Negative effects of crystalline-SiC doping on the critical current density in Ti-sheathed MgB$_2$(SiC)$_y$ superconducting wires

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Abstract

Ti-sheathed MgB$_2$ wires doped with nanosize crystalline-SiC up to concentration of 15wt.% SiC have been fabricated, and the effects of the SiC doping on the critical current density ($J_c$) and other superconducting properties are studied. In contrast with the previously reported results that nano-SiC doping with doping range below 16wt.% usually enhances $J_c$, particularly at higher fields, our measurements show that SiC doping decreases $J_c$ in almost the whole field range from 0 to 7.3 Tesla at all temperatures. Furthermore, it is found that the degradation of $J_c$ becomes stronger at higher SiC doping level, which is also in sharp contrast with the reported results that $J_c$ is usually optimized at doping level near 10wt.% SiC. Our results indicate that these negative effects on $J_c$ could be attributed to the absence of significant effective pinning centers (mainly Mg$_2$Si) due to the high chemical stability of the crystalline SiC particles.

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1. Introduction

Following the first report of Dou et al. [1] on the dramatic enhancement of critical current density ($J_c$) in MgB$_2$ superconductors by nano-SiC doping, extensive studies have been carried out on nano-SiC doped MgB$_2$ wires/tapes/pellets prepared with different SiC grain sizes (5-300 nm), doping levels, sintering conditions, and precursor powders [2-14]. Particularly, the dependence of $J_c$ on SiC doping level is an important issue and was studied by a number of groups in the last few years. The initial study of Dou et al [2] on some Fe-sheathed MgB$_2$(SiC)$_y$ samples, with $y$ being the weight percentage of MgB$_2$ and the size of SiC in the range between 10 nm and 100 nm, indicated that $J_c$ increases dramatically (e.g., up to a maximum factor of $\sim 30$) with $y$ increasing from 0 to 10wt.%. However, $J_c$ was found to be much lower than that of the un-doped sample at higher doping levels of 20wt.% and 30wt.%. The enhancement of $J_c$ due to nano-SiC doping in the doping level range $0 < y < 0.16$ was later reported by some other groups [4-14], showing that SiC doping with doping level up to $y = 16$wt.% usually enhances $J_c$, particularly at higher fields (e.g., $> 2-4$ Tesla at 5 K). In spite of these studies, it is still unclear as to at what doping level and under what synthesizing/fabricating conditions (e.g. sintering temperature and time etc.) SiC doping would optimize or degrade $J_c$. For example, while the study of Soltanian et al [5] suggests that $J_c$ is optimized at SiC doping level $y = 10$wt.% and has a small change in the range of $y$ between 8wt.% and 15wt %, the result of Pachla et al [6], on the other hand, shows that $J_c$ decreases substantially with the increase of $y$ from 5.2wt.% to 15.9wt.%.

Recently, we have successfully fabricated Ti-sheathed, un-doped MgB$_2$ wires with high $J_c$ and demonstrated that the performance of Ti-sheath on $J_c$ is comparable or even better than the Fe-sheath [15]. To further increase $J_c$ for future applications of Ti-sheathed MgB$_2$ wires in lightweight superconducting magnets, we fabricated Ti-sheathed MgB$_2$ wires doped with crystalline nano-SiC (20 nm) at different doping levels. We would like to know if $J_c$ in these wires could be enhanced substantially and the results obtained could help us further understand the dependence of $J_c$ on SiC doping level. However, to our surprise, we found that for SiC doping level up to 15wt.%, the nano-SiC doping actually decreases $J_c$ in these Ti-sheathed MgB$_2$ wires substantially, virtually at all temperatures and at almost all magnetic fields up to 7.3 Tesla (T). Such negative effects of SiC (in either crystalline or amorphous form) doping on $J_c$ in a wide range of fields has not yet been reported for samples with doping levels less than 16wt.%SiC. This paper presents a detailed analysis and discussion on this newly observed phenomenon in Ti-sheathed, nano-SiC doped MgB$_2$ superconducting wires.

2. Experimental details
Ti-sheathed, SiC-doped monocore MgB$_2$ wires were fabricated using the standard *in situ* powder-in-tube (PIT) method with exactly the same fabricating conditions as described previously on the fabrication of the un-doped MgB$_2$ wires [15]. The SiC powder was well mixed with the milled Mg + 2B mixture which has an average particle size of about 1 $\mu$m [15]. The SiC powder with a nominal size of 20 nm was purchased from Alfa Aesar. The MgB$_2$(SiC)$_y$ wires were prepared with the SiC doping levels at $y =$ 5wt.%, 10wt.%, and 15wt.%. Wire sections of about six inches long were cut from the as-drawn wires and sintered in a tube furnace in flowing high purity argon with the following schedule: the temperature was ramped up from room temperature to 800 °C at a rate of 300 °C/hour, kept at 800 °C for 30 minutes, and then cooled down to room temperature at a rate of 100 °C/hour. The cross-sectional areas of the wires are about 1 mm × 1 mm and the MgB$_2$ cores have cross-sectional areas of about 0.34 mm × 0.37 mm.

The size and shape of the commercial SiC particles were measured by transmission electron microscopy (TEM). The impurities, compositions, and microstructures of the SiC-doped MgB$_2$ wires were studied by TEM, x-ray energy dispersive spectroscopy (EDS), and scanning electron microscopy (SEM). The TEM work was carried out using a JEOL 2010 microscope at a working voltage of 200 kV and the SEM work was performed using a JEOL JSM-6330F Field Emission Scanning Electron Microscope. In the TEM measurement, the SiC powders or the slightly ground core powders of the SiC doped MgB$_2$ wires were mixed with ethanol solution in about 0.05wt% concentration, then a small drop of this suspension was transferred to the top surface of a carbon-film supported Cu grid until the suspension was dried. For accurate size measurement, all of the TEM imaging magnifications were calibrated: for magnifications at or above $\times$ 100,000 [16], standards of 6$H$ SiC lattice (hexagonal structure with $a =$ 0.3087 nm and $c =$ 1.5117 nm) fringes were used for the calibration; for lower magnifications at or below $\times$ 80,000, cross-line grating replica was used. The EDS microanalysis was performed using an Oxford Instruments EDS detector with ultra thin window, which enables the elemental qualitative and quantitative analyses from boron ($Z \geq$5). The data acquisition, qualitative and quantitative analyses were done using an Oxford INCA TEM system, and the x-ray elemental mapping was done in the INCA Semi-STEM (scanning TEM) mode.

The x-ray diffraction (XRD) patterns were obtained using a Rigaku x-ray 2005 diffractometer with Cu K$_\alpha$ radiation. To get the core materials out of the Ti-sheathed wires for making XRD slide samples, the wires were rolled into thin tapes, cut by scissors, and then finally peeled off with a knife. The core material was then ground into fine powder for XRD measurement. The temperature ($T$) dependent resistivity, $\rho(T)$, was measured by a standard four-probe dc technique. The temperature dependent magnetization, $M(T)$, was measured in both zero-field-cooled (ZFC) and field cooled (FC) modes using a Magnetic Properties Measurement System (MPMS) magnetometer from Quantum Design. Except for the
30 K hysteresis half loop of the sample with $y = 10\text{wt.\% SiC}$, which was measured using the MPMS, all of the other hysteresis loops were measured using an Oxford Instrument’s Maglab 9-Tesla Vibrating Sample Magnetometer (VSM) with a field ramping step of $\Delta H \approx 33 \text{ Oe}$. In the $M(T)$ or $M(H)$ measurements, the longitudinal axis of each wire sample was oriented along the direction of the applied magnetic field.

3. Results and Discussion

Fig. 1 shows a TEM image for the commercial SiC particles. The particles are seen in multi-angular shape, with a large size variation. The average size is measured as 20.2 nm with a standard deviation of 9.7 nm, based on the measurements of 179 particles. This value of average size is almost identical to the nominal size (20 nm) provided by Alfa Aesar. The XRD pattern shown in Fig. 2 (top) indicates that these SiC nanoparticles are crystallized in the cubic $\beta$-SiC phase with space group $F\bar{4}3m$ (No. 216) and lattice constant $a = 4.362(4)$ Å.

Fig. 2 shows the XRD patterns for the core material of the MgB$_2$(SiC)$_y$ wire samples, with the SiC doping level (in weight percentage of MgB$_2$) $y = 0, 0.05, 0.10,$ and $0.15$. For comparison, the patterns of

Figure 1. The TEM image of the commercial SiC nanoparticles used in this study.
three reference compounds, SiC (20 nm), MgB$_2$ powder (-325 mesh), and Ti powder (-325 mesh), are also shown in Fig.2. All of the patterns are normalized to the intensity of the strongest peak in each pattern. The major peaks for these wire samples can be indexed with the MgB$_2$ hexagonal structure, indicating that the core materials in these wires are primarily MgB$_2$ phase. The SiC (111) peak [PDF # 29-1129] located at $2\theta \approx 35.6^\circ$ is seen in each pattern of the three SiC doped samples but not in the un-doped sample ($y = 0$) indicating that the SiC nanoparticles did not react completely with other elements during

![XRD patterns](image)

**Figure 2.** The XRD patterns for the core materials of the Ti-sheathed, SiC-doped MgB$_2$ wires. For comparison, the XRD patterns of three reference compounds, SiC powder (20 nm), MgB$_2$ powder (-325 mesh), and Ti powder (-325 mesh), are shown.
the 30 minutes sintering at 800 °C. The intensity of this peak is measured, and it is found that the peak intensity is proportional to the doping level. The two weak impurity peaks located at 2θ = 38.4° and 40.1° are due to the (002) and (101) peaks of the α-phase Ti [PDF # 44-1294]. As discussed previously and confirmed by EDS result in Ref. 15 and below, the Ti impurities were extrinsic, which were scraped away from the inner wall of the Ti sheath and had fallen into the powder of the core material during the preparation of the XRD sample slides. Note that since the strongest peak of the cubic Mg2Si phase [PDF #35-0773] is also located at 2θ ≈ 40.1°, the intensity for the Ti (101) peak located at 2θ = 40.1° could have contribution from possible Mg2Si impurities, if any, due to the peak overlapping. However, our XRD data cannot determine definitely the existence of the Mg2Si impurity in the samples. The broad peak located at 2θ = 62.33° can be attributed to the cubic MgO phase [2, 4, 10]. The main source of the oxygen could be from the air trapped in the cores during the short-time crimping sealing of the end of the Ti tube (with Mg+2B mixture packed in) in air. Our quantitative analysis shows that the content of the MgO in the sample is less than 5%, which is consistent with the result reported by Dou et al [1] for their SiC doped MgB2 samples.

In Fig. 3, we show the TEM images and EDS spectra for the powder of the core material of the MgB2 wire with 10wt.% SiC doping. Fig. 3(a) shows an overview image in lower magnification, and its bottom part is magnified in Fig. 3(b). The large grains are identified mainly in the MgB2 phase, as shown by the EDS analysis in Fig. 3(c). Quantitative analysis shows that the sample contains 30.1at.% Mg and 69.1at.% B. Due to the overlapping of B with C peaks, the quantitative analysis of B by EDS is not expected to have high accuracy as compared with other elements. Some particles of irregular shape are identified as amorphous SiO2, an example is marked in Fig. 3(a) and its EDS spectrum is shown in Fig. 3(d). Its composition is found to be 32.1 at.% Si, 67.9 at.% O. The SiO2 could be formed by the reaction between O2 and Si resulted from the decomposition of SiC during the sintering process. Since the SiO2 particles are in amorphous form, they could not be detected by the XRD. Note that this is the first report on the formation of SiO2 in SiC doped MgB2 materials. Fig. 3(b) shows that on the surface of MgB2 grains, there is a distribution of two kinds of particles: the larger ones (darker) with size 17.1± 3.5 nm are SiC, and the well dispersed smaller particles with size less than 10 nm are MgO particles. Because these two kinds of particles are well dispersed, the EDS spectrum shows both MgO and SiC in Fig. (e). The quantified composition is 49.1 at.% C, 9.2 at.% O, 7.2 at.% Mg, and 34.5 at.% Si, which is basically consistent with the stoichiometry of MgO and SiC. The carbon content is higher than the stoichiometric composition due to the contribution from the supported carbon film. The slightly smaller size (17.1 nm) of the SiC particles than that (20 nm) of the initial SiC particles means that the reaction between SiC and other elements (such as O, Mg) is light. The EDS does not detect the existence of the Mg2Si particles, indicating that either Mg2Si impurities were not formed in the SiC doped samples during the sintering
process or their concentration is too low to be detected. This EDS result is consistent with the XRD result discussed above and can be explained by the light reaction of SiC and the formation of SiO₂ which reduces the amount of Si available for forming Mg₂Si. The EDS spectra in Fig. 3 do not show any Ti lines that should be located at 4.508 keV for Ti Kα and 4.931 keV for Ti Kβ lines, respectively. This confirms that the Ti impurities detected by XRD are extrinsic, i.e., they were not formed in the cores of the wires during the sintering process.

The above XRD/TEM/EDS results show some distinct differences between our SiC-doped MgB₂ samples and those previously by Dou et al [1-2] and Soltanian et al [5]. (1) For our samples, the reaction between the doped SiC particles and Mg +2B mixture is very light, as evidenced by the small size change for the SiC particles before and after sintering, whereas for their samples, SiC reacted fully and no SiC peak was observed in the XRD patterns [1, 2, 5]. (2) For our samples, no indication of the formation of significant or detectable amounts of Mg₂Si nanoparticles was observed. For their samples, prominent intensity of Mg₂Si peak was observed in the XRD patterns. (3) Our TEM result shows that the SiC particles are mostly located on the surface boundaries of the MgB₂ grains, whereas their TEM results

Figure 3. (a) is the TEM image of the powder of the 20 nm SiC doped MgB₂ core with doping level of 10wt.% SiC. (b) shows the magnified bottom portion of the image in (a). (c) – (e) are the EDS spectra taken at different sites of the sample, as labeled with letters d, c and e, respectively, in (a) and (b).
[5] indicated that most of the SiC particles are embedded inside the MgB₂ grains. (4) The formation of amorphous SiO₂ impurities was observed for our samples, whereas the existence of SiO₂ impurities in their samples was not mentioned and thus is unclear. Observations (2) and (3) may be consequences of observation (1), which could possibly result from the difference in the sintering time and the crystal structure of the SiC nanoparticles. For example, our samples were sintered at 800 °C for 30 min. whereas the samples reported by Dou et al [2] were sintered at 950 °C for 3 h. Higher sintering temperature and longer sintering time could make the reaction between SiC and other elements more completed and thus decrease the concentration of the SiC particles. In the case of the samples reported by Soltanian et al [5], the size of the SiC nanoparticles used is the same as ours (20 nm) and the average sintering temperature they used (700 °C to 900 °C) is similar to ours (800 °C). However, their sintering time (1 h) is also longer than ours (0.5 h). In addition, the SiC nanoparticles they used were in amorphous form, which could react with other elements in a different way from our crystallized β-phase SiC. Such different effect caused by different structures has been observed from some carbon doped MgB₂ samples for which carbon with different crystal structures could react with Mg +2B precursor forming different impurities [17]. Thus, due to light reaction of SiC in our samples, most of the doped SiC nanoparticles after sintering would be located at the grain boundaries of the formed MgB₂ grains instead of being located inside them. On the other hand, for the samples reported by Dou et al [1, 2] and Soltanian et al [5], since the SiC particles were fully or almost fully reacted with the Mg + 2B, the very fine formed Mg₂Si nanoparticles and the very fine, un-reacted remainders of the SiC particles can be embedded inside the MgB₂ grains.

Shown in Fig. 4 are the SEM images for the cores of the four wire samples. The surfaces of the samples were polished. These images show that large amounts of holes/voids exist in the samples. Most of them are about 1-2 μm in diameter which is close to the size of the Mg particles in the milled Mg + 2B powder precursor. These voids could be produced by the volume reduction in the Mg + 2B → MgB₂ reaction, it also could be partially attributed to the evaporation of the Mg particles during the sintering of the wires. It appears from the SEM images that the density of the hole/void decreases with the increase of the doping level. For example, the hole density for the 15wt.% SiC sample shown in Fig. 4(d) is much less than that of the un-doped sample shown in Fig. 4(a). Since in the process of mixing and packing the Mg, B, and SiC powders, the spaces between the bigger Mg +2B particles (≈ 1 μm in size) were filled by the much smaller SiC nanoparticles (20 nm average size), thus higher doping level of SiC could result in a higher filling or packing factor.

Fig. 5 shows the field dependent magnetic $J_c(H)$ curves for the Ti-sheathed MgB₂(SiC)ₙ wire samples. The cross sections of the MgB₂ cores are $a \times b \approx 0.34 \text{ mm} \times 0.37 \text{ mm} = 0.126 \text{ mm}^2$ and the lengths of the wires range from 7.5 mm to 10 mm. The magnetic $J_c$ of the samples was calculated by
formula $J_c = 20\Delta M / [a(1-a/3b)]$ [15, 18, 19] from the Bean critical state model [20], where $\Delta M$ is the difference between the upper and lower branches of the hysteresis loops. Our field dependent magnetization measurement on the Ti-sheath of the wires indicates that the Ti-sheath is paramagnetic [21, 21] and reversible with field, and thus the magnetization background due to the Ti-sheath should have no any contribution to the calculated $J_c$. The four $J_c$ curves measured at 5 K (Fig. 5(a)) clearly show that SiC doping depresses $J_c$ substantially in the whole field range from 0 to 7.5 T, and the effect of depression is stronger when the SiC doping level $y$ is higher. At 15wt.% SiC, $J_c$ is decreased to only 15%-25% of the value of the un-doped ($y = 0$) sample. For example, the $J_c$ at 2 T for the un-doped sample is about $1.7 \times 10^5$ A/cm$^2$, it decreases to only $2.5 \times 10^4$ A/cm$^2$ at $y = 0.15$. There is no any crossover between the $J_c$ curve of the un-doped and doped samples, indicating that SiC doping does not enhance $J_c$, not even in the high field region. It is seen from the slopes of the $J_c$ curves that for the samples of $y = 0.05$ and 0.10, the drop of $J_c$ with increasing field is slower than that of the un-doped sample. This is the only positive effect of SiC doping observed from Fig. 5. However, for the sample of $y = 0.15$, the slope of the $J_c(H)$ curve is similar to that of the un-doped sample, indicating that at this higher doping level, SiC doping does not

Figure 4. SEM images of the cores of the Ti-sheathed MgB$_2$ wires which are (a) un-doped, (b) doped with 5wt.% SiC, (c) doped with 10wt.% SiC, and (d) doped with 15wt.% SiC. The surfaces of the cores were polished before taking the images. These SEM images show that a large amount of holes/voids exist in the cores of the wires.
slow down the dropping of $J_c$. The $J_c$ curves measured at 20 K (Fig. 5(b)) show similar effects of SiC doping to what shown by the 5K $J_c$ curves, except that the 20 K $J_c$ curve of the $y = 0.05$ sample has a crossover with the 20 K $J_c$ curve of the $y = 0$ sample at about 3.8 T. For the $J_c$ curves measured at 30 K (Fig. 5(c)), the variation of the slope indicates that at 30 K, the SiC doping actually causes the $J_c$ drop faster at higher SiC doping level in the whole doping range of $0 \leq y \leq 0.15$.

Such a substantial depression of $J_c$ in a wide range of field and at all temperatures, caused by nano-SiC doping with doping level ranging from 0 to 15wt.% SiC, is in sharp contrast with earlier results [1-
that SiC doping in this range usually enhances $J_c$, particularly in the high field region. Some researchers [1, 2, 5] have attributed such enhancement of $J_c$ to the nano-SiC and Mg$_2$Si inclusions embedded inside MgB$_2$ grains and believed that these nano-inclusions function as effective pinning centers. If such explanation is valid, then the observed negative effects of SiC doping on $J_c$ for our samples could be explained by the absence of significant amounts of SiC and Mg$_2$Si nanoparticles distributed inside MgB$_2$ grains. For our samples, as discussed above on the XRD/TEM/EDS results, no detectable amount of Mg$_2$Si was formed due to the very light reaction of the β-phase SiC nanoparticles with other elements, and the un-reacted SiC nanoparticles are distributed mainly on the surface boundaries of the MgB$_2$ grains instead of being embedded inside the MgB$_2$ grains. We believe that the suppression of $J_c$ by SiC doping in our samples is a consequence of the competition between two opposite effects: on one hand, there could still be small amounts of very fine SiC and formed impurities (possible Mg$_2$Si) embedded inside the MgB$_2$ grains as effective pinning centers, which can enhance $J_c$. On the other hand, the majority of the un-reacted SiC and the impurities (MgO and SiO$_2$) are distributed around the boundaries of the MgB$_2$ grains serving as weak links, resulting in the substantial decrease of $J_c$. Thus, more SiC nanoparticles located at the grain boundaries mean more degradation of $J_c$. This explains why $J_c$ decreases with the increase of the SiC doping level. This explanation for the negative effect of SiC doping on $J_c$ suggests that unlike the SiC nanoparticles located inside the MgB$_2$ grains, the SiC nanoparticles located at the grain boundaries of the MgB$_2$ grains may not act as effective pinning centers for enhancing $J_c$.

Shown in Fig. 6 are the dc magnetization $M(T)$ curves for the four samples, measured at 20 Oe in both zero-field-cooled (ZFC) and field-cooled (FC) modes. It is observed from the ZFC curves that the onset transition temperature, $T_{c, on}$, defined as the temperature at which the susceptibility starts to drop, is $\sim 35.9$ K. It is observed that the $T_{c, on}$ is almost unchanged with the doping level $y$. In contrast, the midpoint transition temperature, $T_{c, mid}$, defined as the temperature at the half drop of $M(T)$, decreases continuously with increasing $y$ from 35.4 K at $y = 0$ to 30.8 K at $y = 0.15$. The width of the transition, $\Delta T$, defined by the difference between the temperatures at 10% and 90% of the full drop of $M(T)$, increases continuously with increasing $y$ from $\Delta T = 2.2$ K at $y = 0$ to $\Delta T = 9.8$ K at $y = 0.15$. The decrease of the $T_{c, mid}$ and increase of the width $\Delta T$ with increasing $y$ were also observed by Dou et al [2] for their MgB$_2$(SiC)$_y$ samples. However, compared with their observed variations in $T_{c, mid}$ and $\Delta T$, which are about 1.7 K and 1.03 K, respectively, with doping level up to $y = 0.2$, the changes in $T_{c, mid}$ ($\approx 4.6$ K) and $\Delta T$ ($\approx 7.6$ K) for our samples with $y$ up to 0.15 are much larger. These large variations in $T_{c, mid}$ and $\Delta T$ suggest that the co-substitution of B by both Si and C might not occur in our samples.

The inset of Fig. 6 shows the temperature dependent electrical resistivity, $\rho(T)$, for three MgB$_2$(SiC)$_y$
samples with $y = 0$, 0.10, and 0.15. The $T_{\text{con}}$ determined from these $\rho(T)$ curves is 35.9 K for all of these three samples, which matches well with the values determined from the $M(T)$ curves. The un-doped sample ($y = 0$) has the sharpest transition with $\Delta T \approx 0.6$ K. With SiC doping, $\Delta T$ increases to about 1 K at both $y = 0.10$ and $y = 0.15$, reflecting the distribution of the SiC particles at the MgB$_2$ grain boundaries [6]. The transition width $\Delta T$ (0.6-1.1 K) is much narrower than that (2.2-9.8 K) determined from the $M(T)$ curves. Different from the monotonic variation of the $T_{\text{c, mid}}$ observed from the $M(T)$ curves, the $T_{\text{c, mid}}$ determined from the $\rho(T)$ curves first decreases from 35.6 K at $y = 0$ to 34.8 K at $y = 0.10$, and then

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**Figure 6.** Temperature dependent dc magnetization, measured in both ZFC and FC modes in a field of 20 Oe and between 5 and 50 K, for the Ti-sheathed MgB$_2$(SiC)$_y$ wires. In the figure, only the sections of the curves in the temperature range between 5 K and 40 K are shown. The inset shows the temperature dependent electrical resistivity curves in a temperature range between 33 and 42 K.
increases to 35.4 K at \( y =0.15 \). This minimum \( T_{c,\text{mid}} \), occurring at \( y \approx 10\text{wt.\% SiC} \), was also observed by Pachla et al. [6] from their temperature dependent resistance curves, and was explained by the competition of two opposite effects on the \( T_{c,\text{mid}} \): (i) the negative effect due to the SiC-doping induced deterioration of the MgB\(_2\) grain connectivity and (ii) the positive effect due to the SiC-doping caused reduction of the pressure inside the MgB\(_2\) core.

4. Conclusions

We have fabricated and characterized mono-core Ti-sheathed MgB\(_2\) wires doped with crystalline SiC nanoparticles of average size 20 nm and concentrations up to 15 wt.\% SiC. The wires were sintered at 800 °C for 30 minutes. In sharp contrast with the previously reported results that amorphous nano-SiC doping in this doping range usually enhances \( J_c \), at least at higher fields, our measurements show that crystalline-SiC doping decreases \( J_c \) in almost the whole field range from 0 to 7.3 T and at all temperatures. It is found that the degradation of \( J_c \) becomes stronger when the SiC doping level is higher. Our XRD/TEM/EDS analysis indicates that the origins of these negative effects on \( J_c \) could be attributed to the absence of significant pinning centers (mainly very fine SiC and Mg\(_2\)Si particles) embedded inside the MgB\(_2\) grains, which is the consequences of (1) the very light reaction between the \( \beta \)-phase SiC nanoparticles (due to its high chemical stability at the sintering temperature) and the Mg\(+2\)B precursor powder and (2) the distribution of the SiC nanoparticles mainly at the MgB\(_2\) grain boundaries. This study suggests that the formation of significant content of Mg\(_2\)Si by reaction of SiC with Mg might be a necessary condition for the enhancement of \( J_c \) in SiC-doped MgB\(_2\) wires.

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