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S X. Dou

*University of Wollongong, shi@uow.edu.au*

V. Braccini

*University of Wisconsin-Madison, USA*

Saeid Soltanian

*University of Wollongong, saeid@uow.edu.au*

R. Klie

*Brookhaven National Laboratory, New York, USA*

Y. Zhu

*Brookhaven National Laboratory, New York, USA*

*See next page for additional authors*

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**Authors**

S X. Dou, V. Braccini, Saeid Soltanian, R. Klie, Y. Zhu, S. Li, Xiaolin Wang, and D. LARBALÉSTIER

## Nanoscale-SiC doping for enhancing $J_c$ and $H_{c2}$ in superconducting $MgB_2$

S. X. Dou<sup>a)</sup>

*Institute for Superconducting and Electronic Materials, University of Wollongong, Wollongong, New South Wales 2522, Australia*

V. Braccini

*Applied Superconductivity Center, University of Wisconsin–Madison, Wisconsin 53706 and Applied Conductivity Centre, University of Wisconsin–Madison, Madison, Wisconsin 53706*

S. Soltanian

*Institute for Superconducting and Electronic Materials, University of Wollongong, Wollongong, New South Wales 2522, Australia*

R. Klie and Y. Zhu

*Brookhaven National Laboratory, Utpn, New York 11973*

S. Li

*Advanced Materials Research Center, Nanyang Technological University, 639798, Singapore*

X. L. Wang

*Institute for Superconducting and Electronic Materials, University of Wollongong, Wollongong, New South Wales 2522, Australia*

D. Larbalestier

*Applied Superconductivity Center, University of Wisconsin–Madison, Wisconsin 53706 and Applied Conductivity Centre, University of Wisconsin–Madison, Madison, Wisconsin 53706*

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The effect of nanoscale-SiC doping of  $MgB_2$  was investigated in comparison with undoped, clean-limit, and Mg-vapor-exposed samples using transport and magnetic measurements. It was found that there are two distinguishable but related mechanisms that control the critical current-density-field  $J_c(H)$  behavior: increase of upper critical field  $H_{c2}$  and improvement of flux pinning. There is a clear correlation between the critical temperature  $T_c$ , the resistivity  $\rho$ , the residual resistivity ratio  $RRR=R(300\text{ K})/R(40\text{ K})$ , the irreversibility field  $H^*$ , and the alloying state in the samples. The  $H_{c2}$  is about the same within the measured field range for both the Mg-vapor-treated and the SiC-doped samples. However, the  $J_c(H)$  for the latter is higher than the former in a high-field regime by an order of magnitude. Mg vapor treatment induced intrinsic scattering and contributed to an increase in  $H_{c2}$ . SiC doping, on the other hand, introduced many nanoscale precipitates and disorder at B and Mg sites, provoking an increase of  $\rho(40\text{ K})$  from  $1\ \mu\Omega\text{ cm}$  ( $RRR=15$ ) for the clean-limit sample to  $300\ \mu\Omega\text{ cm}$  ( $RRR=1.75$ ) for the SiC-doped sample, leading to significant enhancement of both  $H_{c2}$  and  $H^*$  with only a minor effect on  $T_c$ . Electron energy-loss spectroscopy and transmission electron microscope analysis revealed impurity phases:  $Mg_2Si$ ,  $MgO$ ,  $MgB_4$ ,  $BO_x$ ,  $Si_xB_yO_z$ , and BC at a scale below 10 nm and an extensive domain structure of 2–4-nm domains in the doped sample, which serve as strong pinning centers. © 2004 American Institute of Physics. [DOI: 10.1063/1.1814415]

### INTRODUCTION

The critical current density  $J_c$  in  $MgB_2$  has been a central topic for extensive research efforts since superconductivity in this compound was discovered.<sup>1</sup> A number of techniques have been developed and employed to improve the  $J_c$  performance in magnetic fields. By using nanoparticle SiC doping of  $MgB_2$ , we have achieved a  $J_c$  enhancement in high fields of more than one order of magnitude, with only a slight reduction in  $T_c$ .<sup>2</sup> It was proposed that a high density of nanoinclusions and a possible substitution of SiC for B in  $MgB_2$  was responsible for enhancing  $J_c(H)$  over a wide range of temperatures.

Recently, using high-field transport measurements, Gurevich *et al.* have reported the achievement of record-high upper critical fields  $H_{c2}$  for high-resistivity films and untextured bulk polycrystals.<sup>3</sup> They found that enhancements to the resistivity have a strong influence on  $H_{c2}$ . The observed remarkable  $H_{c2}$  enhancement to almost 50 T is a consequence of the two-gap superconductivity of  $MgB_2$ ,<sup>3</sup> which offers special opportunities for further  $H_{c2}$  increase by tuning the impurity scattering. Nanoscale-SiC doping introduces a large degree of alloying and greatly raises resistivity, too. Thus, we expected that transport measurements on SiC-doped samples would provide additional useful information for understanding the pinning mechanisms and  $H_{c2}$  behavior of alloyed  $MgB_2$ . In this paper, we report on such transport and magnetic measurement evaluations in combination with

<sup>a)</sup>Electronic mail: shi\_dou@uow.edu.au

transmission electron microscope (TEM) observations on the nanoscale-SiC-doped  $\text{MgB}_2$ . A set of four samples ranging from the clean limit to very dirty state have wild variation of normal-state resistivity from 1 to  $300 \mu\Omega \text{ cm}$  and significantly different electromagnetic properties which allow us to understand the mechanisms behind the enhancement of  $J_c(H)$ .

## EXPERIMENTAL DETAILS

$\text{MgB}_2$  pellet samples were prepared by an *in situ* reaction method, described in detail previously.<sup>2</sup> Powders of magnesium (99%) and amorphous boron (99%) were well mixed with SiC nanoparticle powder (size: 10–100 nm) with the atomic ratio of  $\text{MgB}_2$  and with 10 wt % (sample B) and 0 wt % (sample A) SiC addition. Pellets 10 mm in diameter and 2 mm in thickness made under a uniaxial pressure were sealed in an Fe tube and then heated at  $800^\circ\text{C}$  for 30 min in flowing high-purity Ar, followed by furnace cooling to room temperature. These two Wollongong samples were compared to two Madison samples, one being the clean limit (sample C) and the second being the same sample exposed to Mg vapor (sample D), as described in detail elsewhere.<sup>4</sup>

The resistivity versus temperature curves,  $\rho(T)$ , were measured in magnetic fields up to 9 T by a four-probe method at a current density of about  $1 \text{ A/cm}^2$  using a 9-T Physical Property Measurement System (Quantum Design). From the resistivity curves, we defined the upper critical field as  $R(H_{c2})=0.9 R(T_c)$ . Magnetization was measured from 5 to 30 K using an Oxford 14-T vibrating sample magnetometer (VSM). Bar-shaped samples of about the same size were cut from the as-sintered pellets to minimize size-dependent effects.<sup>5</sup> Magnetic  $J_c$  values were determined from the magnetization hysteresis loops using the appropriate critical state model.<sup>6</sup> An empirical magnetic irreversibility line,  $H_M^*$ , was defined as the field at which  $J_c$  falls to  $100 \text{ A/cm}^2$ .<sup>4</sup> High-resolution transmission electron microscopy was employed to characterize the morphology of the samples. Electron energy-loss spectroscopy<sup>7</sup> (EELS) was obtained using a JEOL-3000F field-emission scanning transmission electron microscope TEM, equipped with a Schottky field-emission source operated at 300 keV.

## RESULTS

Figure 1 shows the resistivity curves,  $\rho(T)$ , up to 9 T for the undoped (a) and the SiC-doped (b) samples. The onset  $T_c$  of the undoped sample was 37.5 K. For the 10 wt % SiC-doped sample,  $T_c$  decreased only by 0.6 K. By contrast,  $T_c$  is depressed to about 22 K for C-doped  $\text{MgB}_2$  with a nominal stoichiometry of  $\text{Mg}(\text{B}_{0.8}\text{C}_{0.2})_2$  synthesized from Mg and  $\text{B}_4\text{C}$ .<sup>8</sup> This indicates that the proportion of C added SiC to substitute for B in the lattice is small compared to the pure C substitution case. The majority of SiC ends up with various impurity phases at nanoscale, as evidenced by the EELS analyses. It is noted that the  $\rho(T)$  curves for the doped sample shifted with increasing field much more slowly than the undoped one, as is shown explicitly in Fig. 2, where the  $H_{c2}$  obtained from the 90% values of the resistive transitions from Fig. 1 are shown. It is also noted that the  $\rho(T)$  curve for

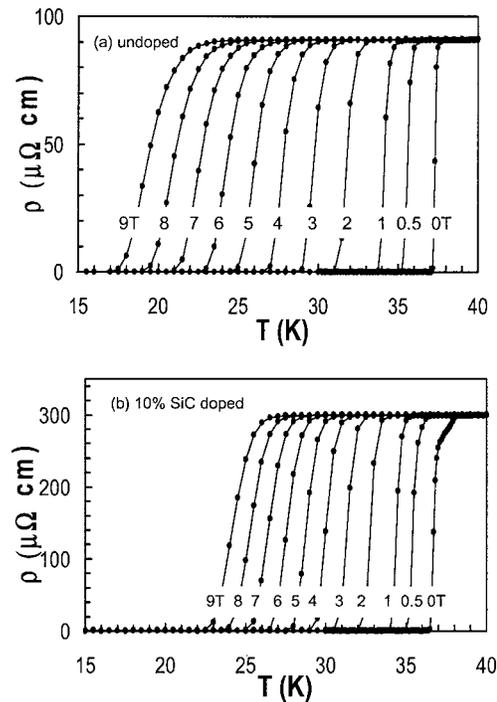


FIG. 1. The resistivity vs temperature in fields up to 9 T for the undoped (a) and SiC-doped (b) samples.

the doped sample in self-field showed a special feature of two-step transition. This is due to the inhomogeneity of this sample because the 10% SiC addition resulted in a number of impurity phases coexisting with  $\text{MgB}_2$ . This will be further confirmed in EELS analyses in a later section. A further important point is that the nominal resistivities of the two samples are very different,  $\rho(40 \text{ K})$  being  $90 \mu\Omega \text{ cm}$  for the undoped sample and  $300 \mu\Omega \text{ cm}$  for the doped sample. We consider that the 90% transition approximates  $H_{c2}$ . Figure 2 also includes the same data taken on the clean-limit [ $\rho(40 \text{ K})=1 \mu\Omega \text{ cm}$ ] and Mg-exposed sample [ $\rho(40 \text{ K})=18 \mu\Omega \text{ cm}$ ] of Braccini *et al.*<sup>4</sup> It is interesting to note that the Mg-exposed sample has the highest  $H_{c2}$ , then the SiC doped, undoped, and the clean limit. Figure 3 shows a typical example of the half  $M-H$  loop at 20 K for the undoped sample (A) and 10% SiC-doped- $\text{MgB}_2$  sample (B). It is clear that the closing field of the  $M-H$  loop for the sample B is

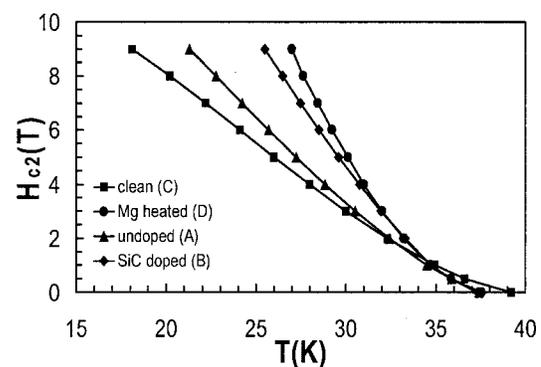


FIG. 2. The 90% of the resistive transition (upper critical field) as a function of the temperature for the undoped (A), the 10 wt % SiC-doped (B), the clean-limit, (C) and the Mg-vapor-treated (D) samples.

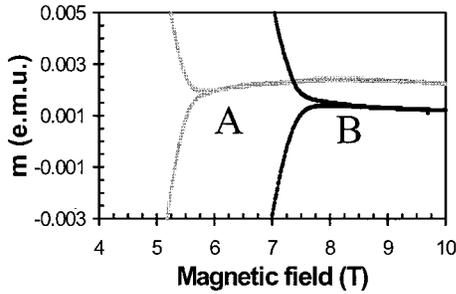


FIG. 3. The M–H loop at 20 K for the undoped sample (A) and the SiC-doped sample (B).

2 T higher than that for the sample A. The magnetic  $J_c$  for all the samples was calculated from the  $M-H$  loops at 4.2, 10, 20, and 30 K. Figure 4 shows the magnetic  $J_c$  vs  $H$  for the four samples at 4.2 and 20 K. Consistent with its higher  $H_{c2}$ , the doped sample shows a smaller dependence of  $J_c$  on magnetic field at all temperatures. At 4.2 K and low field, both the SiC-doped and undoped samples attain about  $10^5$  A/cm<sup>2</sup> while falling to 100 A/cm<sup>2</sup> at 7.4 and 5.6 T at 20 K, respectively. The  $J_c$  values for the Wollongong samples (A and B) are much higher than for the two Madison samples (C and D). At 20 K, the 10 wt % SiC-doped sample achieved  $10^5$  A/cm<sup>2</sup> at 3 T, comparable to that of state-of-the-art Ag/Bi-2223 tapes, and an order of magnitude higher than recent state-of-the-art Fe/MgB<sub>2</sub> tapes.<sup>9</sup> These results significantly strengthen the position of MgB<sub>2</sub> as a competitor for both low- and high-temperature superconductors.

The irreversibility fields ( $H_M^*$ ), derived from the fields at which the magnetic hysteresis loops obtained with the VSM indicate that  $J_c=100$  A/cm<sup>2</sup>, are shown in Fig. 5. Doping with SiC significantly improved  $H_M^*$ . Here, in contrast to the  $H_{c2}$ , we note that the SiC-doped sample has the best  $H_M^*$  at all temperatures while the Mg-exposed sample has  $H_M^*$  even lower than the undoped at high temperatures but crosses over the undoped at low temperatures. For example,  $H_M^*$  for the

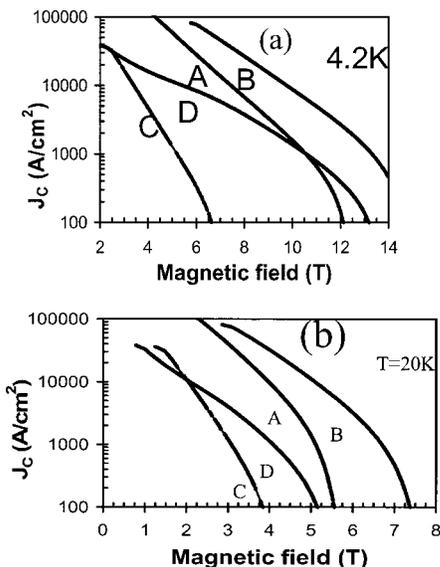


FIG. 4. A comparison of  $J_c(H)$  for the undoped (A), 10 wt % SiC-doped (B), the clean-limit (C), and the Mg-vapor-treated (D) samples at 4.2 K (a) and 20 K (b).

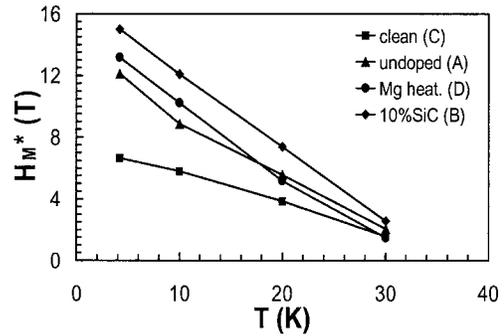


FIG. 5. The irreversibility field  $H_M^*$  vs temperature for the samples A, B, C, and D.

SiC-doped sample reached 7.4 T, compared to 5.6 T for the undoped one, 5.2 T for the Mg-vapor-treated one, and 3.8 T for the clean-limit one at 20 K.

The TEM examination revealed that there are a number of impurity phases in the form of nanometer-size inclusions inside and in between grains in the nano-SiC-doped sample. These impurities include Mg<sub>2</sub>Si, MgB<sub>4</sub>, and MgO detected by x-ray diffraction (XRD) analysis<sup>2,10</sup> and unreacted SiC, amorphous BO<sub>x</sub>, Si<sub>x</sub>B<sub>y</sub>O<sub>z</sub>, and BC detected by using the EELS technique. TEM images show that the grain size of MgB<sub>2</sub> is smaller than 100 nm. Energy dispersive x-ray (EDX) analysis shows that the Mg:Si ratio is identical across the entire sample, indicating that the phase distribution is globally homogeneous. However, the nanoscale impurity phases MgB<sub>4</sub> and MgO are present within the grains. The presence of oxygen within grains is consistent with the results obtained from the above-mentioned 220- $\mu\Omega$  cm thin film with strong pinning, where the ratio of Mg:B:O reached 1.0:0.9:07.<sup>11</sup> Figure 6 is a TEM image showing some unreacted SiC particles and a corresponding lattice image. The EELS analysis [convergence angle ( $\alpha$ )=13 mrad and collection angle ( $\theta_c$ )=18 mrad] shows that this particle is indeed pure SiC without B or any other element in it. The EELS analyses also show other phases present in the SiC-doped sample. Figure 7(a) shows the EELS spectrum of the Si<sub>x</sub>B<sub>y</sub>O<sub>z</sub> phase with no C. The fine structure of both Si and B suggests that the phase is amorphous. Figure 7(b) is the EELS spectrum of the BC phase. Again, the fine structure of B suggests that the phase is amorphous. The EELS of amorphous BO<sub>x</sub> is shown in Fig. 7(c). These phases are often seen in the close vicinity of MgB<sub>2</sub> grains in the sample.

Based on lattice-parameter changes and EDX analysis, we suggested that C and Si might substitute into the lattice in an earlier work.<sup>2</sup> However, in a recent work on SiC-doped MgB<sub>2</sub>, single crystal grown under high pressure (30 kbar) and high temperature (1900–1950 °C) showed there was only C substitution for B but no Si detected in the crystals. The authors revealed that the C substitution for B was as high as 16%, the highest level of substitution in all the C-doping studies so far.<sup>12</sup> There is a clear trend with respect to C substitution in MgB<sub>2</sub> in the literature data.<sup>13–16</sup> The higher the sintering temperature is, the larger the proportion of C that is substituted for B in MgB<sub>2</sub>. As we used relatively low sintering temperatures, 800–850 °C, the C substitution for B is expected to be lower. Figure 8(a) is the Z-contrast

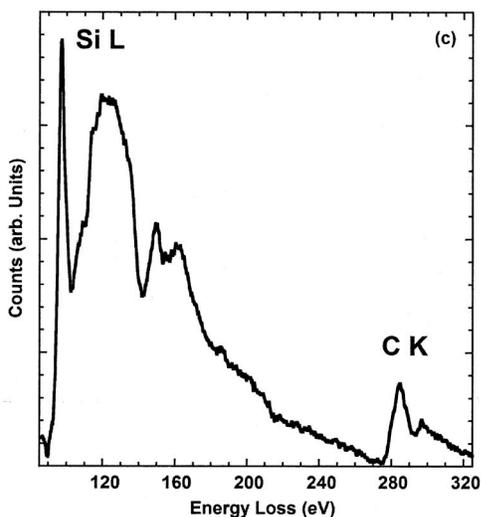
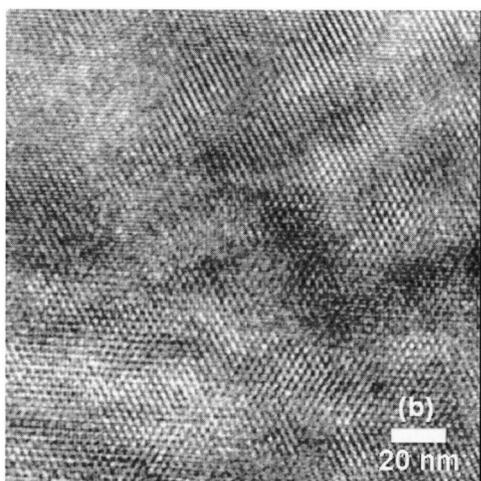
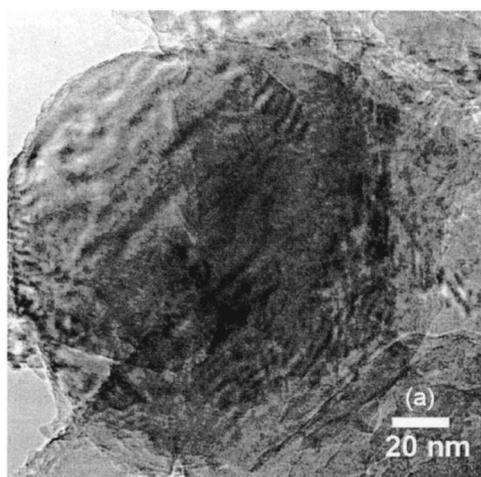


FIG. 6. Conventional TEM image of an unreacted SiC particle, (b) high-resolution TEM image of the bulk of the SiC particle, and (c) EELS spectrum clearly showing the Si L- and the C K-edge.

image<sup>17-19</sup> for the nano-SiC-doped sample, which shows a typical MgB<sub>2</sub> crystal in the [100] orientation. Z-contrast imaging in scanning transmission electron microscopy mode utilizes electrons scattered at high angle (>25 mrad) to form an incoherent image, with an image intensity that is proportional to the square of the average atomic number (i.e.,  $\sim Z^2$ ).

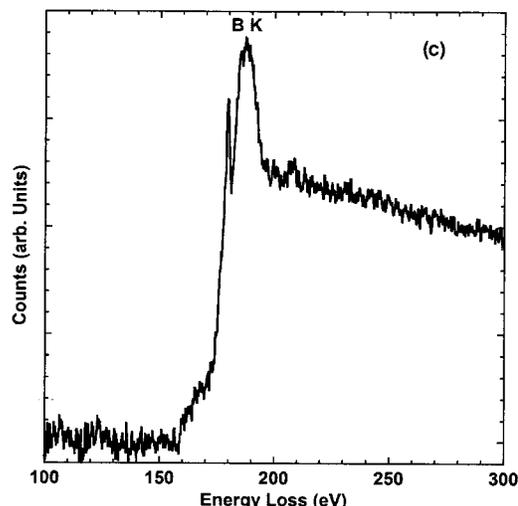
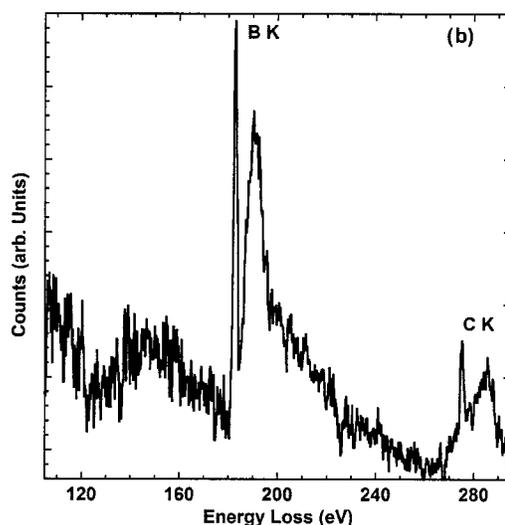
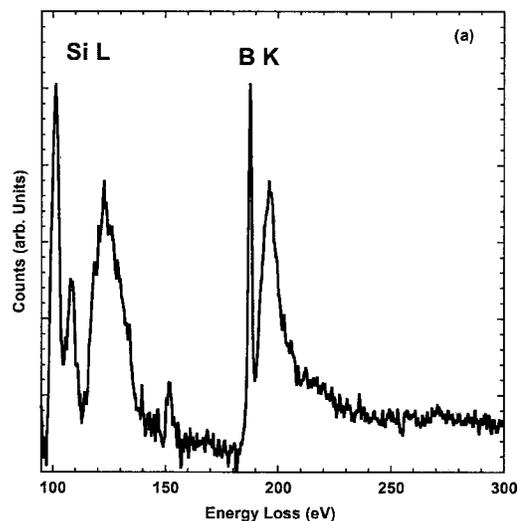


FIG. 7. The EELS spectrum of amorphous (a) Si<sub>x</sub>B<sub>y</sub>O<sub>z</sub>, (b) BC, and (c) BO<sub>x</sub> detected in the SiC-doped MgB<sub>2</sub>.

A close-up look at the atomic structure of the high-resolution lattice image shows that only the Mg columns are visible [Fig. 8(b)], due to the small scattering amplitude of B. The EELS shows the typical fine structure for B in MgB<sub>2</sub>,<sup>20</sup> but no C signal can be detected [Fig. 8(c)]. It should be noted

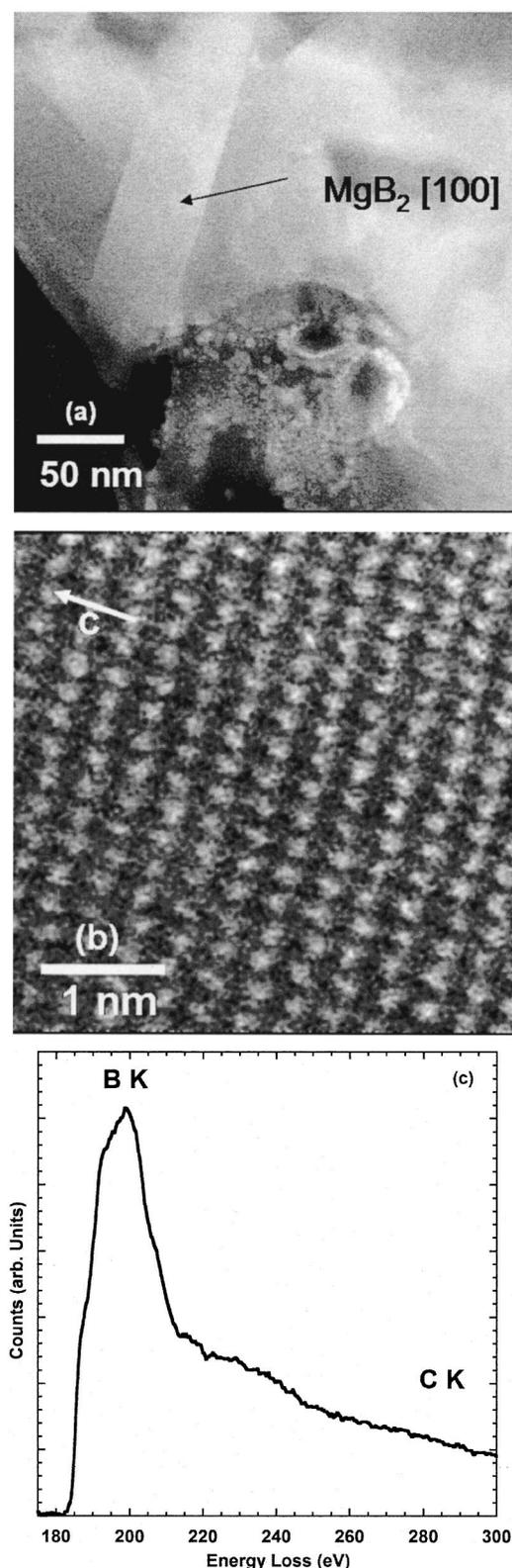


FIG. 8. (a) The Z-contrast image of a typical  $\text{MgB}_2$  grain in the  $[100]$  orientation, (b) high-resolution Z-contrast image of the bulk of the  $\text{MgB}_2$  grain showing the Mg columns only, and (c) EELS spectrum of the B K edge from the  $\text{MgB}_2$  grain.

here that light elements, such as C or B, can be detected in concentrations down to 0.2% with a 10% accuracy in a matrix such as  $\text{MgB}_2$ . However, it is rather difficult to distinguish a small C signal originating from within the lattice or

from surface contamination, as the low signal-to-noise ratio of the C core loss for such low concentrations makes it nearly impossible to distinguish the near-edge fine structure. Due to the large variety of phases present in the SiC-doped sample, it is therefore possible that C substitution at a level of 1%–3%, which is believed to be quite reasonable in the framework of the literature on C substitution,<sup>15–19</sup> cannot be readily identified, and more careful analysis is needed.

In addition to the high concentration of the nanoinclusions, there were structural defects observed in the nano-SiC-doped sample, as reported previously.<sup>21</sup> The majority of nanodomains have a rectangular shape with a domain size of about 2–4 nm. The domain boundaries trap numerous defects to form nanodefekt wells and release the strain caused by the rotation of nanodomains, as reported by Li *et al.*<sup>21</sup> This nanodomain structure may be the result of a small proportion of C substituted for B. In our recent work, we found that C substitution indeed improved flux pinning while also depressing  $T_c$ . It was found that an optimal combination of substitution and addition achieved the best enhancement of flux pinning.<sup>22</sup>

## DISCUSSION

In comparison to all other doping reported so far, the special features of nanoscale-SiC doping into  $\text{MgB}_2$  can be described as follows: (1) the extent of enhancement in  $J_c(H)$  is very large, by more than an order of magnitude above certain fields, (2) the enhancement of  $J_c(H)$  extends to all temperatures up to  $T_c$ , in contrast to most of the other doping studies, which only show that it is effective in enhancing  $J_c(H)$  at low temperatures, (3) although the value of  $H_{c2}$  for the SiC-doped sample is not as high as for the Mg-vapor-treated sample in the field range measured (Fig. 2), the  $J_c(H)$  values for the SiC-doped sample are substantially higher than those of the Mg-vapor-treated sample, in particular, at higher temperatures (Fig. 4). These special features of the SiC-doped samples can be explained in terms of impurity scattering in the framework of two-gap superconductivity theory<sup>3</sup> and the improvement of flux pinning.

### Role of impurity scattering

Recently, Gurevich *et al.*<sup>3</sup> reported a record-high  $H_{c2}(0)=29$  T for untextured sample C and  $H_{c2}^\perp(0)=34$  T and  $H_{c2}^\parallel(0)=49$  T for a high-resistivity film [ $\rho(40$  K) = 220  $\mu\Omega$  cm] using direct, high-field resistivity measurements. In this study, a clean film with a low resistivity of 7  $\mu\Omega$  cm at 40 K had an  $H_{c2}^\parallel$  of 29 T, in comparison to the 49 T of the 220  $\mu\Omega$  cm film. It seems likely that the SiC-doped sample with the highest resistivity of 300  $\mu\Omega$  cm will also have a very high  $H_{c2}$ .

To understand the significant enhancement of  $J_c$  at higher fields for the nano-SiC doping, we measured the resistivity  $\rho$  and residual resistivity ratio RRR for samples A and B, as shown from the resistivity versus temperature curves reported in Fig. 1. For comparison we list some literature data in Table I. The highest value of  $H_M^*$  correlates

TABLE I. Comparison of  $T_c$ , resistivity, and irreversibility field data for samples A, B, C, D, and one literature sample (pure sintered pellet made from  $^{10}\text{B}$ ).<sup>23</sup>

	Sample A: Undoped	Sample B: SiC doped	Sample C: Clean limit	Sample D: Mg vapor treated	Pure bulk
$T_c$ (K)	37.2	36.5	39	36.9	40.2
$\rho$ ( $\mu\Omega$ cm) at 40 K	90	300	1	18	1
RRR	2.1	1.74	14.7	3	19.7
$R(300\text{ K})/R(40\text{ K})$					
$H_M^*$ (20 K) (T)	5.6	7.4	3.9	5.2	3.8

well to the highest value of resistivity, both being found in the SiC-doped sample for which the  $J_c(H)$  characteristics are best, too.

For the sample D, the Mg vapor treatment caused the increase in resistivity from  $1\ \mu\Omega$  cm for the clean-limit sample C to  $18\ \mu\Omega$  cm for sample D. Because Mg vapor treatment will largely affect the Mg sites in the lattice, the disorder in Mg sites will induce *out-of-plane*  $\pi$ -band scattering which will increase the resistivity and  $dH_{c2}/dT$  at low temperatures and hence the  $H_{c2}$  at low temperatures. This was indeed confirmed later by the resistivity measurements in high field, which gave the value of  $H_{c2}(0)$  as about 29 T.<sup>3</sup> Sample D in Table I was measured after aging for two months, during which time the resistivity  $\rho$  dropped from its original value of  $18\text{--}5\ \mu\Omega$  cm at 40 K, while  $T_c$  also increased from 36.9 to 37.7 K, due probably to relaxation of a quenched defect structure. As the Mg vapor treatment is unlikely to introduce impurities at grain boundaries, the increase in resistivity in this case can be considered to be tied to the improvement of  $H_{c2}$ . The improvement of  $H_{c2}$  at low temperatures leads to the improvement in  $J_c(H)$  at low temperatures, as shown from the  $J_c$  versus  $H$  for sample D at 4.2 K in Fig. 4(a). The  $J_c$  of sample D is substantially larger than that of the clean-limit sample C and also crosses over sample A in higher fields.

As for sample B, according to the two-gap superconductivity theory,<sup>3</sup> the nano-SiC doping could lead to two different scattering channels. First, the partial C substitution for B sites causes disorder on the B sites which will result in in-plane  $\sigma$  scattering. The higher  $H_{c2}$  at higher temperatures contributes to the enhancement of  $J_c(H)$  at higher temperatures for the SiC-doped sample. Second, the formation of nanodomain structures is due to the variation of Mg–B spacing which in turn causes disorder at B and Mg sites. These nanodomains with a size of 2–3 nm are also well below the 8–10-nm coherence length of  $\text{MgB}_2$ . These extensive nanodomain defects could result in strong in-plane and out-of-plane scattering and contribute to the increase of resistivity and  $H_{c2}$  in a wide temperature regime. This accounts for the enhancement of  $J_c(H)$  in over a wide temperature range for the SiC-doped sample. Recently, a record-high  $H_{c2}(0)$  value of 37 T for bulk  $\text{MgB}_2$  was achieved from transport measurements on a nano-SiC-doped sample, as reported by Serquis.<sup>24</sup> The strong upturn of  $H_{c2}(T)$  at low temperatures indicates impurity scattering on the Mg sites.

### Role of flux pinning

It should be pointed out that the resistivity of sample B ( $300\ \mu\Omega$  cm) is much larger than that of sample D ( $18\ \mu\Omega$  cm), although sample D has higher  $H_{c2}$  than sample B at least in the field region up to 9 T. In fact, the resistivity for the undoped sample A is already larger than sample D by a factor of 5, although the  $H_{c2}$  for A is much lower than for D. The large difference in resistivity between the Wollongong samples (A and B) and the Madison samples (C and D) is attributable to the different processing parameters used.<sup>25</sup> Samples C and D were prepared at 950–960 °C for 24 h in a highly protected environment (closed in a Ta or Nb tube and then sealed in a quartz tube), while samples A and B were processed at 800 °C for only 30 min in a sealed Fe tube, which cannot prevent oxygen diffusion through to the sample. As a result, samples C and D are of high quality, have large grains, and are free from impurities, as indicated by XRD,<sup>4</sup> while sample A has more impurities such as  $\text{MgO}$  which can be seen in the XRD pattern.<sup>2</sup> For sample B, the 10 wt % SiC addition into the sample A resulted in a further increase in the concentration of impurities through the reaction of SiC with Mg and B, including  $\text{MgSi}_2$ , BC,  $\text{BO}_x$ ,  $\text{SiBO}_x$ , and unreacted SiC as identified by EELS and XRD. The resistivity of samples A and B may be strongly affected by the extrinsic mechanisms, such as scattering on the grain boundaries or second phase inclusions, which do not contribute to the enhancement of  $H_{c2}$  but certainly increases the global resistivity.

On the other hand, the additional impurities at nanoscale introduced by SiC doping can serve as strong pinning centers to improve flux pinning within a certain field region. This is clearly demonstrated by the superior  $J_c$ – $H$  performance of the SiC-doped sample B, as shown in Figs. 3 and 4. It is particularly interesting to note that in Fig. 4(b), the  $J_c$  for B is higher than for D by a factor of as large as 100 at 20 K and 5 T even though the  $H_{c2}$  for D is higher than for B. This is further confirmed by the higher irreversibility line for B than for D, as shown in Fig. 5. The potential pinning centers introduced by SiC doping include inclusions, such as highly dispersed  $\text{MgSi}_2$ , BC,  $\text{BO}_x$ , and  $\text{SiBO}_x$ , which are all at a scale below 10 nm, match the coherence length very well and can act as strong pinning centers. Some large impurity particles such as unreacted SiC, as shown in Fig. 6(a), would not be effective pinning centers but act to reduce the superconducting volume and thus should be eliminated in order to further improve the zero-field  $J_c$ . In addition, the extensive network of nanodomain defects at a scale of 2–3 nm would

provide very effective collective pinning at all the temperatures up to  $T_c$ . All these defects are absent in sample D as Mg vapor treatment would not introduce these impurities. Thus, the flux pinning in sample D is not as strong as in sample B, at least at higher temperatures. At lower temperatures, the strong intrinsic scattering induced by Mg vapor treatment significantly enhances the  $H_{c2}$  which will, in turn, improve  $H_M^*$  and  $J_c$  at low temperatures for sample D. The fact that sample D has higher  $H_{c2}$  but lower  $H_M^*$  than sample B indicates that there are two closely related but distinguishable mechanisms that control the  $J_c(H)$  characteristics:  $H_{c2}$  and flux pinning. The  $H_{c2}$  is the primary factor that sets the upper limit to  $H_M^*$  while the flux pinning is important to bring  $H_M^*$  closer to  $H_{c2}$  and improve the  $J_c(H)$  within certain field regimes. These results suggest that we can manipulate the processing parameters that could lead to the improvement of either  $H_{c2}$  or flux pinning or of both at the same time. Recently, Matsumoto *et al.* reported that their Fe-sheathed SiC-doped MgB<sub>2</sub> wire achieved an  $H_M^*$  value of 23 T at 4.2 K by transport measurements,<sup>25</sup> which is consistent with the results we obtained previously<sup>26</sup> but extends them to higher fields. This is comparable to the conventional metallic superconductor, Nb<sub>3</sub>Sn. The high value of  $H_M^*$  suggests that  $H_{c2}(0)$  for the SiC-doped sample would be greater than that for sample D (29 T).<sup>3</sup> This confirms that there are two mechanisms that are responsible to the significant enhancement of  $J_c(H)$  performance in all the fields and temperatures: increase of  $H_{c2}$  and improvement of flux pinning in the SiC-doped sample.

## CONCLUSION

In the framework of two-gap superconductivity, we have studied a set of four samples with very different resistivity and  $J_c$  characteristics as a result of different processing parameters and SiC doping. We have demonstrated that there are two closely related but distinguishable mechanisms that control the performance of  $J_c(H)$ :  $H_{c2}$  and flux pinning. Mg exposure to the clean-limit sample causes disorder on Mg sites whose scattering leads to the enhancement of  $H_{c2}$ . Nanoscale SiC doping into MgB<sub>2</sub> enhances both  $H_{c2}$  and flux pinning through multiple-scattering channels. Alloying at B and Mg sites due to C substitution and the formation of nanodomain structures causes strong scattering over a wide range of temperatures, leading to enhancement in  $H_{c2}$ . A high concentration of various nanoscale impurity phases results in high resistivity, a low residual resistivity ratio, and a large irreversibility field and upper critical field with modest  $T_c$  reduction. The highly dispersed nanoscale precipitates MgSi<sub>2</sub>, BC, BO<sub>x</sub>, and SiBO<sub>x</sub> and the extensive domain structures at a scale well below 10 nm serve as strong pinning centres. Doping with SiC enhances the critical current den-

sity, the irreversibility field, and the upper critical field in a manner that helps make MgB<sub>2</sub> potentially competitive with both low- and high- $T_c$  superconductors.

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