MgB$_2$ coated superconducting tapes with high critical current densities fabricated by hybrid physical–chemical vapor deposition

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Abstract

The MgB$_2$ coated superconducting tapes have been fabricated on textured Cu (0 0 1) and polycrystalline Hastelloy tapes using coated conductor technique, which has been developed for the second generation high temperature superconducting wires. The MgB$_2$/Cu tapes were fabricated over a wide temperature range of 460–520 °C by using hybrid physical–chemical vapor deposition (HPCVD) technique. The tapes exhibited the critical temperatures ($T_c$) ranging between 36 and 38 K with superconducting transition width ($\Delta T_c$) of about 0.3–0.6 K. The highest critical current density ($J_c$) of $1.34 \times 10^5$ A/cm$^2$ at 5 K under 3 T is obtained for the MgB$_2$/Cu tape grown at 460 °C. To further improve the flux pinning property of MgB$_2$ tapes, SiC is coated as an impurity layer on the Cu tape. In contrast to pure MgB$_2$/Cu tapes, the MgB$_2$ on SiC-coated Cu tapes exhibited opposite trend in the dependence of $J_c$ with growth temperature. The improved flux pinning by the additional defects created by SiC-impurity layer along with the MgB$_2$ grain boundaries lead to strong improvement in $J_c$ for the MgB$_2$/SiC/Cu tapes. The MgB$_2$/Hastelloy superconducting tapes fabricated at a temperature of 520 °C showed the critical temperatures ranging between 38.5 and 39.6 K. We obtained much higher $J_c$ values over the wide field range for MgB$_2$/Hastelloy tapes than the previously reported data on other metallic substrates, such as Cu, SS, and Nb. The $J_c$ values of $J_c$(20 K, 0 T) $\sim$5.8 $\times$ 10$^6$ A/cm$^2$ and $J_c$(20 K, 1.5 T) $\sim$2.4 $\times$ 10$^5$ A/cm$^2$ is obtained for the 2-$\mu$m-thick MgB$_2$/Hastelloy tape. This paper will review the merits of coated conductor approach along with the HPCVD technique to fabricate MgB$_2$ conductors with high $T_c$ and $J_c$ values which are useful for large scale applications.

Keywords: MgB$_2$ film, Cu (0 0 1) tape, Hastelloy tape, HPCVD, Coated conductor
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1. Introduction

The exciting magnesium diboride (MgB$_2$) superconductor is a promising material for superconducting magnets and microelectronic devices at low cost because of its remarkably high critical temperature ($T_c$) of 39 K [1,2], which is 2–4 times higher than those of conventional metallic superconductors, such as Nb-Ti ($T_c = 9$ K) and Nb$_3$Sn ($T_c = 18$ K). The strongly linked nature of the intergrains [3] with a high charge carrier density [4] in this material makes it further an attractive candidate for the next generation of superconductor applications. The practical application of any superconductor strongly depends on its capacity for carrying loss-free currents in high magnetic fields [5]. Therefore, MgB$_2$ superconductors in the form of wires and tapes with high critical current density ($J_c$) are required to be produced in long lengths to replace the conventional metallic superconductors [6,7]. Enormous efforts have been devoted to the fabrication and characterization of MgB$_2$ conductors, such as wires and tapes by the powder-in-tube (PIT) process with encouraging results [8–10]. Already some companies have commercially produced MgB$_2$ wires from conventional powder by using PIT method [11,12]. But the PIT-processed conductors are facing the problems of poor self-field and in-field $J_c$ values of the order of $10^5$–$10^6$ A/cm$^2$ and low packing density of MgB$_2$ (below ~50%) [13,14]. In contrast to PIT bulks and conductors, there are several reports on MgB$_2$ thin films with very high upper critical field ($H_{c2}$), high self-field $J_c$ of $\sim 10^7$ A/cm$^2$ and high $J_c$ in magnetic fields [15–17]. These high $J_c$ values obtained in these films formed by vapor deposition techniques demonstrate the potential for further improving the current carrying capabilities of MgB$_2$ wires and tapes. However, there are only few reports on the MgB$_2$ conductors made by MgB$_2$ film fabrication processes [18–23].

In this article, we report our recent systematic investigation on the fabrication of MgB$_2$ coated superconducting tapes by depositing MgB$_2$ films on textured Cu (0 0 1) and polycrystalline Hastelloy tapes by hybrid physical–chemical vapor deposition (HPCVD) technique. First, we will briefly describe the techniques used for the fabrication of MgB$_2$ wires and tapes. Then we describe the HPCVD technique used for the growth of MgB$_2$ coated conductors. The effect of growth temperature on the microstructure and superconducting properties of MgB$_2$/Cu (0 0 1) superconducting tapes is investigated and discussed. Then we will show how the tape performances can be improved by creating additional defects in the MgB$_2$ films by SiC-coated Cu tapes. The results on MgB$_2$/Hastelloy superconducting tapes are also presented. Finally, we will conclude our results.

2. Methods for the fabrication of MgB$_2$ conductors

A number of techniques have been developed to fabricate MgB$_2$ wires and tapes with high critical current densities. The following section describes those methods briefly.

2.1. Diffusion method

Soon after the discovery of superconductivity in MgB$_2$, the first MgB$_2$ wires fabrication is demonstrated by Canfield et al. by exposing boron filaments to Mg vapor [24]. These wires showed exceptionally low resistivity, high density and high $T_c$ of 39.4 K. The $J_c$ of the order of $5\times10^5$ A/cm$^2$ at 5 K and self-field is obtained for these wires. DeFouw et al. report the synthesis of superconducting MgB$_2$ fibers within a magnesium matrix by infiltration of a B fiber preform with liquid Mg and subsequent in situ reaction at 950 °C [25]. The MgB$_2$ fibers exhibit a critical temperature slightly above 39 K and a critical current density of 360 kA/cm$^2$ at 5 K and self-field is obtained for these wires. DeFouw et al. report the synthesis of superconducting MgB$_2$ fibers within a magnesium matrix by infiltration of a B fiber preform with liquid Mg and subsequent in situ reaction at 950 °C [25]. The MgB$_2$ fibers exhibit a critical temperature slightly above 39 K and a critical current density of 360 kA/cm$^2$ at 5 K and self-field. Diffusion method was also used for the synthesis of MgB$_2$ tapes. Togano et al. fabricated a B/Fe–Mg composite tape and obtained a high-density MgB$_2$ layer by the diffusion reaction between the B plate and a Fe–Mg alloy sheet [26].

2.2. Powder-in-tube process

The powder-in-tube (PIT) process is one of the most commonly used methods for the fabrication of MgB$_2$ wires and tapes. The PIT process is classified into two different processes: in situ and ex situ. The former employs powder
starting materials [27], whereas the latter employs a powder of synthesized superconducting material [28]. The superconducting properties of PIT-processed MgB$_2$ conductors are very sensitive to the quality of the starting powder [29–33], particle size [34–36], porosity of the MgB$_2$ core [37,38], heat treatment temperature [34,39–41], and the sheath material [27,42]. In general, ex situ-processed MgB$_2$ tapes show lower $J_c$ than in situ-processed tapes. Recently, Nakane et al. succeeded in the fabrication of high $J_c$ ex situ MgB$_2$/Fe tapes using MgB$_2$ cores removed from in situ MgB$_2$ tapes [43]. The key factor to obtain a high $J_c$ for an ex situ tape is the high quality of the MgB$_2$ starting powder. The highest $J_c$ values obtained so far at 4.2 K and 10 T is 27 kA/cm$^2$ and at 20 K and 5 T is 10 kA/cm$^2$.

Great progress has been made [44,45] and is going on for the fabrication of MgB$_2$ wires and tapes by PIT process to demonstrate the potential of MgB$_2$ superconductor in practical applications. However, the main problem of the PIT-processed conductors is the low packing density of MgB$_2$ (below ~50%). A further improvement in current carrying capability could be expected if the packing density of MgB$_2$ conductors can be increased.

2.3. Coated conductor technique

The coated conductor technique was basically developed for the cuprates high-$T_c$ superconductors. For example, Y$_{1−x}$Ba$_x$Cu$_2$O$_{7−δ}$ (Y-123) compound, where Y-123 thin films are deposited on various buffer layers on a metallic substrate [46], the highly textured Y-123 is required for high critical current density in Y-123 coated conductor due to the poor superconducting coupling across the grain boundaries of this material [47]. A schematic representation of coated conductor technique is shown in Fig. 1a [48].

The fabrication of MgB$_2$ superconducting films on flexible metallic substrates using the coated conductor approach was firstly demonstrated by Komori et al. [18]. First, they prepared 1 µm thick yttria-stabilized-zirconia (YSZ) buffer layer on Hastelloy (C-276) tapes by bias sputtering. Subsequently, MgB$_2$ films were deposited on YSZ coated Hastelloy tapes using a KrF excimer laser with 400 mJ/pulse operating at 5 Hz. MgB$_2$ superconducting tape showed very high in-field transport critical current density of the order of $1.1 \times 10^5$ A/cm$^2$ at 4.2 K and 10 T. Ferrando et al. succeeded in fabricating carbon-alloyed MgB$_2$ coated conductors on round SiC fibers with a tungsten core by using HPCVD technique [21]. The diameter of the SiC fiber is 100 µm with a tungsten core of 10 µm. The carbon-alloyed fibers show a high upper critical field of 55 T at 1.5 K and a high irreversibility field of 40 T at 1.5 K. From economical and practical application point of view, the SiC fiber used in this work is not the desirable substrate for making wires because it is brittle and costly. MgB$_2$ coatings can also be made on inexpensive and strong metallic wires and tapes such as stainless steel and copper.

Abe et al. reported the growth of MgB$_2$ films on metallic stainless steel substrates using the molten-salts electroplating (MSEP) technique [22]. Transport measurements have shown that the electroplated MgB$_2$ films possess an upper critical field, $H_{c2}(0)$ of 28 T, an irreversibility field, $H_{irr}(0)$ of 13 T, and a critical current density, $J_c$ at (5 K, 0 T) of 25000 A/cm$^2$, which are comparable with the PIT-processed MgB$_2$ tapes. The MgB$_2$ thin films deposited by electron beam evaporation technique on polycrystalline Cu plates by Masuda et al. [23] showed high $J_c$ of $1 \times 10^7$ A/cm$^2$ at 4.2 K and 1 T.
These results for MgB₂ with high upper critical field and high critical current densities demonstrate great potential of MgB₂ coated conductors for superconducting magnets. MgB₂ coated conductors also promise lower cost [49].

3. HPCVD for MgB₂ coated conductors

The film growth of MgB₂ superconductors has been achieved using molecular beam epitaxy (MBE) [50–52], pulsed laser deposition (PLD) [2,53–55], hybrid physical–chemical vapor deposition (HPCVD) [56–58], electron beam evaporation (EBE) [59–61], and the modified reactive method [62,63] etc. Among all these techniques, HPCVD has been proved to be the most effective one for the fabrication of high-quality MgB₂ films with superior superconducting properties [58,64]. For MgB₂ film growth, HPCVD process uses diborane (B₂H₆) as the boron precursor gas, but unlike conventional CVD, which only uses gaseous sources, heated bulk magnesium pieces are used as the Mg source in the deposition process. Since the process involves chemical decomposition of precursor gas and physical evaporation of metal bulk, it is named as hybrid physical-chemical vapor deposition.

Our HPCVD system [57], which is similar to that reported by Zeng et al. [56], consists of a vertical quartz tube reactor having a bell shape (inner diameter: 65 mm, height: 200 mm), gas inlet and flow control system, pressure maintenance system, temperature control system, gas exhaust and cleaning system, inductively coupled heater, and a load-lock chamber. Fig. 2a shows the schematic of the HPCVD system. The main process chamber is connected with the load-lock chamber on the bottom and an induction heater is attached on the outer periphery of the process chamber. A susceptor made from stainless steel is placed coaxially inside the reactor as shown in Fig. 2b. There is a gate-valve between the process chamber and the load-lock chamber. This valve protects the process chamber from contamination during the synthesis of MgB₂ films. HPCVD is capable of producing high quality MgB₂ films free of impurities, such as MgO and oxygen. In PIT-processed conductors, MgO is the main obstacle for the superconducting current transfer from one MgB₂ grain to adjacent grains and results in lower \( J_c \) values [65,66].

The critical current density \( (J_c) \) of the order of \( 10^7 \, \text{A/cm}^2 \) at 5 K in a zero magnetic field is obtained for HPCVD MgB₂ films [67,68]. Zhuang et al. [69] report the observation of a very high \( J_c(0) \) of \( 1.6 \times 10^8 \, \text{A/cm}^2 \) by the transport measurement in a clean-limit MgB₂ film grown by HPCVD, the highest self-field critical current density observed so far in MgB₂. These high \( J_c \) values in HPCVD films suggest that a significant improvement in current carrying capability is possible in MgB₂. In addition, strong enhancement of critical current density [70] and record high values of upper critical field \( (H_{c2}) \) over 60 T [71] have been reported for carbon-doped MgB₂ films produced by HPCVD. Moreover, the deposition of MgB₂ films on the metallic substrates, such as stainless steel (SS), Nb and Cu have been demonstrated by Chen et al. using HPCVD [72]. In MgB₂ superconductor, it is well known that the grain boundaries are transparent to current flow [3]. In addition, grain boundaries in MgB₂ are proven to be the effective flux-pinning centers for enhancing \( J_c \) [60]. This is very advantageous compared to cuprates high-\( T_c \) superconductors where grain alignment is essential to improve grain coupling and to obtain high \( J_c \) values.

![Fig. 2. (a) Schematic diagram of the HPCVD system. (b) The susceptor on which substrate is mounted on the top. About 6.5g of Mg can be accommodated around the substrate holder, good enough to maintain high Mg vapor pressure during deposition.](image-url)
Therefore, the direct deposition of MgB₂ onto the substrate is possible and can avoid the necessity of buffer layers to reduce the cost as well as the complexity of the overall deposition process (Fig. 1b). The micron-thick polycrystalline MgB₂ films on stainless steel substrate and Nb sheets have been reported by Li et al. [73] and Zhuang et al. [74], respectively using the same method of HPCVD. The polycrystalline MgB₂ films on flexible YSZ substrates deposited by HPCVD has been reported by Pogrebnyakov et al. [75] with $T_c$ of 39 K and self-field $J_c$ over $10^7$ A cm⁻². The repeated bending over a radius of 10 mm does not change the superconducting properties of the films including $T_c$ and $J_c$. These good superconducting and mechanical properties are important for MgB₂ coated-conductor wires. Therefore, the impact of HPCVD technique along coated conductor approach is of great interest for the fabrication of MgB₂ tapes and wires.

4. Results on MgB₂/Cu (0 0 1) superconducting tapes

In order to fabricate long length MgB₂ tapes and wires with high critical current densities the selection of substrate material is very important [27,42]. In PIT process, the Fe sheath material most commonly used for the fabrication of MgB₂ tapes and wires has low thermal conductivity and is not suitable from the aspect of the high thermal stability of the tapes and wires. Superconducting MgB₂ wires and tapes have also been fabricated using Cu sheaths by PIT method [27,76–81]. Here we selected textured Cu (0 0 1) tape as a substrate due to its high thermal and electrical conductivity. Also Cu was found to be very helpful in the accelerated formation of the MgB₂ phase at low temperatures [82]. Although Cu reacts with Mg or MgB₂ [77] at high temperatures, in contrast to magnetic materials like Fe and Ni, this reaction can be reduced by lowering the growth temperature.

4.1. Fabrication of MgB₂/Cu tapes

In previous studies the growth of MgB₂ films on Cu substrates have been reported [83,84], whereas there is no report on the deposition of MgB₂ films on textured Cu (0 0 1) tapes. The textured Cu tapes have been prepared by one hour heat treatment of Cu ingot in a tube furnace at 900 °C under a flow of a mixture of argon and hydrogen (5%) gases [85]. The thickness of as-prepared Cu tapes was ~90 μm and the tapes used in this study were cut into sizes of $10 \times 10$ mm².

The MgB₂ coated superconducting tapes were fabricated by depositing MgB₂ films on Cu tapes using the HPCVD process. In this process, the Cu (0 0 1) tape was placed on top surface of a susceptor and Mg solid pieces were placed around it. For the deposition of MgB₂ film on a Cu tape, the reactor was firstly evacuated to a base pressure of ~10⁻³ Torr using rotary pump and purged several times by flowing high purity Ar and H₂ gases. Prior to the film growth, the susceptor along with Cu tape and Mg pieces were inductively heated towards the set temperature under a fixed reactor pressure in ambient H₂ gas. Upon reaching the set temperature, a boron precursor gas, B₂H₆ (5% in H₂) was introduced into the reactor to initiate the film growth. The flow rates were 50 sccm for the H₂ carrier gas and 5 sccm for the B₂H₆/H₂ mixture. Finally, the deposited MgB₂/Cu tape was cooled down to room temperature in a flowing H₂ carrier gas. The MgB₂/Cu tapes were fabricated in the temperature range of 460 to 520 °C. The thickness of all fabricated MgB₂ tapes was ~1.3 μm with a growth rate of 65 nm min⁻¹.

4.2. XRD analysis

The X-ray diffraction (XRD) patterns of pure Cu (0 0 1) tape and MgB₂ superconducting tapes fabricated at various temperatures are shown in Fig. 3. Only the (0 0 l) peaks of MgB₂ are observed, indicating a c-axis-oriented crystal structure normal to the Cu (0 0 1) tape. The peaks of Cu tape are marked with asterisk (*). The MgCu₂ impurity phase is detected in all samples which come from the reaction between Mg and Cu tape during the film deposition process. It is noted that the amount of MgCu₂ impurity phase increases with increasing the growth temperature, implying that Cu is more reactive with Mg at higher temperatures.
Fig. 3. X-ray diffraction patterns of textured Cu (0 0 1) tape and MgB$_2$ films grown on Cu (0 0 1) tapes at temperatures of 460–520 °C.

4.3. Temperature versus resistance curves

The temperature dependences of the normalized resistances for MgB$_2$ films on textured Cu tapes are plotted in Fig. 4. The growth conditions, the superconducting transition temperature $T_c$, and $T_{c,zero}$ of MgB$_2$ superconducting tapes are summarized in Table 1. The tapes exhibited critical temperatures ranging between 36 and 38 K with superconducting transition width ($\Delta T_c$) of about 0.3–0.6 K. Among our samples, the tape fabricated at a temperature of 500 °C has the highest $T_c$ of 37.5 K, which is comparable with the data reported by Li et al. [84] and much higher than that obtained for the MgB$_2$ films on Cu plates by Masuda et al. [23]. It is observed that the $T_c$ of the MgB$_2$ tapes decreased as we lowered the growth temperature. However, the tape prepared at a temperature of 520 °C has lower $T_c$ than the other tapes but shows a sharp superconducting transition.

Table 1

| $T_s$ °C | $P$ (Torr) | $B_2H_6$ (sccm) | $H_2$ (sccm) | $T_c$ (K) | $T_{c,zero}$ (K) |
|---------|------------|-----------------|--------------|-----------|------------------|
| 460     | 2          | 5               | 50           | 37.0      | 36.4             |
| 480     | 5          | 5               | 50           | 37.2      | 36.7             |
| 500     | 10         | 5               | 50           | 37.5      | 37.2             |
| 520     | 20         | 5               | 50           | 36.7      | 36.3             |

4.4. SEM analysis

Fig. 5 shows the surface morphologies of MgB$_2$ tapes observed by scanning electron microscopy (SEM). The MgB$_2$ grains are well connected for the tapes deposited at low temperatures of 460 and 480 °C, as can be seen in Fig. 5a and b. The hexagonal crystallites of about 300–500 nm in size were observed for the tape grown at 460 °C. However, the grain sizes of the tapes grown at high temperatures of 500 and 520 °C are irregular and the connectivity between the MgB$_2$ grains is very poor. Fig. 5c and d, reflects the severe reactions of Mg vapor with Cu tape such as eutectic melting or diffusive movement between Mg vapor and Cu tape [86].
4.5. Effect of growth temperature on $J_c$

The critical current density ($J_c$) is evaluated from the magnetic hysteresis ($M$–$H$) loops by using the Bean’s critical state model. Fig. 6 shows the magnetic field dependence of the $J_c$ at 5 and 20 K for the MgB$_2$ superconducting tapes grown at various temperatures from 460 to 520 °C. Among all the tapes, the highest $J_c$ of $1.34 \times 10^5$ A/cm$^2$ at 5 K under 3 T is obtained for the tape prepared at a temperature of 460 °C. The $J_c$ of our tape over the wide field range is higher than the previous reported data for the film prepared on metallic Cu substrate by short-time sintering method [83] and the films deposited on polycrystalline Cu substrates by electron beam evaporation technique and HPCVD method [23,72]. Furthermore, on increasing the growth temperature a decrease in $J_c$ under applied field was observed for MgB$_2$ tapes. In 3 T field, the $J_c$ of the MgB$_2$ tape grown
at 460 °C is about 2.6 times higher than that grown at 520 °C. This could be attributed to the high density of grain boundaries for the tape grown at 460 °C, which may act as effective flux pinning centers. Another reason for the decrease of $J_c$ of 520 °C sample is most likely due to increase of the extent of MgCu$_2$ impurity phase as we observed in XRD. The presence of these impurity phases reduces the density of superconducting MgB$_2$ and also acts as an obstacle against the flow of supercurrent [70,87]. Therefore, the tape grown at 460 °C shows higher $J_c$ than that grown at 520 °C.

4.6. Flux pinning and SiC doping

The self-field $J_c$ values as high as in the HPCVD MgB$_2$ films are obtained for the HPCVD MgB$_2$/Cu tapes. However, $J_c$ of pristine MgB$_2$/Cu tapes is decreased with an increasing external magnetic field because of their poor flux pinning properties [88]. To increase the potential use of MgB$_2$ superconductor in the practical applications, enhancement in the critical current density under high magnetic fields is essential. In these applications, strong pinning of vortices is an important requirement for an operation with low noise and lower power loss. Basically, the Lorentz force, $F_L = J \times B$, causes vortices to move in the superconductor, where $J$ is the applied current density and $B$ is the average magnetic induction inside the superconductor. If we understand the optimal pinning mechanism, we can obtain a higher $J_c$ because pinning sites, such as point defects, planar defects, etc., block the flux flow against the Lorentz force. The flux pinning force density ($F_p$) is defined as the product of the critical current density ($J_c$) and the induced magnetic field ($B$) in the sample, $F_p = -J_c \times B$. Pinning of vortices can be achieved by creating pinning sites, such as point defects [89], planar defects [90] and impurities [5,87].

Many researchers have attempted to improve the flux pinning behavior through several types of processes, such as proton [91] and neutron irradiation [92,93], oxygen alloying of MgB$_2$ thin films [16] and chemical doping using different metallic and nonmetallic phases, and nanoparticle admixing [94–96]. These studies have shown that chemical doping may be an effective and feasible approach for increasing critical current density of MgB$_2$ superconductors. Therefore, it is necessary to study the doping effect of suitable elements in MgB$_2$. Out of several dopants, carbon C [97] and carbon containing compounds, such as SiC [98], C and CaCO$_3$ co-addition [99] have been reported to be effective for improving the flux pinning properties of MgB$_2$. Employing HPCVD technique, so far carbon has been doped in MgB$_2$ films, using (MeCp)$_2$Mg [100] and methane [101] as the carbon source. Recently, we report the doping of Cu and Ag in MgB$_2$ films by two step method [102]. Firstly, amorphous Cu and Ag-impurity layers were deposited on c-cut Al$_2$O$_3$ substrates at room temperature by pulsed laser deposition (PLD) system. Subsequently, MgB$_2$ films were grown on top of Cu or Ag/Al$_2$O$_3$ substrates at 560 and 600 °C by using HPCVD. It was found that both Cu and Ag are effective for improving the flux pinning property due to creation of extra pinning sources in MgB$_2$ films. In this work, the amorphous SiC layer which is grown on metallic Cu (0 0 1) tape is used as an impurity for creating additional pinning sites in MgB$_2$ [103].

The PLD system was used for the deposition of SiC-impurity layers. The amorphous SiC-impurity layers of thickness 20 nm were deposited on the textured Cu tapes by PLD at room temperature with a background pressure of $\sim 10^{-6}$ Torr. Laser beams were generated using a Lambda Physik KrF excimer laser ($\lambda = 248$ nm). The pulse energy was set at 300 mJ with a repetition rate of 8 Hz. The SiC target (99.99% purity) was used for the deposition of impurity layers. The PLD system used in this study is described in more detail [104,105]. After depositing the SiC-impurity layers on the Cu tapes, we used HPCVD system for the fabrication of MgB$_2$ coated superconducting tapes by growing MgB$_2$ films on SiC/Cu tapes.

In this process, the Cu tape coated with SiC-impurity layer was placed on top surface of a susceptor and Mg chips were placed around it. For the deposition of MgB$_2$ film on a SiC/Cu tape, the reactor was firstly evacuated to a base pressure of $\sim 10^{-3}$ Torr using rotary pump and purged several times by flowing high purity argon and hydrogen gases. Prior to the film
growth, the susceptor along with SiC/Cu tape and Mg chips were inductively heated towards the set temperature under a reactor pressure of 50 Torr in ambient H₂ gas. Upon reaching the set temperature, a boron precursor gas, B₂H₆ (5% in H₂) was introduced into the reactor to initiate the film growth. The flow rates were 90 sccm for the H₂ carrier gas and 10 sccm for the B₂H₆/H₂ mixture. Finally, the fabricated MgB₂/SiC/Cu tape was cooled down to room temperature in a flowing H₂ carrier gas. Since the critical properties, such as T_c, J_c, and Hc₂ of superconductors are strongly influenced by heat treatment conditions. Therefore, the performance of any superconductor can be improved further by optimizing the growth temperature. Thus, it is necessary to investigate the heat treatment effect on the J_c–H behavior of MgB₂/SiC/Cu tapes. The MgB₂/SiC/Cu tapes were fabricated in the temperature range of 460 to 600 °C. A pure MgB₂/Cu tape was also prepared at a temperature of 460 °C for comparison. The thickness of all fabricated MgB₂ tapes was ~2 μm with a growth rate of 0.2 μm min⁻¹.

The J_c (H) curves for MgB₂/Cu and MgB₂/SiC/Cu tapes as a function of growth temperature measured at 5 and 20 K are shown in Fig. 7a and b, respectively. We can see that both at 5 and 20 K, the self-field J_c for SiC-coated MgB₂ tapes is higher than that of pure MgB₂ tape. It is well established that SiC doping in MgB₂ using the conventional in situ PIT technique shows a decrease in T_c and reduction of the self-field J_c due to the high level of C substitution on the B sites [98,106]. Recently Li et al. obtained enhancement in the self-field J_c for SiC-MgB₂ composite caused by thermal strain on the interface between SiC and MgB₂ during the diffusion process, when no chemical reaction between SiC and MgB₂ was observed [107]. No reduction of the T_c and the absence of any chemical reaction between SiC-impurity layer and MgB₂ might be the reason for improvement in the self-field J_c of our MgB₂/SiC/Cu tapes. Furthermore, on increasing the deposition temperature an increase in J_c under applied field was observed for SiC-coated tapes. The optimal growth temperature for the strong improvement in J_c was observed to be 540 °C. These results are in contrast to our previously reported data for MgB₂/Cu tapes, where MgB₂ films were directly deposited on the textured Cu tapes and they showed a decrease in J_c with increasing growth temperature [88]. The enhanced J_c for MgB₂/SiC/Cu tapes could be due to the improved flux pinning by grain boundaries along with the additional defects created by SiC-impurity layer which act as effective flux pinning centers [108]. It is also noted that the in-field J_c of MgB₂/Cu tape around 4 and 2 T at 5 and 20 K, respectively is comparable with the MgB₂/SiC/Cu tapes. This might be due to the
smaller grain size of 350 nm for MgB$_2$/Cu tape as a result this sample has a high density of grain boundaries, and grain boundaries are known to be the main pinning source in MgB$_2$ superconductor [109].

Fig. 8 shows the magnetic field dependences of the flux pinning force densities ($F_p$) at 5 K for MgB$_2$/Cu and MgB$_2$/SiC/Cu tapes grown at various temperatures, where $F_p$ is derived from the corresponding $J_c$ ($H$) data (Fig. 7a). The $F_p$–$H$ graph clearly shows an increase in the flux pinning force density with increasing growth temperature for SiC-coated MgB$_2$ tapes as compared to MgB$_2$/Cu tape. The largest enhancement of $F_p$ is observed for the MgB$_2$/SiC/Cu tape fabricated at 540 °C which has a maximum $F_p$ value, about 5.4 times higher than the MgB$_2$/Cu tape. It indicates that the MgB$_2$/SiC/Cu tapes have higher density of pinning centers, such as grain boundaries and additional defects as a consequence of SiC-impurity layer as compared to MgB$_2$/Cu tape.

5. Results on thick MgB$_2$/Hastelloy superconducting tapes

The role of substrate material is very crucial in the fabrication of MgB$_2$ conductors to obtain high $J_c$ values. The $J_c$ of our MgB$_2$/Cu tapes decreased with an increasing applied field because of their poor flux pinning as well as high reactivity of Mg with Cu tape, as a result, reduction of the superconducting cross sectional area and hence the decrease of $J_c$ values [88]. Thus, we need to use a metallic substrate which does not react with Mg and B. Komori et al. obtained very high in-field transport critical current density for MgB$_2$ grown on YSZ buffered Hastelloy tapes without any chemical reactions between Mg and Hastelloy tapes [18]. In addition, Hastelloy tape has good material properties, such as good flexibility, good conductivity, and lower corrosiveness. Thus, Hastelloy tape could be expected as a good metallic substrate for MgB$_2$ tape conductor fabrications.

For coated conductor applications, it is important to have a thick superconducting film at high growth rate [48, 110], so that the high critical current density ($J_c$) translates to high engineering critical current density ($J_e$) of a conductor [85]. To see the effect of film thickness and deposition rate on the superconducting properties, mainly on $J_c$ of MgB$_2$ superconductor a study is needed to be performed. Wang et al. [111] and Hanna et al. [112] have reported the correlation between film thickness and critical current density for the MgB$_2$ films fabricated on single-crystal MgO (111) and (0001) 6H-SiC substrates, respectively by ex-situ methods. In order to use MgB$_2$ for coated-conductor applications, a clear idea about the relationship between film thickness and critical current density for the MgB$_2$ films fabricated on metallic substrates is necessary. Recently, we have investigated the influence of B$_2$H$_6$ gas mixture flow rate and deposition time on superconducting properties of MgB$_2$ films deposited on metallic Hastelloy tapes by HPCVD [113].

5.1. Fabrication of thick MgB$_2$/Hastelloy tapes

In this work, we use the polycrystalline Hastelloy tapes of thickness ~45 μm and the tapes were cut into sizes of 10 × 10 mm$^2$. The MgB$_2$ coated superconducting tapes were prepared by growing MgB$_2$ films on Hastelloy tapes with different B$_2$H$_6$ gas mixture flow rates and for different deposition times. All tapes were fabricated at a temperature of 520 °C under
Table 2

The growth conditions, thickness, $T_c$, and $T_{c,\text{zero}}$ of MgB$_2$/Hastelloy tapes fabricated at a temperature of 520 °C under a pressure of 200 Torr.

| $H_2$ (sccm) | $B_2H_6$ (sccm) | Deposition time (min.) | Thickness ($\mu$m) | $T_c$ (K) | $T_{c,\text{zero}}$ (K) |
|-------------|----------------|-----------------------|-------------------|---------|-------------------|
| 90          | 10             | 10                    | 0.5               | 38.5    | 38.1              |
| 80          | 20             | 10                    | 1.5               | 38.8    | 38.2              |
| 70          | 30             | 10                    | 2.0               | 38.6    | 38.2              |
| 90          | 10             | 30                    | 1.6               | 38.9    | 38.6              |
| 80          | 20             | 30                    | 4.5               | 39.6    | 39.1              |
| 70          | 30             | 30                    | 6.0               | 39.3    | 39.0              |

a pressure of 200 Torr. The $B_2H_6$ gas flow rate was varied from 10 to 30 sccm at $H_2$ of 90–70 sccm, respectively under different deposition times as summarized in Table 2. For the deposition of MgB$_2$ film on Hastelloy tape, the reactor was firstly evacuated to a base pressure of $\sim 10^{-3}$ Torr using rotary pump and purged several times by flowing high purity Ar and $H_2$ gases. Prior to the film growth, the susceptor along with Hastelloy tape and Mg chips were inductively heated towards the set temperature in ambient $H_2$ gas. Upon reaching the set temperature, a hydrogen-diluted $B_2H_6$ (5% in $H_2$) gas was introduced into the reactor to initiate the film growth. Finally, the fabricated MgB$_2$/Hastelloy tape was cooled down to room temperature in a flowing $H_2$ carrier gas. The MgB$_2$ layer thicknesses range between 0.5 and 6.0 $\mu$m.

5.2. XRD analysis

Fig. 9 shows the X-ray diffraction patterns for pure Hastelloy tape and MgB$_2$/Hastelloy tapes fabricated at 520 °C under various $B_2H_6$ gas mixture flow rates and deposition times. The MgB$_2$ tapes are polycrystalline in nature, as expected because of the use of polycrystalline Hastelloy tapes. It should be noted that the MgB$_2$ tapes deposited for longer time have preferred orientation along the $(00l)$ plane in addition to diffraction peaks of different MgB$_2$ planes compared to those deposited for shorter time. The $c$-axis lattice parameter was calculated using the $(002)$ reflections of the tapes. For the tapes deposited under various $B_2H_6$ gas mixture flow rates and deposition times with the thicknesses of 0.5, 1.6, 2.0, and 6.0 $\mu$m (Table 2), it is observed to be 3.526, 3.525, 3.524, and 3.521 Å, respectively. The compression of $c$-axis lattice parameter with the increase of MgB$_2$ layer thickness indicates that these tapes are
under tensile strain [114]. Moreover, there is no indication for any chemical reaction between MgB$_2$ and Hastelloy tape, neither any impurity or secondary phases, such as Mg, MgO, and MgB$_4$ were observed.

5.3. Temperature versus resistance curves

The temperature dependences of the normalized resistances for the MgB$_2$/Hastelloy tapes are plotted in Fig. 10. The MgB$_2$ tapes have critical temperature ranging between 38.5 and 39.6 K with superconducting transition width ($\Delta T_c$) of 0.3–0.6 K. From Fig. 10, we can see that there is a little effect of B$_2$H$_6$ gas mixture flow rate on $T_c$, and the MgB$_2$ tapes deposited for longer time have higher $T_c$ than those deposited for shorter time. It indicates that $T_c$ is depending on the thickness of MgB$_2$ layer. The thicker films show higher $T_c$ than the thinner films have been reported for the films.
prepared by the same method of HPCVD [115]. The $T_c$ of our tapes is higher by 2 K than we previously obtained for MgB$_2$/Cu tapes [88], and it is much higher than that reported by Komori et al. of 29 K for post annealed MgB$_2$/YSZ/Hastelloy tapes [18]. The high $T_c$ of our samples is most probably due to high purity of the samples and the tensile strain in the films. The tensile strain is reported to be the possible cause of higher $T_c$ in MgB$_2$ films, and the thicker the films the larger the tensile strain [114].

5.4. Microstructure observations

The plane views and the cross-sectional views of the MgB$_2$/Hastelloy tapes were observed by scanning electron microscopy (SEM) and focused ion beam (FIB) technique, respectively. The surface morphologies and the cross-sectional views of MgB$_2$/Hastelloy tapes fabricated with different B$_2$H$_6$ gas mixture flow rates and for different deposition times are shown in Fig. 11a–f. Grains of various orientations could be seen in the images, suggesting that the MgB$_2$ tapes are polycrystalline, which is consistent with XRD results. In addition, the surfaces of the MgB$_2$ tapes look porous. The tape deposited with high B$_2$H$_6$ flow rate and for longer time, Fig. 11d, results in the tilted and isolated grain growth of few μm in sizes. It indicates that the microstructure degradation occurs for the tapes deposited with high B$_2$H$_6$ flow rate and for longer time. Fig. 11e and f are the cross-sectional views of MgB$_2$/Hastelloy tapes. Porosity was observed at the interfaces of Hastelloy tapes and MgB$_2$ layers whereas the inner areas of the MgB$_2$ layers are composed of dense structure without porosities.

5.5. MgB$_2$ layer thickness dependence of $J_c$

Fig. 12 shows the magnetic field dependence of $J_c$ at 5 and 20 K for MgB$_2$ tapes fabricated with various B$_2$H$_6$ flow rates and deposition times. In order to explain the $J_c(H)$ behavior, we use the MgB$_2$ tape thickness of 0.5, 1.5, 2.0, and 6 μm, as shown in Table 2. On increasing the thickness an increase in $J_c$ was observed for MgB$_2$ tapes. The optimal thickness for the strong improvement in $J_c$ was observed to be 2.0 μm. These results are in contrast to the data reported by Wang et al. [111] and Hanna et al. [112], where they obtained a decrease in $J_c$ with increasing MgB$_2$ film thickness, in the similar trend as observed in YBCO-coated conductors [116]. In our previous study, we report that the thick MgB$_2$ film has higher $J_c$ than the thin one, due to much higher density of pinning centers with weak intercolumnar regions in the thicker film [57]. However, further increasing the MgB$_2$ layer thickness over 2 μm, caused a reduction of $J_c$, the rapid fall of $J_c$ can be seen for the 6-μm-thick MgB$_2$ tape. The decrease of $J_c$ is most likely due to the degradation of the microstructure and the growth of larger grain sizes, as a result, less density of pinning centers. Our MgB$_2$/Hastelloy tapes showed very high $J_c$ values over the wide field range than the previously reported data on other metallic substrates, such as Cu, SS, and Nb [22,23,72,88].

6. Conclusions

In this article, we report our recent results on MgB$_2$ coated superconducting tapes fabricated using coated conductor approach by HPCVD technique. Our HPCVD MgB$_2$ tapes showed self-field $J_c$ values at 5 K of the order of $10^7$ A/cm$^2$ as high as already obtained in the
MgB$_2$ films and 1–2 order higher than those reported for PIT-processed MgB$_2$ conductors. The MgB$_2$/Cu tape fabricated at 460 °C showed the $J_c$ values of $J_c(20~K,~0~T) \sim 5.8 \times 10^6$ A/cm$^2$ and $J_c(20~K,~1.5~T) \sim 2.4 \times 10^6$ A/cm$^2$. Even the MgB$_2$/Hastelloy tape as thick as 6 µm showed high $J_c$ values as compared to MgB$_2$/Cu tapes and exhibited opposite trend in the dependence of $J_c$ with growth temperature. The strong improvement in $J_c$ for the MgB$_2$/SiC/Cu tapes could be attributed to the improved flux pinning by the additional defects created by SiC-impurity layer along with the MgB$_2$ grain boundaries. The MgB$_2$/Hastelloy tapes grown at 520 °C with optimal thickness of 2 µm showed the highest $J_c$ values of $J_c(20~K,~0~T) \sim 5.8 \times 10^6$ A/cm$^2$ and $J_c(20~K,~1.5~T) \sim 2.4 \times 10^6$ A/cm$^2$. The HPCVD MgB$_2$ tapes fabricated at 460 °C showed $J_c$ values of $J_c(20~K,~0~T) \sim 1.9 \times 10^6$ A/cm$^2$ and $J_c(20~K,~1~T) \sim 2.3 \times 10^6$ A/cm$^2$. The HPCVD MgB$_2$ tapes on Hastelloy showed very high $J_c$ values over the wide field range than for those deposited on other metallic substrates, such as Cu, SS, and Nb. In addition, no chemical reaction was observed between Mg and Hastelloy tape. Thus, the Hastelloy tape could be a promising metallic substrate for the fabrication of MgB$_2$ tape conductors for practical applications. These results on our HPCVD MgB$_2$ coated superconducting tapes suggest that practical applications of MgB$_2$ at 20 K in low magnetic fields are promising. Although the self-field and low field $J_c$ values are high in these MgB$_2$ tapes than the existing superconducting wires of NbTi and Nb$_3$Sn, further improvement in the current carrying capability under the applied field is essential to realize the potential of MgB$_2$ superconductors as high-field magnets. Therefore, the main challenge now is to transfer the already achieved high $H_c2$ and $J_c$ values in carbon-doped MgB$_2$ films to the conductor forms which are more suitable for large scale applications.

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