of the pristine MB400nm becomes slightly broader after being annealed at 500 °C, probably due to decomposition of Mg during the thermal annealing [30], which could be the reason why the $T_c$ of irradiated films does not recover completely after thermal annealing.

The SC transition temperature ($T_c$) and upper critical field ($H_{c2}$) of the irradiated films (1E15), after annealing them at $T_A = 500$ °C, were determined from electrical resistivity measurements and compared for various fluxes of C-ion with 2E14 fluence, respectively. It will be shown that 2E14 is an optimal fluence for the enhancement of $H_{c2}$ in this study. Here, $\rho(T)$ is normalized by the normal state resistivity ($\rho_{n0}$) at each magnetic field for comparison, and $H_{c2}$ is determined as the middle point of the SC transition, as indicated by arrows in figures 6(a) and (b). $H_{c2S}$ of MB400nm and MB800nm plotted as a function of temperature for various fluxes are shown in figures 6(c) and (d), respectively, where all $H_{c2}(T)$ curves are well explained by the two-band Ginzburg–Landau (GL) theory [33]. The solid lines in figures 6(c) and (d) are representative fitting curves for $H_{c2}(T)$ of the pristine and MgB$_2$ films irradiated with a fluence of 2E14.

The upper critical field at zero Kelvin, $H_{c2}(0)$, estimated from the data in figures 6(c) and (d) is plotted as a function of fluence and $T_c$ in figures 7(a) and (b), respectively. $H_{c2}(0)$ of the pristine MB400nm and MB800nm is 6.16 and 5.85 T, respectively, which increases rapidly with increasing fluence and reaches a maximum of 11.09 T for MB400nm and 9.78 T for MB800nm at 2E14, and decreases slowly with further increasing fluence. Both the films show the largest enhancement of $H_{c2}(0)$ at the same fluence of 2E14. Even though there occurs a large improvement in $H_{c2}(0)$ by C-ion irradiation, the $T_c$ suppression in MgB$_2$ films by C-ion irradiation, as shown in figure 7(b), is relatively small and comparable with the results of irradiation of MgB$_2$ by other particles, such as neutrons, $\alpha$ particles, oxygen ions, and so on [12, 15, 17, 32]. In addition, the $T_c$ of MB800nm is insensitive to the irradiation compared to that of MB400nm, suggesting that undamaged SC layer is helpful to improve SC critical properties, such as $H_{c2}$ and critical current density ($J_c$), while minimizing $T_c$ reduction.
irradiation. The results were similar to the $M$–$T$ results presented in figure 10, and the $T_c$ of the annealed films is slightly lower than that of the pure films, as shown in figure 11(a). $T_c$s of the annealed films decrease linearly with the corrected residual resistivity ($\rho_{corr}$) which is relatively larger for the annealed films than the pristine films, and may be ascribed to amorphization in some regions or Mg decomposition during the thermal annealing procedure [30]. The fact that $H_{c2}(0)$ of the annealed films is slightly higher than that of the pristine samples, as presented in figure 11(b), may be due to the same reason as in the case of $\rho_{corr}$. The fluence of 1E15 corresponds to a concentration of $\sim 8 \times 10^{19}$ C atoms cm$^{-3}$ at $R_p$, which is much lower than the concentration of MgB$_2$, which is $\sim 1.15 \times 10^{22}$ Mg atoms cm$^{-3}$ and $\sim 2.29 \times 10^{22}$ B atoms cm$^{-3}$ [36, 37]. This relatively small fraction of C ions may be one of the possible reasons for the negligible effects of the implanted C ions as impurities in the annealed MB800nm. More detailed studies, such as multiple injections and microscopic investigation, will be needed to make a definitive conclusion on the role of implanted C ions.

Figure 12 shows the magnetic field dependence of critical current density ($J_c$) at 5 K for MB400nm and MB800nm: pristine (squares), 1E15 (circles), and annealed at 500°C (triangles). The $J_c$ was estimated from the magnetization hysteresis ($M$–$H$) loops by using Bean’s critical state model ($J_c = 30 \Delta M/r$) [5], where $\Delta M$ is the height of the $M$–$H$ loops and $r$ is the corresponding radius of the total area of the film’s surface, and the $M$–$H$ loops were measured in magnetic fields applied perpendicular to the film’s plane. A large self-field $J_c$ of pristine films and its rapid drop in magnetic fields indicate a high quality of MB400nm and MB800nm. The field performance of $J_c$ for both the irradiated films is much stronger than that in the pristine state because of additional pinning sites created by C-ion irradiation, whereas a significant reduction of the $J_c$ at zero field is due to the suppression of SC regions and $T_c$ reduction. The $J_c$ value of irradiated MB800nm is slightly larger than that of irradiated MB400nm because of the presence of a non-irradiated layer in MB800nm films. Surprisingly, the field performance of $J_c$ for both the irradiated films almost recovered to that of the

Figure 6. Temperature dependence of normalized resistivity, $\rho(T)/\rho_n$, for (a) pristine (pri.) and (b) $2 \times 10^{14}$ C atoms cm$^{-2}$ (2E14) C-ion irradiated MB400nm in magnetic fields. Upper critical field ($\mu_0 H_{c2}$) as a function of temperature for (c) MB400nm and (d) MB800nm for various fluences. Solid lines in figures 6(c) and (d) representatively show the fitting curves for the $H_{c2}(T)$ of the pristine and irradiated (2E14) MgB$_2$ films, obtained from two-band GL theory.
pristine state after thermal annealing at $T_A = 500\, ^\circ \text{C}$, which is in sharp contrast to the thermal annealing effects on $J_c(H)$ for neutron-irradiated MgB$_2$ polycrystals \cite{14}. A similar field performance of $J_c$ between the annealed MB400nm and MB800nm is consistent with the results on the $H_{c2}(0)$, as described in figure 11.

In neutron-irradiated MgB$_2$, SC properties, such as $T_c$, the $c$-axis lattice constant, and $H_{c2}(0)$, are close to the values in the pristine state after annealing at 500 °C, whereas the field performance of $J_c$ is stronger than that of the pristine state. Since $J_c$, the depinning critical current density, is largely influenced by non-SC regions, such as defects, the thermal annealing effects on $J_c(H)$ for neutron-irradiated MgB$_2$ implies that thermodynamically irreversible defects are formed in MgB$_2$ by neutron irradiation \cite{14, 38, 39}. Considering that the effects of irradiation and thermal neutron capture reactions are complex, however, it
is difficult to understand the exact nature of the defects produced by neutron irradiation. On the other hand, the modified SC properties of high-quality MgB2 thin films, such as \( T_c, H_{c2}, \) and the \( c\)-axis lattice constant, as well as the enhanced \( J_c \) by C-ion irradiation were almost fully reversed to the values in the pristine state after thermal annealing for a short period of time. These results indicate that the degraded superconductivity of MgB2 by C-ion irradiations is mainly associated with the atomic lattice distortion caused by displacement damage, which is a reversible deformation, and the recovery of the SC characteristics to the pristine state by thermal annealing could be ascribed to the recovery of its crystallinity.

4. Conclusions

In conclusion, we studied the effects of carbon-ion irradiation on the SC critical properties of two MgB2 thin films with different thicknesses: 400 and 800 nm. One film was thinner than the mean projected range \((R_p \sim 560 \text{ nm})\) of irradiated C ions and the other was thicker than \( R_p \). The SC transition temperatures \((T_c)\) of both the irradiated MB-400nm and MB-800nm films gradually decreased with increasing fluence of the C ions, and were accompanied by an expansion of the \( c\)-axis lattice parameter. In addition, the residual resistivity \((\rho_{41K})\) and upper critical field \((H_{c2})\) considerably increased after C-ion irradiation because of the defects produced by the irradiation. On the other hand, the effect of the implanted C ions on \( H_{c2} \) was not prominent compared to other irradiation effects due to the strong interaction between the irradiated and non-irradiated layers in MgB2 films. Interestingly, the degraded superconductivity and expanded lattice parameter of the irradiated MgB2 were almost regained to that in the pristine state after thermal annealing at 500 °C for 30 min. In particular, the improved field performance of the critical current density \((J_c)\) by C-ion irradiation was also restored to that in the pristine state.
underpinning the fact that the lattice distortion induced by the irradiation is the main origin of the modified SC properties of irradiated MgB$_2$ films.

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