Coercivity Enhancement of Hot-deformed Nd-Fe-B Magnets by the Eutectic Grain Boundary Diffusion Process

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Abstract

Nd-M (M= Al, Cu, Ga, Zn, Mn) alloys with compositions close to eutectic points were investigated as diffusion sources for the grain boundary diffusion process to hot-deformed Nd-Fe-B magnets. Coercivity enhancement was observed for most of the alloys. Among them, the sample processed with Nd₉₀Al₁₀ exhibited the highest coercivity of 2.5 T at room temperature. However, the sample processed with Nd₇₀Cu₃₀ exhibited the highest coercivity of 0.7 T at 200 °C. Microstructural observations using scanning transmission electron microscope (STEM) showed that nonferromagnetic Nd-rich intergranular phase envelops the Nd₂Fe₁₄B grains after the diffusion process. Abnormal grain growth and the dissolution of Al into the Nd₂Fe₁₄B grains were observed in the sample processed with Nd₉₀Al₁₀, which explains its inferior thermal stability of coercivity compared to the sample processed with Nd₇₀Cu₃₀. The coercivity enhancement and poor thermal stability of the coercivity of the Nd₉₀Al₁₀ diffusion-processed sample are discussed based on microstructure studies by transmission electron microscopy.

Keywords: Permanent magnets; Nd-Fe-B; Coercivity; Diffusion process; Grain boundary phase

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1. Introduction

The anisotropic Nd-Fe-B magnets produced by hot-pressing and subsequent hot-deformation of melt-spun isotropic nanocrystalline powder, known as hot-deformed magnets, are comprised of [001]-textured platelet-shaped Nd₂Fe₁₄B grains of typically ~100 nm in thickness and 300-400 in width [1,2]. The strong (001) texture develops during the hot-deformation or upsetting process [3,4], which is considered to be due to the orientation dependent grain growth of tetragonal Nd₂Fe₁₄B phase [4-7] or the anisotropy of its elastic properties [8]. Owing to the ultrafine grain size that is comparable to the single domain size, the hot-deformed magnets show relatively larger coercivity compared to sintered magnets. The remanent magnetization $\mu_0 M_r$ of 1.4 T reported for a hot-deformed magnet is comparable to that of commercial Nd-Fe-B sintered magnets [9,10], indicating that the overall permanent magnetic properties of hot-deformed magnets are competitive with those of sintered magnets. Considering the fact that the average grain size is one order magnitude smaller than that of sintered magnets, the hot-deformed magnets should have potential for achieving higher coercivity at elevated temperature. DyF₃ and DyCu alloy additions to melt-spun Nd-Fe-B ribbons followed by hot-pressing and hot-deformation were found to be capable of improving coercivity by formation of the thin (Nd,Dy)₂Fe₁₄B layer [11,12]. However, even with the Dy addition, the highest coercivity so far reported for the anisotropic hot-deformed magnets was limited to around 2.0 T.

Lewis *et al.* [13-15] attributed this relatively low coercivity to the exchange coupling between Nd₂Fe₁₄B grains through a ferromagnetic intergranular phase that contains high concentration of Fe and Co. They also pointed out that the intergranular phase does not form along the side surface of the platelet-shaped grains [16]. Based on more recent three-dimensional atom probe (3DAP) study, Liu *et al.* [10] have shown a clear correlation between the Nd concentration in the grain

boundary phase and the coercivity of hot-deformed magnets, suggesting that the modification of the composition of the grain boundary phase may lead to substantial improvement in coercivity.

Sepehri-Amin *et al.* [17] reported the coercivity of hydrogen-disproportionation-decomposition-recombination processed Nd-Fe-B magnet powder can be largely enhanced by infiltrating a low melting temperature Nd₇₀Cu₃₀ eutectic alloy. Thereafter, they applied this eutectic grain boundary diffusion process to hot-deformed magnets and demonstrated that the coercivity is enhanced from 1.4 T to 2.3 T using the 1 mm thick hot-deformed Nd-Fe-B magnets [18]. They have also shown that this enhancement of coercivity mainly stems from the enrichment of Nd in the intergranular phases by the infiltration of Nd-Cu liquid through grain boundaries. This method was later extended to bulk hot-deformed magnets and a balanced combination of coercivity (1.9 T) and remanence (1.35 T) was attained using the expansion constraint eutectic diffusion process [19]. However, a higher level of coercivity of 2.5 T is required for automotive traction motor applications.

Neodymium has eutectic reactions with various types of Nd_xM_y compounds, where M includes Al, Cu, Ga, Zn, Mn, Co, Ni and Fe. However, only a few eutectic systems, Nd-Cu and Pr-Cu, were explored for the eutectic grain boundary diffusion process [18, 20]. Previous studies [21,22] suggested that the addition of Al and Ga improve the wettability between Nd₂Fe₁₄B and intergranular phases, which may smoothen the intergranular phase for better separation of Nd₂Fe₁₄B grains. With partial substitution of Al for Fe [23], Nd₂Fe_{14-x}Al_xB compounds are known to exhibit higher anisotropy field accompanied by the reduction in magnetization.

In order to explore which alloy system will give the highest coercivity among various Nd-M systems, we employed various eutectic alloys as the diffusion source for the grain boundary diffusion process to hot-deformed Nd-Fe-B magnets. The mechanism of the coercivity

enhancement of hot-deformed magnets is discussed on the basis of detailed microstructure characterizations by scanning electron microscopy (SEM) and transmission electron microscopy (TEM).

2. Experimental

Hot-deformed magnets in $4 \times 4 \times 2 \text{ mm}^3$ size with the composition of $\text{Nd}_{13.2}\text{Fe}_{76}\text{Co}_{5.6}\text{B}_{4.7}\text{Ga}_{0.5}$ (at.%) upset processed at 780°C were used as starting samples. Flakes of the Nd-rich eutectic alloys, $\text{Nd}_{90}\text{Al}_{10}$, $\text{Nd}_{82}\text{Al}_{18}$, $\text{Nd}_{80}\text{Ga}_{20}$, $\text{Nd}_{70}\text{Cu}_{30}$, $\text{Nd}_{75}\text{Zn}_{25}$, $\text{Nd}_{50}\text{Pr}_{20}\text{Cu}_{30}$ and $\text{Nd}_{60}\text{Al}_{30}\text{Cu}_{10}$ were fabricated with a single-roller melt-spinning machine. The diffusion process was carried out by wrapping the hot-deformed magnets with eutectic alloy ribbons in Nb foil, annealing at various temperatures for 1 h. The annealing temperature was chosen based on the melting temperature of the diffusion sources. In order to investigate the upper limit of the coercivity enhancement through the diffusion process, no control of the amount of the infiltration was made in this work, *i.e.*, no consideration was made for the loss of remanence.

Magnetic measurements of the hot-deformed magnet and the diffusion-processed magnets were carried out in a superconducting quantum interface device vibrating sample magnetometer (SQUID-VSM) for the specimen sliced crossing sample surface and center. Correction of demagnetization field was made using the equation of $H_{\text{eff}} = H_{\text{ext}} - N_d M$, where H_{eff} denotes as effective field, H_{ext} refers to external field, while N_d is the demagnetization factor which was calculated for the prism specimen configuration described in ref[24]. Overall microstructural characterization was conducted using Carl ZEISS CrossBeam 1540EsB and surfaces of samples were cleaned using Focused Ion Beam (FIB) before SEM observation. Scanning transmission electron microscopy energy-dispersive spectroscopy (STEM EDS) maps were constructed using

Nd- L_{α} , Dy- M_{α} , Fe- K_{α} , Co- K_{α} and Cu- K_{α} spectrum. The TEM specimens were prepared with the lift-out method using a focused ion beam on FEI Nanolab Helios 650.

3. Results

Fig. 1(a) shows the demagnetization curves of the initial hot-deformed magnet and the ones diffusion-processed using various diffusion sources. Large enhancement of coercivity ($\mu_0 H_c$) was realized by the diffusion process using Nd₉₀Al₁₀, Nd₈₂Al₁₈, Nd₈₀Ga₂₀, Nd₇₀Cu₃₀, Nd₇₅Zn₂₅, Nd₅₀Pr₂₀Cu₃₀ and Nd₆₀Al₃₀Cu₁₀ alloys as the diffusion sources. Degradation of coercivity was observed when the diffusion process was applied at 750°C for Nd₇₅Mn₂₅ and Nd₈₂Al₁₈ because of grain coarsening. Remanence ($\mu_0 M_r$) of the diffusion-processed samples decreased as a result of the infiltration of the diffusion sources into the magnet.

Fig. 1(b) summarizes the remanence and the coercivity of the diffusion-processed samples, indicating the infiltration of Nd₉₀Al₁₀ is the most effective in enhancing the coercivity. After the diffusion process at 700°C for 1 h using Nd₉₀Al₁₀, coercivity increased from 1.3 T to 2.5 T, with the remanence decreases from 1.51 T to 1.21 T. Using Nd₇₀Cu₃₀ as a diffusion source, $\mu_0 H_c$ of 2.3 T was obtained at a relatively low diffusion process temperature of 600°C. This low processing temperature for Nd₇₀Cu₃₀ compared to other alloys is due to the low Nd/NdCu eutectic temperature of 520°C.

Fig. 2(a) presents the temperature dependence of the coercivities of the samples diffusion-processed with Nd₉₀Al₁₀ and Nd₇₀Cu₃₀. The Nd₉₀Al₁₀ diffusion-processed sample exhibits faster degradation in coercivity with increasing temperature compared to that of Nd₇₀Cu₃₀ diffusion-processed sample; as a result, Nd₉₀Al₁₀ diffusion-processed sample possesses lower coercivity of

0.6 T than that of the Nd₇₀Cu₃₀ diffusion-processed sample, 0.7 T, at 200°C. Figure 2 (b) shows $H_c(T)/M_s(T)$ plotted as functions of $H_A(T)/M_s(T)$, where $H_c(T)$ is experimentally measured temperature dependent coercivity, and $M_s(T)$ and $H_A(T)$ are the temperature dependent magnetization and anisotropy field for Nd₂Fe₁₄B taken from ref [25]. $H_c(T)/M_s(T)$ shows a near linear correlation against $H_A(T)/M_s(T)$ with α as the slope, N_{eff} as the intercept of the phenomenological micromagnetic equation, $H_c(T) = \alpha H_A(T) - N_{\text{eff}} M_s(T)$. In this model, α reflects decreased anisotropy field induced by the defect regions at grain boundaries as well as grain misalignment while N_{eff} , known as effective demagnetization factor which reflects local demagnetizing field [25]. By linear fit of H_c/M_s vs. H_A/M_s plot, α values derived for Nd₉₀Al₁₀ and Nd₇₀Cu₃₀ diffusion processed samples are 0.79 and 0.61, respectively, substantially increased from 0.34 of the hot-deformed magnet. However, N_{eff} values increase from 0.79 to 1.47 and 0.94 for Nd₉₀Al₁₀ and Nd₇₀Cu₃₀ diffusion-processed samples, respectively. The larger N_{eff} for the Nd₉₀Al₁₀ diffusion processed sample is due the larger grain size because the processing temperature for Nd₉₀Al₁₀ is 100°C higher than that for Nd₇₀Cu₃₀.

Fig. 3(a) and (b) show the backscattered electron (BSE) SEM images of the hot-deformed and Nd₉₀Al₁₀ diffusion-processed samples, respectively, taken from the side cross-sections of the samples (c-axis in-plane upward), while Fig. 3(d) and (e) are the images observed from the top surface of the samples (c-axis out-of-plane). The average grain size of the hot-deformed magnet is $\sim 282 \pm 94$ nm in width and $\sim 72 \pm 22$ nm in thickness. The grain size increased to $\sim 299 \pm 138$ and $\sim 100 \pm 66$ nm respectively for the Nd₉₀Al₁₀ diffusion-processed sample. The average grain size of the Nd₉₀Al₁₀ diffusion-processed sample is comparable to that of the initial hot-deformed magnet. However, the large standard deviation of Nd₉₀Al₁₀ infiltrated sample implies a much

larger distribution in the grain size, which is caused by the high annealing temperature for the $\text{Nd}_{90}\text{Al}_{10}$ diffusion-process. Such abnormally grown grains in the $\text{Nd}_{90}\text{Al}_{10}$ diffusion-processed sample (indicated by the red arrows in Fig.3b and d) have not been observed in the $\text{Nd}_{70}\text{Cu}_{30}$ diffusion-processed sample [18]. These images also show good isolation of $\text{Nd}_2\text{Fe}_{14}\text{B}$ grains by the intergranular Nd-rich phase, which forms not only along the c-plane but also at the side grain boundaries. In comparison, the $\text{Nd}_2\text{Fe}_{14}\text{B}$ grains of the hot-deformed magnets are mostly in contact with each other in both parallel and perpendicular directions to the platelets. The areal fractions of the Nd-rich phase increased from 5 % of the hot-deformed magnet to 27 % of $\text{Nd}_{90}\text{Al}_{10}$ infiltrated sample. With this large volume fraction of Nd-rich intergranular phase, the $\text{Nd}_2\text{Fe}_{14}\text{B}$ grains are expected to be magnetically isolated. Note that two types of contrasts are observed in the intergranular phase; one with a relatively bright contrast and the other has a dim contrast, suggesting non-uniform distributions of Co and Ga in the Nd-rich intergranular phase, the details of which will be described later.

Bright field (BF) transmission microscopy (TEM) and Lorentz TEM images of the hot-deformed and $\text{Nd}_{90}\text{Al}_{10}$ diffusion-processed samples are shown in Fig. 4. Large amount of wavy-patterned strain-induced fringes are observed in the $\text{Nd}_{90}\text{Al}_{10}$ diffusion-processed sample (Fig. 4b), which are not evident in the initial hot-deformed sample. This is believed to originate from the lattice mismatch between the main phase and the crystalline Nd-rich intergranular phase. The over-focused Lorentz-TEM images show domain walls in bright and dark contrasts. The domain wall appears to be continuous across the grain boundaries parallel to the broad surface of the platelet-like grains in the initial hot-deformed sample (Fig. 4c). On the other hand, the domain walls are interrupted in the intergranular phase in the $\text{Nd}_{90}\text{Al}_{10}$ diffusion-processed sample (Fig.4d), suggesting the grain boundary phase is non-ferromagnetic after the diffusion process. In

spite of the exchange decoupling of the grains, the domain walls lie linearly because of the magnetostatic interaction among the grains.

STEM-EDS maps of the hot-deformed and Nd₉₀Al₁₀ diffusion-processed samples are shown in Fig. 5(a) and (b), respectively, with c-axis in-plane. In the initial hot-deformed sample, Nd-rich intergranular phase is not clearly visible. In most cases, Nd₂Fe₁₄B grains are in direct contact with each other; Co shows uniform distribution while Ga appears to be enriched in the intergranular phase. In the Nd₉₀Al₁₀ diffusion processed sample, Fig. 5(b), thick Nd-rich intergranular phase forms parallel to the flat surface of the platelet-shaped Nd₂Fe₁₄B grains. Thinner Nd-rich intergranular phase is observed at the side boundaries; however, many platelet grains appear to be interconnected in the longitudinal direction. Two types of contrasts are observed in the Co, Ga and Al maps, indicating the Nd-rich intergranular phase is separated into two phases, *i.e.*, (Nd,Co)-rich and Nd(Ga,Al)-rich phases.

The chemical composition and the crystal structure of these two different types of Nd-rich phases were examined by localized STEM-EDS quantification and nanobeam electron diffraction (NBED). Fig. 6(a) shows a superimposed elemental map of Nd and Co for visualizing the distribution of the (Nd,Co)-rich phase. The diffraction pattern shown in Fig. 6(b) is taken from the Nd₂Fe₁₄B grain in the [100] zone axis. Fig. 6(d) shows the NBED pattern taken from the Nd-rich phase, which can be indexed as an *fcc* structure in the [013] zone axis. The lattice constant of this type of Nd-rich phase was calculated to be ~0.54 nm, which is consistent with the fcc Nd-rich phase reported in sintered Nd-Fe-B magnets [26,27]. From the NBED pattern taken from the interface region of Nd₂Fe₁₄B phase and the Nd-rich intergranular region, the crystallographic orientation relationship (OR) can be identified to be $[100]_{Nd_2Fe_{14}B} // [013]_{fcc-Nd}$,

$(0\bar{4}6)_{Nd_2Fe_{14}B} // (1\bar{3}1)_{fcc-Nd}$. The localized EDS quantification suggests that the composition of the Nd-rich intergranular phase is $\sim Nd_{79}Co_{14}Fe_6Al_1$ (at.%).

The chemical composition of Nd(Ga,Al)-rich regions shown in Fig. 7(a) shows the chemical composition is about $\sim Nd_{62}Co_{22}Fe_8Al_8$ (at.%) with the amorphous structure (Fig. 7b). A layered contrast is seen in the HAADF-STEM image (Fig. 7d), which comes from the periodical atomic layer stacking of the newly-formed phase, which has the composition of close to $Nd(Fe,Al)_2$ and the diffraction pattern of this phase is shown in Fig. 7(b). The lattice fringe shows a periodicity of 1.2 nm and calculated d-spacings are 3.4 nm and 4.2 nm, respectively. Fidler *et al.* [28-30] referred this phase as $\alpha 2$ or δ phase that was first found in Al_2O_3 -doped sintered magnets and they pointed out that this phase has a complex tetragonal structure. They also suggested that the $\alpha 2$ phase has a large anisotropy field and good wettability with $Nd_2Fe_{14}B$ grains, which was attributed to the high coercivity of Al-doped or Al_2O_3 -doped sintered magnets.

Fig. 8 shows the configuration of the Nd-rich intergranular phase in the direction perpendicular to the flat surface of the $Nd_2Fe_{14}B$ grains. The selected peaks for the elemental maps are Nd L_{α} , Fe K_{α} , Co K_{α} , Ga K_{α} and Al K_{α} . Al is segregated at the interface between the intergranular phase and the $Nd_2Fe_{14}B$ phase. Line-scan analysis shows around 7.5 at. % of Al is segregated at the Nd-rich intergranular phase/ $Nd_2Fe_{14}B$ grain interface and about 3 at.% of Al was detected within the $Nd_2Fe_{14}B$ phase. The Nd-rich intergranular phase in the direction perpendicular to the flat surface of the platelet contains Nd close to 80 at. % with less than 10 at.% Fe, suggesting the intergranular region is non-ferromagnetic.

4. Discussion

This work has shown that the coercivity of hot-deformed magnets can be substantially enhanced by the grain boundary diffusion process using various RE-TM eutectic alloys as diffusion sources. Note that the coercivity of hot-deformed magnets is degraded if the diffusion-process is applied above the temperature for the hot-deformation process. In fact, the hot-deformed magnets processed at 750°C using Nd₇₅Mn₂₅ and Nd₈₂Al₁₈ alloys did not show any improvement in coercivity because of abnormal grain growth at this high processing temperature.

The enhancement of the coercivity is mainly attributed to the exchange decoupling of the Nd₂Fe₁₄B grains as a result of the formation of Nd-rich intergranular phase by the infiltration of eutectic melts. This is evidenced from the microstructural characterizations by SEM and TEM. The concentration of Fe in the intergranular phase is much decreased after the diffusion process, indicating that the ferromagnetism in the grain boundary phase is diluted by the infiltration of the RE-TM eutectic alloy. Nearly complete magnetic isolation has been realized by the diffusion process when Nd₉₀Al₁₀ alloy is used as the diffusion source. The micromagnetic parameters α in $H_c(T) = \alpha H_A(T) - N_{\text{eff}} M_s(T)$ was improved for Nd₉₀Al₁₀ and Nd₇₀Cu₃₀ diffusion-processed samples. Improved α suggests that the defect density at the interface that cause reduction of anisotropy field is reduced by diffusion of Nd-Cu and Nd-Al alloys. The Nd₉₀Al₁₀ diffusion processed sample has a slightly larger α value than that of Nd₇₀Cu₃₀ diffusion processed sample. This can originate from segregation of Al at the interface of Nd-rich phases/Nd₂Fe₁₄B grains as shown in Fig. 8 that may recover the reduced local anisotropy field at the surface of Nd₂Fe₁₄B grains. Also, the Al substitution for Fe atoms is expected to increase anisotropy field of the Nd₂Fe₁₄B phase

[23]. The formation of the α_2 phase in the grain boundary may also have a similar effect. These factors could contribute to the high coercivity of $\text{Nd}_{90}\text{Al}_{10}$ infiltrated sample at room temperature.

The larger N_{eff} value of $\text{Nd}_{90}\text{Al}_{10}$ infiltrated sample is considered to be due to the dispersed grain size distribution after the high processing temperature required for the $\text{Nd}_{90}\text{Al}_{10}$ infiltration. Non-uniform grain size distribution leads to poorer thermal stability as a result of the non-uniform nucleation of reversed magnetic domains due to the larger stray field from those abnormally grown grains [31]. In addition, enrichment of 3 at.% Al in the $\text{Nd}_2\text{Fe}_{14}\text{B}$ phase can reduce the Curie temperature of $\text{Nd}_2\text{Fe}_{14}\text{B}$ phase which can be an additional reason for the poorer thermal stability of coercivity in $\text{Nd}_{90}\text{Al}_{10}$ diffusion-processed sample compared with that of $\text{Nd}_{70}\text{Cu}_{30}$ diffusion-processed sample.

The coercivity of $\text{Nd}_{70}\text{Cu}_{30}$ diffusion-processed sample at 200°C (typical operation temperature of traction motors of hybrid vehicles) is 0.7 T, which is close to 0.8 T of the currently used $(\text{Nd}_{0.7}\text{Dy}_{0.3})\text{-Fe-B}$ sintered magnets. With the development of cooling technique in traction motors, the application of the Nd-Cu diffusion-processed magnets may become practical in the near future.

The remaining issue of the eutectic diffusion process is the optimization of the thickness of the intergranular phase, especially those located parallel to the broad surface of the platelet-shaped $\text{Nd}_2\text{Fe}_{14}\text{B}$ grains. As seen in Fig. 1, large coercivity enhancement was achieved at the expense of remanence because of the large volume fraction of non-ferromagnetic intergranular phase. In order to keep the remanence high, a smaller amount of diffusion source should be infiltrated. For this purpose, the usage of diffusion sources with higher diffusivity with lower melting-point may be effective. The diffusion process using lower melting temperature alloys will also be beneficial in optimizing the grain size distribution of the diffusion-processed magnets

since grain coarsening during the eutectic diffusion process is detrimental for the thermal stability of the hot-deformed magnets.

5. Conclusion

Coercivity improvement has been realized for hot-deformed Nd-Fe-B magnets using Nd-Al, Nd-Cu, Nd-Zn, Nd-Ga, Nd-Cu-Al and Nd-Pr-Cu eutectic alloys as diffusion sources. At room temperature, the sample diffusion processed at 700°C using Nd₉₀Al₁₀ shows the highest coercivity of 2.5 T. However, it exhibits pronounced degradation of coercivity with increasing temperature compared to the sample diffusion-processed with Nd₇₀Cu₃₀. Thus, at 200°C, Nd₇₀Cu₃₀ diffusion-processed sample possesses the highest coercivity of 0.7 T, indicating that Nd-Cu alloy is still the most promising diffusion source for the eutectic grain boundary diffusion process.

Co-enriched Nd-rich phase was determined to be an *fcc* phase which has a good crystallographic match with the Nd₂Fe₁₄B phase. An Al-containing phase (al2) was observed in the Nd₉₀Al₁₀ diffusion-processed sample with the composition of Nd(Fe,Al)₂ and a tetragonal structure.

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Fig. 1. (a) Demagnetization curves of the hot-deformed and diffusion-processed samples by various eutectic alloys, (b) remanence vs. coercivity plot of the hot-deformed and the diffusion-processed samples.

Fig. 2. (a) Temperature dependence of coercivity of the hot-deformed, Nd-Al and Nd-Cu diffusion-processed samples, (b) dependence of H_c/M_s on H_a/M_s of the hot-deformed, Nd-Al and Nd-Cu diffusion-processed samples. Micromagnetic parameters α and N_{eff} are calculated from the slope and intercept of the linear fit line of these plots, respectively.

Fig. 3. BSE-SEM images of the hot-deformed (a,c) and the NdAl-infiltrated (b,d) samples. The observation direction is in-plane (a,b) and out-of-plane in (c,d).

Fig. 4. BSE-SEM images of the hot-deformed (a,c) and the NdAl-infiltrated (b,d) samples. The observation direction is along c-axis in-plane (a,b) and out-of-plane in (c,d).

Fig. 5. TEM-EDS elemental mapping images of the hot-deformed (a) and the NdAl-infiltrated (b) samples with c-axis in-plane upward. The selected analysis peaks for the elemental maps are Nd L_{α} , Fe K_{α} , Co K_{α} , Ga K_{α} and Al K_{α} .

Fig. 6. (a) STEM-EDS elemental mapping images of Fe overlapped with Co maps of NdAl-infiltrated sample. The selected analysis peaks for the elemental maps are Fe K_{α} , and Co K_{α} . (b-d) NBED patterns taken from $\text{Nd}_2\text{Fe}_{14}\text{B}$ phase, interface and intergranular Nd-rich regions, respectively; (e) schematic models of the crystallographic ORs between $\text{Nd}_2\text{Fe}_{14}\text{B}$ phase and the Nd-rich phase; (f) NBED pattern taken from another intergranular Nd-rich grain.

Fig. 7. (a) STEM-EDS elemental mapping images of Ga overlapped with Al maps of NdAl-infiltrated sample. The selected analysis peaks for the elemental maps are Fe K_{α} , and Co K_{α} . (b,c) NBED patterns taken from amorphous and Nd(Fe,Al)₂ intergranular Nd-rich regions, respectively; (d) STEM-HAADF image and EDS maps taken from the Nd(Fe,Al)₂ phase.

Fig. 1

Fig. 2

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