Magnetic Properties of Cluster/Thin-Film Laminated Hybrids of Fe–Pt and Fe–Co Alloys

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Both Fe–Pt cluster/Fe–Co film and Fe–Co cluster/Fe–Pt film laminated hybrids have been prepared by combination of a plasma-gas-cluster-deposition and a helicon-plasma sputter-deposition and studied by X-ray diffraction and magnetization measurements. In Fe–Pt cluster/Fe–Co film laminated hybrids, $M_s$ increases, while $H_C$ and $BH_{MAX}$ monotonically decrease with the Fe–Co film thickness. In Fe–Co cluster/Fe–Pt film laminated hybrids, $M_s$, $H_C$ and $BH_{MAX}$ increase with the Fe–Co cluster layer thickness in comparison with those of simple Fe–Pt films. However, the remarkable improvement of their hard magnetic properties cannot be attained in these laminated hybrids owing to both the low packing density of Fe–Co clusters and the difficulty in the diffusion-control of Fe and Co atoms from Fe–Co to Fe–Pt layers.


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1. Introduction

Monodispersive size transition metal clusters have been prepared in a clean inert gas atmosphere using a plasma gas condensation cluster deposition (PGCCD) system.1–3) Alloy clusters have been also been prepared with adjusting their chemical compositions by setting two different metal plates on the two facing target holders of the single sputtering room of the PGCCD system and by supplying electric powers independently to these targets.3–13) Their chemical homogeneities and structures reflect the equilibrium alloy phase diagrams.14) B2 type ordered phase formations in Co/Al and Fe/Al clusters are consistent with those in the equilibrium phase diagrams, random solid solutions in Co/Pd, Co/Pt, Fe/Pt and Fe/Ni clusters contradictory to L1_0-type ordered phases in the equilibrium phase diagrams, and Ag- and Nb-rich phase formations partly consistent with the complete immiscibility in the equilibrium phase diagram. These clusters form porous nanoparticle assemblies on substrates: their packing densities are 30–40% of those of bulk solids.15–17)

Since superparamagnetism becomes prominent in nanoparticles whose volumes are directly related to the effective magnetic anisotropy, their magnetic applications such as magnetic memory devices have been hindered. L1_0-type ordered fct FePt and CoPt alloy nanoparticles are hopeful candidates for such applications because their high magnetocrystalline anisotropy overcomes the thermal instability of magnetization in magnetic clusters at an ambient temperature. Indeed, L1_0-type ordered FePt and CoPt phases have been obtained in clusters and nanoparticles prepared via vapor and chemical solution routes and post-annealing.6,7,18–21) Mixing of Fe–Pt alloys with various additives and uses of noble metal buffer layers are effective to minimize particle agglomeration, crystal growth and size distribution broadening induced by heat treatments. They reduce the ordering temperatures, but modify morphology of Fe–Pt nanoparticles.22–24) Periodical arrangements (self-organizations) of these particles have been performed via colloidal-solution routes, but their coalescence during annealing processes degrades their magnetic properties.19,20) Though physical isolations of nanoparticles have been achieved by surface coating or matrix embedding,25–27) nonmagnetic materials modify their order degrees and morphologies.

Online annealing (annealing of clusters during the flight period prior to the cluster deposition) has been reported to obtain ordered FePt alloy clusters using high pressure plasma sputtering methods.28–31) They are basically similar to the present PGCCD system where ordered FePt nanoparticles formed in the growth room were ejected through a small orifice and transported through a heating zone. Ordered FePt alloy nanoparticles have been also prepared in a colloidal solution at high reaction temperatures,32) and directly by high pressure plasma sputtering with adjusting the Ar gas flow and concentrating Ar plasma,33) where sputtered atoms gain sufficient energy from energetic Ar ions for both the crystalization and chemical ordering. They reveal coherent rotation-type magnetization curves expected for isolated magnetic particles, however, the low $H_C$ value indicates incomplete ordering of Fe–Pt nanoparticles. Moreover, it has been emphasized that the ordered phase cannot be formed in Fe–Pt nanomaterials whose diameters and/or thicknesses are less than 4 nm.34–36)

Agglomeration, growth and heterogeneous morphology of Fe–Pt nanoparticles induced by heat treatment are serious for their application to magnetic memory devices, but not for the application to spring magnets, in which hard magnetic (high magnetic anisotropy) phases coexist with soft magnetic (high saturation magnetization) phases and keep magnetically good
contacts. Their effective magnetic exchange-coupling reveals superior hard magnets properties, saturation magnetizations and energy products are much higher than the single phase counterparts, magnetic remanences are higher than those expected for isotropic magnets, etc. So far, Fe–Pt particles and Fe particles/Fe–Pt thin film hybrids prepared by cluster deposition methods have been reported to show excellent hard magnetic properties.

In this study, we prepared Fe–Pt cluster/Fe–Co film and Fe–Co cluster/Fe–Pt film laminated hybrids using both a PGCCD system and a helicon-plasma-type-sputter-deposition system and observed their structures and magnetic properties. Based on these result, we are discussing a perspective of these cluster/film laminated hybrids for the application to spring magnets.

2. Experimental Procedures

Alloy clusters were prepared by the PGCCD system (see Fig. 1), which was based on plasma-glow-discharge vaporization and inert gas condensation, being composed of the three main parts: a sputtering chamber, a growth room and a deposition chamber. The back ground pressures of all chambers were evacuated by turbo molecular pumps to be kept at about 1.3 × 10⁻⁵ Pa. During the cluster deposition, flow rates, R, of Ar and He gases were adjusted by mass-flow controllers and steadily introduced into the sputtering chamber through a nozzle, whereas the sputtering chamber and growth room were evacuated by a mechanical booster pump to be kept at 4 × 10⁻¹ Pa and the deposition chamber by a turbo molecular pump to be kept at about 3 × 10⁻¹¹ Pa. We set Fe and Pt metal targets, whose diameter was 80 mm and thickness was 5 mm, face to face and supplied different electric powers to them for adjusting chemical composition of clusters for Fe–Pt alloy cluster preparations. We also set Fe–30 at%Co alloy disc targets for Fe–Co clusters and supplied the same electric powers to them. Atoms sputtered into the inert gas space were decelerated by collisions with inert gas atoms, lost their kinetic energies and collided with each other to form clusters. The mean cluster sizes were controlled by changing the flow rates (the partial pressures) of Ar and He gases. Volumes of cluster layers were estimated for some Fe–Co clusters we chose the following compositions for Fe–Pt cluster assemblies using an energy dispersive X-ray (EDX) analyzer installed in TEM. Morphologies and average chemical compositions were also measured for Fe–Pt and Fe–Co clusters heavily deposited on Si wafers by a scanning electron microscope (FE-SEM, Hitachi Co., S-4700) in which an EDX analyzer was installed. Structures were checked by an X-ray diffractometer (Rigaku 4037) in the Bragg–Bretano geometry with Cu-Kα radiations monochromatized by a graphite crystal. Magnetization curves were measured at room temperature in magnetic fields up to 5 T for cluster/thin film laminated hybrids of Fe–Pt and Fe–Co alloys by a superconducting quantum interface device magnetometer (Quantum Design Co., MPMS-5).

3. Results

3.1 Fe–Pt cluster/Fe–Co film laminated hybrids

Clustering of Fe–Pt clusters were chosen the following conditions: R = 6.7 × 10⁻⁶ m³/s for Ar gas, the input electric power, P_W = 150 W for the Pt target and P_W = 70 W for the Fe target. The average diameter is about 7–8 nm with the standard deviation of 20% and their average Fe composition, c_{Fe}, is about 46 at% Fe with the standard deviation of 15%.
Disordered fcc phases in the as-deposited state become ordered FePt phases by annealing at temperature, \( T_A > 770 \text{ K} \). The degree of order, which roughly estimated by the intensity ratio of (100) and (111) diffraction lines, became maximum at \( T_A = 970 \text{ K} \) and decreased at \( T_A = 1070 \text{ K} \). As shown in the SEM images (see Fig. 4), the as-prepared specimen consists of well separated particles, while the specimen annealed at \( T_A = 970 \text{ K} \) consists of aggregated particles. The magnetization curves of Fe–Pt clusters at room temperature indicate that the saturation magnetization per volume, \( M_S \), is about 0.6 T, being much smaller than 1.4 T for the bulk alloy at around \( c_{Fe} = 50 \text{ at}\% \text{Fe} \). This is due to the porous stacking of Fe–Pt clusters whose packing densities are 30–40% of bulk specimens. By annealing at temperature, \( T_A \), for a duration, 600 s, the magnetic coercivity, \( H_C \), rapidly increases to 0.68 MA/m at \( T_A = 870 \text{ K} \), becomes a maximum of 0.96 MA/m at \( T_A = 970 \text{ K} \) and decreases to 0.80 MA/m at \( T_A = 1070 \text{ K} \). By annealing at \( T_A = 970 \text{ K} \), \( H_C \) becomes 0.96 MA/m for 600–1200 s and 1.27 MA/m for 5400 s. The average composition of Fe–Co films were 65 at\%Fe. The magnetization measurement of an Fe–Co film with a thickness of 50 nm indicates that \( M_S \) is about 1.2 T. This value is about a half of 2.3 T for the bulk Fe–35 at\%Co alloy, being attributed to the oxidation of the Fe–Co film.

Then, Fe–Co films with \( x = 0–30 \text{ nm} \) and Fe–Pt cluster assemblies with \( t = 30 \text{ nm} \) were deposited layer by layer as shown in Fig. 2, where the preparation conditions of Fe–Pt clusters and Fe–Co films were same as mentioned above. Figure 5 depicts X-ray diffraction patterns of Fe–Pt cluster/Fe–Co film laminated hybrids annealed at \( T_A = 970 \text{ K} \) for 600 s. In the figure, Fe–Co oxide diffraction lines are detected for \( x \neq 0 \text{ nm} \) and bcc Fe–Co diffraction lines for \( x \geq 20 \text{ nm} \), while the FePt diffraction lines slightly shift to the high angle side, being ascribed to mixing of Fe–Pt clusters with Fe–Co films.42

Figure 6(a) depicts magnetization curves at room temperature for Fe–Pt cluster/Fe–Co film laminated hybrids. The saturation magnetization, \( M_S \), increases with \( x \) (see also Fig. 6(b)) because the \( M_S \) values of Fe–Co alloys are higher than those of Fe–Pt alloys. However, it is not so high as expected from the \( x \) values, being due to the oxidation of Fe–Co films. Two step demagnetization behaviors are detected for \( x \leq 15 \text{ nm} \), suggesting the incomplete ordering
of fcc Fe–Pt alloys. We also deposited only Fe-rich Fe–Pt clusters on a substrate with supplying $P_W = 140$ W into both Fe and Pt targets. As shown in Fig. 7(a), the two step demagnetization behavior cannot be removed even after very long annealing: the magnetization decreases just after the remanence state in the second quadrant. This is attributed to Fe-rich fcc phases because very wide compositional distribution is observed by the chemical analysis of individual Fe–Pt clusters (see Fig. 7(b)). Figure 6(b) depicts the magnetic characteristics estimated from Fig. 6(a). With increasing $x$, $M_S$ gradually increases, $H_C$ rapidly decreases, and $BH_{MAX}$ monotonically decreases.

3.2 Fe–Co cluster/Fe–Pt film laminated hybrids

We chose the following conditions for preparing Fe–Co clusters: $R = 5.8 \times 10^{-6}$ m$^3$/s for Ar gas and the input electric power, $P_W = 300$ W for the Fe–30 at%Co alloy targets: the average diameter thus obtained was about 18 nm (see Figs. 8(a) and 8(a')). They were disordered bcc Fe–Co alloys whose average Fe composition was about 65 at%/Fe with the standard deviation of 5 at%. The magnetization measurement indicated that $M_S = 0.34$ T, being much smaller than $M_S = 2.3$ T of bulk alloys owing to the low packing density (30–40%) and severe oxidation of these Fe–Co clusters. $H_C$ was about 0.03 T. In Fe–Pt films with nearly equiaxial compositions prepared by the helicon sputter deposition, the degree of order gradually increases at $T_A < 1170$ K. $H_C$ = 5.97, 8.76 and 11.54 MA/m, while $M_S$ = 0.9, 0.9 and 0.75 T, and $M_R$ = 0.8, 0.85 and 0.7 T after annealed at $T_A = 970$, 1070 and 1170 K for 600 s, respectively. The $M_R$ values are higher than 0.5 $M_S$, being ascribed to the preferred growth of these films.

Then, Fe–Co cluster assemblies with $t = 0.5–15$ nm and Fe–Pt films with $x = 20$ nm were stacked layer by layer (see Fig. 2), where the preparation conditions of Fe–Co clusters and Fe–Pt films were same as mentioned above, and the annealing treatment was carried out at $T_A = 970$ K for 600 s. As shown in Fig. 9, diffraction lines allotted to the ordered FePt phase are mainly detected and the peak angles shift to the higher side, suggesting diffusion of Fe and Co atoms from Fe–Co films to Fe–Pt clusters.42) Figure 10 depicts the magnetization curves at room temperatures of Fe–Co cluster/Fe–Pt film laminated hybrids with $t < 5$ nm after annealing at $T_A = 970$ K for 600 s. $H_C$ values are higher than 0.56 MA/m and $M_S$ values are higher than 0.9 T even for $t < 3$ nm. The faint two step demagnetization behaviors are detectable in all specimens: magnetizations slightly decrease just below the remanence state (in the second quadrant) even though the $M_R$ values are comparable to the $M_S$ values. Figure 11(a) depicts the magnetic characteristics estimated from Fig. 10(a). With increasing $t$, $M_S$ gradually increases, $H_C$ monotonically decreases, and $BH_{MAX}$ slightly increases. We also annealed the same specimens at $T_A = 1070$ K for 600 s. As shown in Fig. 11(b), however, the hard magnetic characters have not been much improved with increasing $t$.

Spring magnet characters have been expected in composite materials where small soft magnetic nano-particles and/or films are finely dispersed in hard magnet matrices to strengthen the magnetic exchange-coupling between soft and hard magnetic regions. Thus, we made Fe–Co clusters
smaller by mixing He gas with Ar gas in the cluster preparation process. Since thermal conductivity and viscosity of He gas are higher than those of Ar gas, He gas effectively cool vapor atoms to form clusters and to transport clusters from the growth room to the deposition chamber. We chose the following preparation conditions:

\[ R = 4.2 \times 10^{-6} \text{m}^3/\text{s} \] for Ar gas, and \[ R = 14 \times 10^{-6} \text{m}^3/\text{s} \] for He gas, and the input electric power, \[ P_W = 300 \text{W} \] for the Fe-30 at%Co alloy targets.

The average diameter of Fe-Co clusters thus obtained was about 7 nm with the standard deviation of 12% (see Figs. 8(b) and 8(b')). Here, we adjusted the chemical composition of Fe-Pt film with \( x = 20 \text{nm} \) to be \( c_{Fe} = 43 \text{ at}\% \) for taking into account of the atomic diffusion of Fe and Co from Fe-Co clusters to Fe-Pt films. The squareness of magnetization curves was improved in Fe-Co cluster/Fe-Pt film laminated hybrids after annealed at 970 and 1070 K for 600 s. However, \( M_S, H_C \) and \( B_{H\text{MAX}} \) values were not...
so much improved with $t$ in comparison with those of pure Fe–Pt films annealed at 970 and 1070 K for 600 s.

Finally, we deposited five layers of Fe–Co clusters (the average size: 7 nm) with $t$ and six layers of Fe–Pt films ($c_{Fe} = 43$ at%Fe) with $x = 10$ nm in order to increase the interface areas between clusters and films, where $t$ was limited within 2 nm to minimize Fe and Co diffusion from Fe–Co clusters to Fe–Pt films. Figure 12(a) depicts the magnetization curves of Fe–Co cluster/Fe–Pt film laminated hybrids annealed at 970 K for 600 s. As seen in this figure, all of these curves are very smooth, i.e., no two step demagnetization behavior and the magnetic squareness is improved. As shown in Fig. 12(b), $M_s$ and $H_C$ increase with $t$ and $BH_{MAX}$ values are much improved for $t < 1$ nm. It is worth to mention that these specimens reveal a spring back effect as shown in Fig. 13. The recoil permeability values, $\mu$, estimated from Fig. 13 are shown in Table 1.

4. Discussion

As mentioned above,fcc disordered alloy phases are obtained in as-deposited Fe–Pt clusters in contrast to formation of B2 type ordered phases in as-deposited Fe–Al and Co–Al clusters. When metal atoms sputtered out of the targets collide with each other at early stages and form alloy clusters. If the condensation energy of vapor atoms and cohesive energy of alloys are partly stored, they induce internal atomic diffusion and preferential atom pairing, leading to formation of ordered phases in these clusters. Referring to the theoretical and thermodynamical data, the cohesive energies, $-\Delta H$, of ordered FeAl and CoAl alloys are about 2 times larger than that of an ordered FePt alloy.

Fig. 11 Saturation magnetization, $M_s$, magnetic coercivity, $H_C$ and maximum energy product, $BH_{MAX}$ for Fe–Co cluster/Fe–Pt film laminated hybrids prepared at $P_w = 300$ W for the Fe–Co target and annealed at 970 K for 600 s. (a) Results estimated from the data in Fig. 8 and (b) those for the same specimens annealed at $T_A = 1070$ K for 600 s. The thickness of an Fe–Co cluster layer, $t < 5$ nm and the thickness of an Fe–Pt film layer, $x = 20$ nm.

Fig. 12 (a) Magnetization curves at room temperature and (b) saturation magnetization, $M_s$, magnetic coercivity, $H_C$ and maximum energy product, $BH_{MAX}$ for Fe–Co cluster/Fe–Pt film laminated hybrids prepared at $P_w = 300$ W for the Fe–Co target in an Ar and He gas mixed atmosphere and annealed at 970 K for 600 s. The thickness of an Fe–Pt film layer, $x = 10$ nm and the thickness of an Fe–Co cluster layer, $t < 2$ nm.

Fig. 13 Demagnetization curves and spring back characters in the second and third quadrants for Fe–Co cluster/Fe–Pt film laminated hybrids prepared at $P_w = 300$ W for the Fe–Co target in an Ar and He gas mixed atmosphere. The thickness of an Fe–Pt film layer, $x$ and the thickness of an Fe–Co cluster layer, $t$ are as follows: (a) $t = 0$ nm and $x = 50$ nm $\times$ 1, (b) $t = 2$ nm $\times$ 2 and $x = 20$ nm $\times$ 3, and (c) $t = 1$ nm $\times$ 5 and $x = 10$ nm $\times$ 6.
Comparing the equiatomic composition ranges of their equilibrium phase diagrams, B2 type ordered phases appear just below the high temperature liquid phases in the Fe–Al and Co–Al alloy systems (congruent types), while a disordered fcc phase between the liquid phase and the low temperature ordered phase in the Fe–Pt alloy system (an incongruent type). The disordered phase formation in Fe–Pt clusters infers that the stored energy is not enough for Fe and Pt atoms to internally diffuse and form the ordered phase in the present experimental conditions. The in-flight heating processes and direct formations of ordered FePt alloy nanoparticles are promising, however, Fe–Pt clusters thus obtained show isotropic hard magnetic properties. For production of nanocomposite spring magnets from magnetically hard Fe–Pt phases and magnetically soft Fe–Co phases, post heat treatments are requisite because the degree of order must be maximized in Fe–Pt alloys and good contacts between Fe–Co and Fe–Pt layers is important to enhance their exchange coupling. Therefore, it is worthwhile to discuss the appropriate annealing procedures of cluster/thin film laminated hybrids of Fe–Pt and Fe–Co alloys prepared by the present PGCCD system.

For Fe–Pt clusters, the annealing effects on both $H_C$ and the degree of order are consistent: the reduction of the ordering temperature in small clusters leads to the lowering of $H_C$ at $T_A = 1070 \text{K}$. The monotonic increase in $H_C$ with the annealing period at $T_A = 970 \text{K}$ clearly demonstrates that this is the optimum ordering temperature to keep the high degree of order against the thermal agitation. During several annealing procedures, however, particle–particle coalescence must be minimized, because it changes the magnetization process from a coherent rotation type to a domain-wall movement type. In particular, the diffusivity of Fe atoms in L1$_0$ ordered FePt alloy films is enhanced even at low temperatures by a highly correlated mechanism, which is based on the coporative movements of neighboring atoms and higher than that extrapolated from the high temperature data. Once L1$_0$ type ordered FePt clusters accept Fe atoms via diffusion they are forming Fe-rich fcc Fe–Pt phases within a short range distance. Therefore, mixing of Fe and Pt clusters with Fe–Co films must be suppressed during annealing. Moreover, there are two steps in the demagnetization curves of Fe–Co films leading to the small $M_S$ values and no strong exchange-coupling between soft Fe–Co layers and hard Fe–Pt layers. $M_S$ increases and $H_C$ rapidly decreases for $x > 15 \text{nm}$, being ascribed to an increase of Fe–Co alloy layers. Since Fe and Co diffuse at interfaces from Fe–Co films to Fe–Pt clusters, Fe-rich fcc phases are formed and $BH_{MAX}$ monotonically decreases with $x$.

Since Fe–Pt alloys are not so easily oxidized as Fe–Co alloys and $M_S$ of Fe–Co alloys is much larger than that of Fe–Pt alloys, a remarkable increase in $M_S$ has been expected for Fe–Co cluster/Fe–Pt film laminated hybrids. In Fig. 10, the two step demagnetization behaviors are slightly detected for Fe–Co cluster (about $18 \text{nm}$ in diameter)/Fe–Pt film ($x = 20 \text{nm}$) laminated hybrids in contrast to the one step demagnetization behavior for the Fe–Pt thin film. With increasing $t$ in Fig. 11, $M_S$ only slightly increases and $H_C$ gradually decreases, where the low packing density of Fe–Co clusters (about 30%) does neither contribute to the increase in $M_S$ nor enhance the magnetic exchange coupling between Fe–Co clusters and Fe–Pt films.

In Fe–Co cluster (about $7 \text{nm}$ in diameter)/Fe–Pt film ($x = 20 \text{nm}$) laminated hybrids, the magnetic squareness is much improved probably due to the increase in the contact area between Fe–Co clusters and Fe–Pt films, however, $M_S$ and $H_C$ values do not much increase probably due to the diffusion of Fe and Co atoms into Fe–Pt clusters, i.e., the increase in the Fe-rich Fe–Pt alloy phases. In Fe–Co cluster (7 nm in diameter)/Fe–Pt film ($x = 10 \text{nm}$) laminated hybrids, moreover, the exchange coupling between Fe–Co clusters and Fe–Pt films is more enhanced. In Fig. 12 the increase of $H_C$ with $t$ is ascribed both to the chemical composition shift to the equi-atomic composition in Fe–Pt films by diffusion of Fe and Co atoms from Fe–Co clusters, and to the thickness reduction of the soft Fe–Co phase which depresses the exchange coupling and magnetic anisotropy at the interfaces. Though $M_S$ does not much increase because of the low dense stacking of Fe–Co clusters, $BH_{MAX}$ becomes largest at $t = 1 \text{nm}$.

As mentioned in Introduction, excellent hard magnetic properties were reported for Fe particles/Fe–Pt thin film hybrids prepared by cluster deposition methods. However, the magnetization evaluated per weight was indicated and no information on the packing density of cluster assemblies. Since the magnetization per volume is necessary to estimate $BH_{MAX}$, we cannot compare the present results with the reported ones.

### Table 1. Recoil permeability $\mu_r$ for Fe–Co cluster/Fe–Pt film hybrids. $x$ is the thickness of an Fe–Pt film layer, and $t$ the thickness of an Fe–Co cluster layer.

<table>
<thead>
<tr>
<th>Fe–Pt film layer</th>
<th>Fe–Co cluster layer</th>
<th>Fe–Co in vol%</th>
<th>$\mu_r$ ($\text{m}^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>($x = 50 \text{nm}$) × 1</td>
<td>($t = 0 \text{nm}$)</td>
<td>0</td>
<td>$2.3 \times 10^{-13}$</td>
</tr>
<tr>
<td>($x = 20 \text{nm}$) × 3</td>
<td>($t = 2 \text{nm}$) × 2</td>
<td>7</td>
<td>$4.4 \times 10^{-13}$</td>
</tr>
<tr>
<td>($x = 10 \text{nm}$) × 6</td>
<td>($t = 1 \text{nm}$) × 5</td>
<td>8</td>
<td>$4.9 \times 10^{-13}$</td>
</tr>
</tbody>
</table>

5. Summary

By combining the PGCCD system with the helicon-plasma-type-sputter-deposition system, we have prepared Fe–Pt cluster/Fe–Co film and Fe–Co cluster/Fe–Pt film laminated hybrids and measured the magnetic and structural properties to understand the possibility and limitation of these cluster/film hybrids for spring magnets. In Fe–Pt cluster/Fe–Co film laminated hybrids (see Fig. 5(b)), $M_S$ increases, while $H_C$ and $BH_{MAX}$ monotonically decrease with the Fe–Co film thickness: the hard magnetic characteristics have not been improved in comparison with those of pure Fe–Pt films. In Fe–Co cluster/Fe–Pt film hybrids (see Fig. 11), on the other hand, $M_S$, $H_C$ and $BH_{MAX}$ increase with the Fe–Pt film thickness in comparison with those of Fe–Pt films. Even then the low packing density of Fe–Co clusters impedes the remarkable improvement of their hard magnetic properties.
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