The combined influence of connectivity and disorder on $J_c$ and $T_c$ performances in MgxB2+10 wt % SiC

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The combined influence of connectivity and disorder on $J_c$ and $T_c$ performances in Mg$_{1+x}$B$_2$+10 wt % SiC

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The influences of connectivity and disorder on the critical current density $J_c$ are discussed to clarify the different mechanisms of $J_c(H)$ enhancement in different magnetic field ranges. Excess Mg in Mg$_{1+x}$B$_2$+10 wt % SiC composites effectively improves the connectivity, as evidenced by both the resistivity properties and the Raman scattering. The promising $J_c(H)$ of Mg$_{1.15}$B$_2$+10 wt % SiC is attributed to both the high connectivity and the improved irreversibility field, $H_{irr}$, which is in agreement with the Raman fitting analysis. Raman scattering measurements suggest a strengthened electron-$E_{2g}$ coupling and weakened disorder with Mg addition. © 2009 American Institute of Physics. [doi:10.1063/1.3253757]

I. INTRODUCTION

Study of the flux pinning mechanism in MgB$_2$ superconductor is of fundamental interest both for understanding its supercurrent properties and for breaking through the technical obstacles that inhibit its practical applications. The deparing current density, $J_p$, is $\sim 8.7 \times 10^8$ A/cm$^2$ for pure MgB$_2$, as estimated from the Ginzburg–Landau formula $J_p = \Phi_0 / (3/4) \pi \rho_0 \lambda^3(T) \xi(T)$, where $\Phi_0$ is the flux quantum, $\rho_0$ is the permeability of vacuum, $\lambda$ is the penetration depth, and $\xi$ is the coherence length.1 The self-field critical current density, $J_s(0)$, in the best connected samples indicates the ultimate current-carrying potential in the superconductor, which has been reported as $3.5 \times 10^7$ A/cm$^2$ at 4.2 K and $1.6 \times 10^8$ A/cm$^2$ at 2 K in clean films made by hybrid physical-chemical vapor deposition.2,3 These values are about 4% and 20% of the $J_s$ values. Compared with these values, the $J_s(0)$ values in polycrystalline MgB$_2$ bulks and wires are very low and have great potential to be improved. Although the $J_s(H)$ values can be improved through chemical doping, they are still much lower than $J_s$ and drop exponentially with applied magnetic field ($H_H$).4 How to keep the $J_s(H)$ close to $J_s$ is the key to obtaining higher practicable supercurrents in MgB$_2$. Chen et al.5 reported the possibility of improved $J_s$ in low magnetic fields from a high Mg content in MgB$_2$. Zeng et al.6 showed that 10 wt % Mg excess could improve the $J_s$ values of MgB$_2$ wires over the entire range of measurement fields, 0–8.5 T.

In this work, the effects of excess Mg combined with nano-SiC codoping on the current transport properties and flux pinning mechanisms of MgB$_2$/Fe wires were studied systematically. To confirm the cooperation between the connectivity and lattice distortion, Raman scattering measurements were employed in this study. Raman scattering can disclose the crystal strain caused by lattice distortion and chemical pressure, which is related to the flux pinning force ($F_p$).7

II. EXPERIMENTAL

The powder-in-tube process was employed to make practical MgB$_2$ wires from starting powders of Mg (99%), B (99%, amorphous), and SiC (<30 nm). The starting powders were weighed according to the molecular formula for MgB$_2$, Mg$_{1.15}$B$_2$, and Mg$_{2.1}$B$_2$+10 wt % SiC, with $x=1.00$, 1.15, 1.20, 1.25, and 1.30. Sintering was carried out at 650 °C for half an hour under a flowing argon gas atmosphere. The MgB$_2$ cores extracted from the annealed wires were ground and subjected to x-ray diffraction (XRD) measurements. The microstructures were observed by field emission gun scanning electron microscopy (FEG-SEM: JSM-6700F). The magnetic $J_s$ was derived from the height of the magnetization loop $\Delta M$ using the Bean model: $J_s = 15 \Delta M / [\pi a^2 h]$, where $a$ and $h$ are the radius and height of a cylindrical sample. The temperature dependent resistivity, $\rho(H,T)$, was measured with $H$ applied perpendicular to the current direction, using the four-probe method in the temperature range from 15 to 300 K and the field range from 0 to 8.7 T. The upper critical field, $H_{c2}$, and the irreversibility field, $H_{irr}$, were defined as the magnetic field values at 90% and 10% of the superconducting transition on the $\rho(H,T)$ curve, respectively. The Raman scattering was measured with a confocal laser Raman spectrometer (Renishaw inVia plus) with a 100× microscope. The 514.5 nm line of an Ar$^+$ laser was used for excitation.

III. RESULTS

The Mg excess and nanosize SiC doping effects on the phases and superconducting properties were first studied to explore their different functions in the behavior of the critical temperature, $T_c$, and the critical current density, $J_c$. Figure 1 shows the XRD patterns for the wire cores of MgB$_2$, ...
Mg$_{1.15}$B$_2$ and Mg$_{2}$B$_2$+10 wt % SiC. These XRD patterns reveal that nearly single-phase MgB$_2$ was obtained in pure, stoichiometric MgB$_2$ samples along with a small amount of MgO. A small amount of residual Mg could be found in Mg$_{1.15}$B$_2$, while the Mg$_x$Si content was obvious in Mg$_{2}$B$_2$+10 wt % SiC. Figure 2 shows the field dependence of $J_c$ at 5 and 20 K in fields of 0–8.5 T. The pure MgB$_2$ shows a gradual decrease in $J_c$ with increased measurement field at 5 and 20 K. The $J_c$ values of Mg$_{1.15}$B$_2$ in low magnetic field, $<\sim$ 2.5 T at 5 K and $<\sim$ 1.1 T at 20 K, are the highest. Although this behavior cannot be observed directly at 5 K because the $J_c$ values at 5 K are difficult to calculate due to the flux jumping, it is obvious that the $J_c$ values for Mg$_{1.15}$B$_2$ wires at 20 K exceed those of all the other samples. The nano-SiC addition has improved the $J_c$ over that of the pure stoichiometric MgB$_2$ in high measurement fields, $>\sim$ 7 T at 5 K and $>\sim$ 4 T at 20 K. However, the $J_c$ values are very low at low field, as in previous results. The superconducting transitions are shown in the inset of Fig. 2. The additional Mg degrades the superconductivity of MgB$_2$, and the $T_c$ value of Mg$_{1.15}$B$_2$ is 0.7 K lower than that of the stoichiometric MgB$_2$ because the residual Mg is a source of impurity scattering effects. The $T_c$ value, 35.3 K, for Mg$_{2}$B$_2$+10 wt % SiC is even lower than that of Mg$_{1.15}$B$_2$ because of both the C substitution (estimated as $\sim$ 2.2%, as shown in Table I) and the impurity scattering.

To take advantage of the doping effects of Mg and SiC, Mg$_{2}$B$_2$+10 wt % SiC samples were studied with $x=1.15$, 1.20, 1.25, and 1.30. The residual Mg contents are similar in all the samples, as indicated by the XRD patterns in Fig. 1. As estimated from the XRD patterns, the C substitution content is $\sim$ 2.3% in Mg$_{1.15}$B$_2$+10 wt % SiC, which is similar to that in Mg$_{2}$B$_2$+10 wt % SiC, and then it drops to 2.1% in Mg$_{2.20}$B$_2$+10 wt % SiC and 1.8% in both Mg$_{2.25}$B$_2$+10 wt % SiC and Mg$_{1.30}$B$_2$+10 wt % SiC. It seems that the MgB$_2$ lattice excludes C atoms due to the excess Mg, as in the trend shown in Table I. The most impressive advantage of the Mg excess samples is the improved $J_c(H)$ in low magnetic field, especially at 5 K. However, the $J_c$ values drop quickly for Mg$_{2.20}$B$_2$+10 wt % SiC, Mg$_{2.25}$B$_2$+10 wt % SiC, and Mg$_{1.30}$B$_2$+10 wt % SiC because of the high impurity content. The decreased $T_c$ values with increasing Mg content are responsible for the $J_c$ drop at 20 K. The $J_c$ performance of Mg$_{1.15}$B$_2$+10 wt % SiC improves over the whole range of applied magnetic field due to the optimum ratio of the component elements.

Although the $T_c$ values decrease with the dopant contents, the different superconducting transition processes suggest quality differences between the MgB$_2$ crystals according to the normalized ac susceptibility, as shown in the inset of Fig. 2. The transition width, $\Delta T_c$, is defined as the temperature difference between the onset value of the transition (the point where the ac susceptibility deviates from zero), $T_{c,onset}$, and the termination point of the transition (the point where the ac susceptibility reaches $-1$), $T_{c,zero}$. All the values of $T_{c,onset}$ and $\Delta T_c$ are listed in Table I. The superconducting transition of MgB$_2$ is quite narrow, with a $\Delta T_c$ of 0.98 K. However, it is even sharper for Mg$_{1.15}$B$_2$, with a $\Delta T_c$ of 0.85 K, which is attributed to the improved quality and connectivity of the MgB$_2$ because the flux is easy to expel with the decreased temperature. The transition of Mg$_{2}$B$_2$+10 wt % SiC is very broad, with a $\Delta T_c$ of 3.36 K, because of the strong disorder. Excess Mg in Mg$_{2}$B$_2$+10 wt % SiC is of benefit to the transition: the $\Delta T_c$ of Mg$_{1.15}$B$_2$+10 wt % SiC drops significantly to 1.69 K. More Mg addition decreases the value of $\Delta T_c$ to $\sim$ 1.55 K in Mg$_{2}$B$_2$+10 wt % SiC ($x>1.20$). The stable $\Delta T_c$ values indicate that the quality cannot be improved with more Mg addition in Mg$_{2}$B$_2$+10 wt % SiC.

### IV. DISCUSSION

#### A. Connectivity

Based on the collective pinning model, the disorder-induced spatial fluctuations in the vortex lattice can be clearly divided into different regimes according to the strength of the applied field: single-vortex, small-bundle, large-bundle, and charge-density-wave type relaxation of the vortex lattice. For high current purposes, the practicable $J_c$ for MgB$_2$ is in the regions of single-vortex pinning and small-bundle pinning in the phase diagram. It is believed that the quality of the connectivity is responsible for the $J_c$ performance in the single-vortex pinning regime (low magnetic field region) due to its weak field dependence, while the
disorder is responsible for the $J_c$ performance in the small-bundle regime (high magnetic field) due to the strong $F_p$.

To explore the effect on connectivity of Mg excess, microstructures of all samples were observed with SEM, as shown in Fig. 3. The grains in MgB$_2$ samples show an independent growth process, which is responsible for their isolated distribution. The grains in Mg$_{1.15}$B$_2$ have clearly melted into big clusters because the additional Mg can extend the liquid reaction time. The grain shapes in MgB$_2$ +10 wt % SiC are different from that in pure, stoichiometric MgB$_2$ because the crystals are grown under strain due to the C substitution effect. The strain is also strong in Mg$_{1.15}$B$_2$ +10 wt % SiC, as long bar grains can be observed under SEM. The strain is released in the high Mg content samples, judging from the homogeneous grain sizes and shapes. Compared with MgB$_2$ +10 wt % SiC, the grain connectivity improved greatly with the increasing Mg addition. The grains are merged into big particles. This means that the grain boundaries have replaced the gaps between grains. However, more impurities are induced in forms such as residual Mg and MgO.

Compared with the microstructure observations, the resistivity is a more reliable and quantitative method to estimate the connectivity. The resistivity dependences on temperature are shown in Fig. 4. The $\rho_{300 \, K}$ and $\rho_{40 \, K}$ of MgB$_2$ wires are always high compared with thin films and bulks due to the low density of the wire samples. Although residual Mg exists in Mg$_{1.15}$B$_2$, its resistivity is decreased due to the

<table>
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<th>Samples</th>
<th>MgB$_2$</th>
<th>Mg$_{1.15}$B$_2$</th>
<th>MgB$_2$</th>
<th>Mg$_{1.15}$B$_2$</th>
<th>Mg$_{1.20}$B$_2$</th>
<th>Mg$_{1.25}$B$_2$</th>
<th>Mg$_{1.30}$B$_2$</th>
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<td>$T_{c,\text{onset}}$ (K)</td>
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<td>36.50</td>
<td>35.32</td>
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<td>34.28</td>
<td>34.04</td>
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<td>$\Delta T_c$ (K)</td>
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<td>0.85</td>
<td>3.36</td>
<td>1.69</td>
<td>1.57</td>
<td>1.53</td>
<td>1.54</td>
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<tr>
<td>$a$-axis (Å)</td>
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<td>3.0881</td>
<td>3.0771</td>
<td>3.0766</td>
<td>3.0773</td>
<td>3.0786</td>
<td>3.0788</td>
</tr>
<tr>
<td>C content (%)</td>
<td>N/A</td>
<td>N/A</td>
<td>2.2</td>
<td>2.3</td>
<td>2.1</td>
<td>1.8</td>
<td>1.8</td>
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<tr>
<td>$N(0)$ (μΩ cm)</td>
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<td>0.354</td>
<td>0.332</td>
<td>0.331</td>
<td>0.333</td>
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<tr>
<td>$\rho_{40 , K}$ (μΩ cm)</td>
<td>86.8</td>
<td>70.0</td>
<td>277.2</td>
<td>152.2</td>
<td>168.6</td>
<td>191.3</td>
<td>213.3</td>
</tr>
<tr>
<td>$\rho_{300 , K}$ (μΩ cm)</td>
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<td>246.3</td>
<td>269.8</td>
<td>307.7</td>
<td>347.3</td>
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<td>94.1</td>
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<td>$A_F$</td>
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<td>0.062</td>
<td>0.096</td>
<td>0.089</td>
<td>0.077</td>
<td>0.067</td>
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FIG. 3. SEM images of MgB$_2$ (a), Mg$_{1.15}$B$_2$ (b), MgB$_2$+10 wt % SiC (c), Mg$_{1.15}$B$_2$+10 wt % SiC (d), Mg$_{1.20}$B$_2$+10 wt % SiC (e), Mg$_{1.25}$B$_2$+10 wt % SiC (f), and Mg$_{1.30}$B$_2$+10 wt % SiC (g).

FIG. 4. Temperature dependence of resistivity curves of MgB$_2$, Mg$_{1.15}$B$_2$, and Mg$_{1.20}$B$_2$+10 wt % SiC (x=1.00, 1.15, 1.20, 1.25, and 1.30).
highly connected grain boundaries. The resistivity increases greatly in MgB₂+10 wt % SiC because of the strong band scattering effects due to the C substitution on B sites. The strong localization of the boron pₓ orbitals on the boron plane, which is sensitive to the carbon substitution for boron, is responsible for the great resistivity enhancement. Additionally, the reduction in charge carriers has to be considered because the conducting holes are annihilated by the electrons introduced into the system by the substitutional carbon atoms. The most amazing effect of Mg excess is that the resistivity of MgB₂+10 wt % SiC is reduced by ~45% with 15% Mg addition. This is because the grain boundary scattering effects are greatly weakened. Although a greater Mg 15% Mg addition. This is because the grain boundary scattering effects are greatly weakened. Although a greater Mg excess (x > 1.15) ought to also reduce the resistivity of MgB₂+10 wt % SiC, the increased scattering effects of the impurity phases make the resistivity a little higher than that of Mg₁.₁₅B₂+10 wt % SiC.

A practical quantity to evaluate the connectivity is the active area fraction, \( A_F = \Delta \rho_{\text{ideal}} / \Delta \rho (300 \text{ K}) \), where \( \Delta \rho_{\text{ideal}} = \rho_{\text{ideal}}(300 \text{ K}) - \rho_{\text{ideal}}(40 \text{ K}) \approx -9 \mu \Omega \text{ cm} \) is the resistivity of fully connected MgB₂ without any disorder, and \( \Delta \rho (300) = \rho(300 \text{ K}) - \rho(T) \). The \( A_F \) values for all the samples are listed in Table I. It should be noted that the connectivity is far removed from that found in ideal crystals, as reflected by the low \( A_F \) values. Although the \( A_F \) values of pure and 10% SiC doped MgB₂ are just 0.106 and 0.062, the additional Mg can improve them to 0.162 and 0.096 for 15 wt % Mg excess samples, respectively. High \( A_F \) values are the reflection of a broad channel of supercurrents, while impurities reduce the connectivity in large x samples, as shown in Table I. Considering the weak field dependence of \( J_c \) in the ranges of \( H_i < \sim 0.7 \text{ T} \), 20 K and \( H_i < \sim 2 \text{ T} \), 5 K, as shown in Fig. 2, it is believed that the high connectivity improves the supercurrent channels because the currents can easily meander through the well-connected grains.

B. \( H_{\text{irr}}, H_{\text{irr}}, \) and \( F_p \)

In contrast with the high connectivity mechanism in the single-vortex region, a strong flux pinning force is the dominant feature that keeps the \( J_c \) high in the small-bundle region. Most defects, lattice distortions, and grain boundaries are believed to act as flux pinning centers. Carbon substitution for boron in the MgB₂ lattice can cause great lattice distortion because of the shorter C–B bonds. Furthermore, \( H_{\text{irr}} \) and \( H_{\text{irr}} \) are significantly enhanced by the increased band scattering due to carbon substitution. The competitive \( J_c \) in high field for MgB₂+10 wt % SiC and Mg₁.₁₅B₂+10 wt % SiC should be ascribed to the enhanced \( H_{\text{irr}} \) and \( H_{\text{irr}} \). Figure 5 compares the \( H_{\text{irr}} \) and \( H_{\text{irr}} \) performances for all the samples. The \( H_{\text{irr}} \) and \( H_{\text{irr}} \) values of pure, stoichiometric MgB₂ increase gently as the temperature is decreased, while the enhancement is a little slower in Mg₁.₁₅B₂. MgB₂+10 wt % SiC has high \( H_{\text{irr}} \) and \( H_{\text{irr}} \) values in the range of \( T < \sim 24 \text{ K} \), and the additional Mg in Mg₁.₁₅B₂+10 wt % SiC does not degrade the good performance of \( H_{\text{irr}} \) and \( H_{\text{irr}} \). The high \( H_{\text{irr}} \) and \( H_{\text{irr}} \) of Mg₁.₁₅B₂+10 wt % SiC are attributed to both the carbon substitution and the high amount of impurity phase.

The \( H_{\text{irr}} \) is responsible for the \( J_c \) performance in high magnetic field because the field dependence of \( F_p \) is proportional to \( h^2(1-h)^{m} \), with \( h = H/H_{\text{irr}} \), and the values of \( m \) and \( \alpha \) depend on the different flux pinning mechanisms. The \( h \) dependence of \( F_p/F_{p,max} \) is shown in Fig. 6. The x-axis is normalized because the values of \( H_{\text{irr}} \) are higher than those in Refs. 15 and 16 due to the different standards for \( H_{\text{irr}} \). The most striking phenomena in Fig. 6 are (i) the \( F_p/F_{p,max} \) for Mg₁.₁₅B₂+10 wt % SiC is effective both in the single-vortex region and in the small-bundle region; (ii) the \( F_p/F_{p,max} \) for Mg₁.₁₅B₂+10 wt % SiC is the most effective in the small-bundle region; (iii) the \( F_p/F_{p,max} \) values for MgB₂ and Mg₁.₁₅B₂ are uncompetitive in the small-bundle region; and (iv) the \( F_p/F_{p,max} \) values for MgB₂+10 wt % SiC (\( x = 1.20, 1.25, \) and 1.30) are also effective in the small-bundle region. These phenomena are in agreement with the \( J_c \) behavior shown in Fig. 2: (i) the \( J_c \) performance for Mg₁.₁₅B₂...
+10 wt% SiC is promising over the whole range of the \( H_A \); (ii) the \( J_a \) values for MgB\(_4\) remain constant with the \( H_A \), and it can be imagined that the \( J_a \) values are higher than those of Mg\(_{1.15}\)B\(_2\) +10 wt% SiC in the stronger \( H_A \); (iii) the \( J_a(0) \) values of MgB\(_2\) and Mg\(_{1.15}\)B\(_2\) are outstanding, however, they drop quickly with \( H_A \); and (iv) the high impurity content degrades the \( J_a \) performance in low \( H_A \) for Mg\(_{1.15}\)B\(_2\) +10 wt% SiC \((x = 1.20, 1.25, \text{ and } 1.30)\), whereas these impurities are also efficient flux pinning centers in high \( H_A \). The high \( J_a \) of Mg\(_{1.15}\)B\(_2\) +10 wt% SiC is ascribed to the high \( A_e \), high \( H_{12} \), and high \( H_{13} \). The \( H_{12} \) and \( H_{13} \) values of Mg\(_{1.15}\)B\(_2\) +10 wt% SiC \((x \approx 1.20)\) are lower than those of Mg\(_{1.15}\)B\(_2\) +10 wt% SiC, which is attributed to the high impurity content of the samples.

C. Raman spectra

Raman scattering is employed to study the combined influence of connectivity and lattice distortion. Chemical substitution\(^{13} \) and lattice distortion\(^{17} \) are expected to modify the phonon spectrum, by changing the phonon frequency and the electron-phonon interaction. The effects of C substitution include an increase in impurity scattering and band filling, which reduces the density of states (DOS) and alters the shape of the Fermi surface. The \( E_{2g} \) phonon peak shifts to the higher energy side, and the peak is narrowed with increasing \( x \) in Mg\(_{1-x}\)B\(_2\)\(_{1-x}\)C\(_x\). As a carbon source, nano-SiC shows a similar influence, due to its C atoms, on the \( J_a \), \( H_{12} \), and even Raman spectra in MgB\(_2\). Figure 7 shows the Raman spectra fitted with three peaks: \( \omega_1 \), \( \omega_2 \), and \( \omega_3 \). The \( \omega_1 \) and \( \omega_3 \) peaks are understood to arise from sampling of the phonon density of states (PDOS) due to disorder, while \( \omega_2 \) is associated with the \( E_{2g} \) mode, which is the only Raman active mode for MgB\(_2\).\(^{18} \) A reasonable explanation for the appearance of \( \omega_1 \) and \( \omega_3 \) is the violation of Raman selection rules induced by disorder. All three peaks are broad, as in previous results, due to the strong electron-phonon coupling. The influence of \( \omega_2 \) on the superconducting performance is negligible compared with those of \( \omega_2 \) and \( \omega_3 \) because of its weak contribution to the Raman spectrum. The frequency and full width at half maximum (FWHM) of \( \omega_2 \) and \( \omega_3 \) are shown in Fig. 8. Both \( \omega_2 \) and \( \omega_3 \) are hardened with SiC addition. The \( \omega_2 \) frequency is reduced with further Mg addition, whereas the \( \omega_3 \) frequency remains almost stable. The frequencies of \( \omega_2 \) for the \( x \approx 1.20 \) samples are even lower than that of the pure, stoichiometric MgB\(_2\). The FWHM of \( \omega_2 \) decreases with SiC doping, while the Mg excess remedies this trend. On the contrary, the \( \omega_3 \) FWHM increases with SiC addition and becomes narrow with more addition of Mg.

The Raman scattering properties are the direct reflection of the phonon behaviors of MgB\(_2\). The parameters of Raman spectra vary with the composition of MgB\(_2\)\(_{1-x}\)C\(_x\) and the influences of their surroundings, which depend on both the connectivity and the disorder of the samples. Furthermore, the disorder should be considered as intrinsic and extrinsic parts based on their different sources. The crystallinity and chemical substitution are believed to be responsible for the intrinsic disorder effects, while the grain boundaries and impurities are treated as responsible for the extrinsic disorder effects. The influences of intrinsic disorder on the basic characteristics of Raman spectra are significant because the
physical properties of MgB$_2$ depend on the intrinsic disorder. The Raman parameters can also be tuned by the extrinsic disorder. Especially in the samples with good connectivity, the influences of grain boundaries and impurities on the Raman spectra are taken into account because of their strain effects on the MgB$_2$ crystals. The differences between shifts and FWHMs in Raman spectra for MgB$_2$, Mg$_{1.15}$B$_2$, MgB$_2$+10 wt % SiC, and Mg$_{1.15}$B$_2$+10 wt % SiC are mostly attributed to the intrinsic characteristics because of their different chemical compositions. The Raman spectra of MgB$_2$, MgB$_2$+10 wt % SiC (x > 1.20) can be considered as the gradual modification of that of Mg$_{1.15}$B$_2$+10 wt % SiC. The weakened C substitution effects are responsible for the decreased frequencies and slightly increased FWHMs of $\omega_2$ with Mg addition. Accordingly, the FWHMs of $\omega_3$ decrease due to the weakened lattice distortion. Although the $A_F$ values are quite low for MgB$_2$, MgB$_2$+10 wt % SiC (x > 1.20), the effects of extrinsic disorder on Raman parameters are considerable through the MgB$_2$-MgB$_2$ and MgB$_2$-impurities interfaces and the connectivity degrades with the increased x values due to the decreased numbers of MgB$_2$-MgB$_2$ interfaces. The high FWHM value of $\omega_2$ is correlated with the high self-field $J_s$ due to the high carrier density, while the high FWHM value of $\omega_3$ is correlated with the high-field $J_s$ because of the strong flux pinning force due to the large disorder. The FWHM behaviors show that high connectivity and strong disorder are best combined in Mg$_{1.15}$B$_2$+10 wt % SiC among all the samples.

D. Electron-phonon coupling

Not only are the Raman parameters related to the superconducting properties, but they are also related to the electron-phonon coupling (EPC) strength. The broad FWHM of $\omega_2$ and the narrow FWHM of $\omega_3$ in Raman spectra imply a strong electron-$E_{2g}$ coupling in MgB$_2$. According to the frequency and the linewidth of the $E_{2g}$ mode reflected in the Raman spectra, direct evaluation of the contribution of the $E_{2g}$ mode to the EPC is possible due to the negligible anharmonic effects in the system. The constant of the electron-$E_{2g}$ coupling is given by the Allen equation:

$$\Gamma_2 = 2\pi\lambda_{E_{2g}}N(0)\omega_2^2,$$

where $\lambda_{E_{2g}}$ is the strength of the electron-$E_{2g}$ coupling and $N(0)$ is the DOS (per spin per unit energy per unit cell) on the Fermi surface and is the only electronic property explicitly occurring in this equation. The measured phonon frequency and phonon linewidth, in the absence of anharmonic contributions, are simply and directly related to the EPC constant, $\lambda_{E_{2g}}$. The total DOS at $E_F$ in pure MgB$_2$ is taken as $N(0)$=0.354 states/eV/cell/spin, with the contribution from the $\sigma$ band being 0.15 states/eV/cell/spin and that from the $\pi$ band being 0.204 states/eV/cell/spin, respectively. The $N(0)$ is assumed to be constant for the small changes of electron and holes in Mg$_{1.15}$B$_2$ and MgB$_2$. The factor of $N^2(0)$ dependence on the carbon concentration is about $-1$ in the range of 0-0.04, while $N^3(0)$ remains constant. Taking the C substitution contents shown in Table I into account, the $\lambda_{E_{2g}}$ values for the different samples can be obtained from the Allen equation, as shown in Fig. 9. These values are in agreement with 2.5 ± 1.1 obtained for the $q=0.2\Gamma_A E_{2g}$ mode from inelastic x-ray scattering measurements. It is expected that the $\lambda_{E_{2g}}$ of MgB$_2$+10 wt % SiC is just a little smaller than that of the pure, stoichiometric MgB$_2$. A possible explanation is that both the samples are imperfect because of the low temperature sintering and impurity phases. The crystallinity is improved for all the Mg excess samples according to the higher $\lambda_{E_{2g}}$ values. The $\lambda_{E_{2g}}$ values in MgB$_2$+10 wt % SiC (x > 1.20) are even higher than that for Mg$_{1.15}$B$_2$, which is attributed to the low $N(0)$ values and the large amount of Mg excess. This phenomenon is in agreement with the strong FWHM values of $\omega_2$ and the low FWHM values of $\omega_3$.

Although their $E_{2g}$ mode contribution to the EPC is higher, the total EPC constants are degraded by the high impurity contents, which can be estimated with the McMillan formula, as modified by Allen and Dynes:

$$T_c=(\omega_0^2)/(1.2\exp(-1.04(1+\lambda)/\mu^2(1+0.62A)),$$

where $\omega_0$=(390 $\times$ $\omega_{E_{2g}}$ $\times$ 690) is the averaged phonon frequency, with 690 and 390 cm$^{-1}$ being the phonon frequencies of the other modes in the MgB$_2$ system (taken from Ref. 27), $\mu^2$ is the Coulomb pseudopotential, taken as equal to 0.13, and $\lambda$ is the EPC constant. Taking these values, the $\lambda$ for each sample is calculated and is shown in the inset of Fig. 9. The $\lambda$ of MgB$_2$ is the highest one because of its high $T_c$ and low impurity content. Both Mg and SiC addition depress $\lambda$, which is in agreement with the C substitution effect and the high impurity contents shown in the XRD patterns.

V. CONCLUSIONS

In conclusion, connectivity and disorder are both important factors in improvement of the $J_s$ performance of MgB$_2$. The connectivity is responsible for the high $J_s$ performance in the single-vortex regime, while strong disorder is responsible for the promising $J_s$ in the small-bundle pinning regime. The promising $J_s$ values in Mg$_{1.15}$B$_2$+10 wt % SiC are the result of optimized connectivity and disorder, which are reflected in the Raman spectrum, with both a strong $E_{2g}$ mode and a strong PDOS. The Raman scattering measure-
ments imply that excess Mg is effective in improvement of the connectivity of MgB2 grains, while nano-SiC is responsible for the great lattice distortion in the SiC-doped samples. The superconductivity transition is advanced in Mg excess samples according to the decreased $\Delta T_c$. The EPC analysis shows that the excess Mg can also improve the electron-$E_{2g}$ coupling. However, the impurity phases depress the total EPC strengths.

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