Alignment of c-Axis Orientation Perpendicular to the Substrate Surface in FePt Alloy Thin Film with L1₀ Structure

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An FePt alloy thin film with L₁₀ structure formed on a (001)-oriented polycrystalline underlayer or on a (001)-oriented single-crystal substrate may involve (100) variant with the c-axis lying in the film plane in addition to (001) variant with the c-axis normal to the substrate surface. In the present study, FePt(2–40 nm) films with and without MgO(2 nm) cap layers are prepared on MgO(001) single-crystal substrates by employing a two-step formation method consisting of low-temperature deposition followed by high-temperature annealing. Disordered FePt(001) single-crystal films with and without MgO(001) single-crystal cap layers epitaxially grow on the substrates at 200 °C. For the films thinner than 10 nm, the in-plane and out-of-plane lattices are respectively expanded and shrunk due to accommodation of the lattice misfit of film with respect to the substrate. With increasing the thickness up to 40 nm, the strain is relaxed. The phase transformation from A₁ to L₁₀ occurs when the films are annealed at 600 °C. The cap layer seems to prevent the strain release and to promote the phase transformation along the perpendicular direction. The c-axis orientation is controlled to be perpendicular by using the two-step method for the FePt films of thicknesses less than 10 nm with MgO cap layers. On the contrary, the film of 40 nm thickness with cap layer and the films of 2–40 nm thicknesses without cap layers involve small volumes of L₁₀(100) variant. The order degrees of films with cap layers are higher than those of films without cap layers. The films with cap layers have flat surfaces with the arithmetical mean roughness values lower than 0.3 nm for all the investigated thicknesses, whereas an island-like surface is formed for the 2-nm-thick film without cap layer. The film consisting of only L₁₀(001) variant shows a strong perpendicular magnetic anisotropy and a very low in-plane coercivity. Introduction of MgO cap layer is useful in aligning the c-axis perpendicular to the substrate surface, enhancing the L₁₀ ordering, forming a flat surface, and achieving a strong perpendicular magnetic anisotropy.

Key words: FePt alloy thin film, L₁₀ ordered structure, c-axis orientation, perpendicular magnetic anisotropy

1. Introduction

L₁₀ ordered FePt₁, CoPt₂, and FePd₃ alloys, respectively, show uniaxial magnetocrystalline anisotropy energies of \(6.6 \times 10^7\), \(4.9 \times 10^7\), and \(1.8 \times 10^7\) erg/cm³ along the c-axis and the thin films with the easy magnetization axis normal to the substrate surface have been investigated for ultra-high density perpendicular magnetic recording media⁴⁻⁷⁸. When an L₁₀ film is formed on a polycrystalline underlayer consisting of (001)-oriented grains or on a (001) single-crystal substrate, the film may involve (100) variant with the c-axis lying in the film plane in addition to (001) variant with the c-axis normal to the substrate surface. For practical application, alignment of c-axis orientation perpendicular to the substrate surface is essential as well as achievements of flat film surface, high order degree, and strong perpendicular magnetic anisotropy. In the present paper, the notations of plane and direction based on face-centered cubic (fcc) structure [Fig. 1(a)] are applied to the L₁₀ structure for simple comparison with the disordered structure (A₁), though the accurate structure is body-centered tetragonal (bct) [Fig. 1(b)].

In order to investigate the mechanism of thin film growth, it is useful to employ an epitaxial film, since the crystallographic orientation can be controlled by that of single-crystal substrate. FePt,⁶⁻⁷⁷, CoPt,⁶¹⁻⁷³, and FePd⁶⁸⁻⁶⁹,⁶⁶⁻⁶⁷,⁷⁴⁻⁷⁹ epitaxial films have been prepared on (001) substrates of MgO, SrTiO₃, etc. The c-axis orientation has been investigated by high-resolution transmission electron microscopy⁶⁸⁻⁶⁹, selected area electron diffraction⁶⁸, in-plane⁶⁴⁻⁶⁷, and pole-figure⁶⁸ X-ray diffractions (XRDs), and reflection high-energy electron diffraction (RHEED)⁶⁴⁻⁶⁷.

In our previous studies⁶⁴⁻⁶⁹, FePt epitaxial films of 40 nm thickness were prepared on MgO(001) substrates by using two different methods. One was a one-step method consisting of high-temperature deposition at...
600 °C, where an L1₀ crystal nucleated on the substrate. The other was a two-step method consisting of low-temperature deposition at 200 °C followed by high-temperature annealing at 600 °C, where phase transformation from A1 to L1₀ occurred. The film prepared by one-step method possessed an island-like surface involving facets, whereas a flat film surface was realized by using the two-step method. The c-axis orientation was carefully studied by combination of out-of-plane and in-plane XRDs. Although the preparation methods were different, both films involved L1₀(100) variant in addition to L1₁(001) variant.

The chlorine ratio of L1₁₁FePt crystal \(a = 0.3842 \text{ nm, } c = 0.3702 \text{ nm}\) is less than one. Thus, application of out-of-plane lattice compression or in-plane lattice expansion for an A1-FePt(001) film before annealing is considered to be useful for aligning the c-axis normal to the substrate surface. The phase transformation along the perpendicular direction seems to preferentially occur in the strained film. The lattice constant of MgO crystal \(a_{\text{MgO}} = 0.4211 \text{ nm} \) is larger than that of A₁-FePt crystal \(a_{\text{FePt}} = 0.386 \text{ nm} \) \(c_{\text{FePt}} \times (10/17.5) + (0.380 \text{ nm} \times (7.5/17.5) = 0.383 \text{ nm})\). When an A₁-FePt(001) film with MgO(001) cap layer deposited on MgO(001) substrate is annealed, the cap layer is considered to prevent the relaxation of strain caused by accommodation of the lattice misfit of film with respect to the MgO substrate and to effectively expand the in-plane lattice of FePt film.

In the present study, FePt films with and without thin MgO cap layers are prepared by using the two-step method. The film thickness, which influences the lattice strain, is varied in a range between 2 and 40 nm. The influences of MgO cap layer and film thickness on the lattice strain, the c-axis orientation, the order degree, and the magnetic property are investigated.

### 2. Experimental Procedure

A radio-frequency (RF) magnetron sputtering system equipped with an RHEED facility was used for film preparation. The base pressures were lower than 4 \(\times 10^{-7} \text{ Pa}\). Polished MgO(001) substrates were heated at 600 °C for 1 hour in the chamber before deposition to obtain clean surfaces. Figure 2(a) shows the RHEED pattern of an MgO substrate after heating observed by making the incident electron beam parallel to MgO[100]. A clear diffraction pattern involving Kikuchi lines is recognized. The observed pattern corresponds to a diffraction pattern simulated for clean \(B_1\)-MgO(001) single-crystal surface shown in Fig. 2(b). Figure 2(c) shows the atomic force microscope (AFM) image observed for the substrate. A flat surface is confirmed. The arithmetical mean roughness value, \(R_a\), is 0.1 nm.

The distance between target and substrate was set at 150 mm. Fe₃₅Pt₆₅ (at. %) and MgO targets of 3 inch diameter were employed and the respective RF powers were fixed at 43 and 200 W. The Ar gas pressure was kept constant at 0.67 Pa. Under the conditions, the deposition rates were 0.020 and 0.015 nm/s for the FePt and the MgO materials, respectively.

An FePt film and an MgO cap layer were sequentially deposited on the substrate at 200 °C. The thickness of FePt film was varied in a range between 2 and 40 nm, whereas that of cap layer was fixed at 2 nm. The sample was then annealed at 600 °C for 1 hour for enhancing the \(L_1_0\) ordering. FePt films without cap layers were also prepared for comparison with those with cap layers. The FePt film compositions were confirmed by energy dispersive X-ray spectroscopy and the errors were less than 4 at. % from the target composition.

The surface structure was studied by RHEED. The structural properties were investigated by 2θ/ω-scan out-of-plane and 2θ/φ-scan in-plane XRDs with Cu-Kα radiation \(\lambda_{\text{Kα}} = 0.15418 \text{ nm}\). The surface morphology was observed by AFM. The magnetization curves were measured by vibrating sample magnetometry.

### 3. Results and Discussion

#### 3.1 Surface morphology

Figures 3(a-1)–(c-1) and (a-2)–(c-2) show the AFM images observed for FePt films of different thicknesses without cap layers before and after annealing at 600 °C, respectively. The films of 10 and 40 nm thicknesses before annealing have flat surfaces and the flatness is kept after annealing. On the contrary, an island-like surface is recognized for the 2-nm-thick film after annealing, though the 2-nm-thick film before annealing possesses a flat surface. Condensation of deposited atoms on MgO substrate occurs by annealing a thinner film.

Figure 4 shows the AFM images observed for FePt films with 2-nm-thick MgO cap layers. Flat surfaces are realized for the films after annealing by introducing the cap layers, even though the film thickness is decreased to 2 nm.
3.2 Formation of disordered FePt films

Figure 5 shows the RHEED patterns observed for FePt films without cap layers deposited on MgO(001) substrates at 200 °C. Diffraction patterns consisting of streaks are recognized. The films epitaxially grow on the substrates. Very broad streaks are recognized for the films of 2 and 10 nm thicknesses [Figs. 5(a), (b)]. The films are strained due to accommodation of the lattice mismatch with respect to the MgO substrate. The films of 2 and 10 nm thicknesses [Figs. 5(a), (b)] are observed, since the patterns of Figs. 6(a) and (b) are similar. The crystal structure seems to be not enough to detect reflections from superlattice and fundamental reflections, respectively. The result indicates that the strain decreases with increasing the thickness.

Figures 6(a), (b), and (c), respectively, show the schematic diagrams of diffraction patterns calculated for A1(001), L10(001), and L10(100) crystals in the orientation relationships of

\[
\begin{align*}
&A1\text{-FePt(001)[100]} \parallel MgO(001)[100], \\
&L10\text{-FePt(001)[100]} \parallel MgO(001)[100], \\
&L10\text{-FePt(100)[001], (100)[010]} \parallel MgO(001)[100].
\end{align*}
\]

The observed RHEED patterns correspond to the simulated pattern from A1(001) or L10(001) crystal. The identification is not easy when a streak pattern is observed, since the patterns of Figs. 6(a) and (b) are similar. The crystal structure is therefore determined by XRD.

Figures 7(a-1)–(c-1) and (a-2)–(c-2), respectively, show the out-of-plane and in-plane XRD patterns of FePt films without cap layers epitaxially grown on MgO substrates. Only the fundamental reflections of FePt(002) and FePt(200) are respectively observed in the out-of-plane and in-plane patterns of 10-nm-thick film [Fig. 7(b)]. The crystal structure is thus A1. In the patterns of 2-nm-thick film [Fig. 7(a)], no reflections from film are recognized. The XRD sensitivity is considered to be not enough to detect reflections from the very thin strained film.
where \( r \) is the ratio of Fe atom positions occupied with absorption factor (\( A \)). In the L10 structure shown in Fig. 1(a), the positions of Fe atom are \((0, 0, 0)\) and \((1/2, 1/2, 0)\), while those of Pt atom are \((1/2, 0, 1/2)\) and \((0, 1/2, 1/2)\). The structure factor \( F(hkl) \) of \((hkl)\) is thus expressed as

\[
F(hkl) = \{ rf_{\text{Fe}} + (1-r)f_{\text{Pt}}\} \{ 1 + e^{2\pi i (hk+lj+mk)} \} + \{ rf_{\text{Pt}} + (1-r)f_{\text{Fe}}\} \{ e^{2\pi i (jxj+kyl+lzk)} \},
\]

where \( r \) is the ratio of Fe atom positions occupied with Pt atoms to the total Fe atom positions \((0.5 \leq r \leq 1)\) and \( f \) is the atomic scattering factor of Fe or Pt. \( \text{SL10}(100) \) is given as

\[
\text{SL10}(100) = 2r - 1.
\]

Therefore, \( F(001) = 2\text{SL10}(100)(f_{\text{Pt}} - f_{\text{Fe}}) \), \( F(100) = F(002) = 0 \), and \( F(002) = F(100) = 2(f_{\text{Pt}} + f_{\text{Fe}}) \).

The order degree of \( \text{L10}(100) \) crystal, \( \text{SL10}(100) \), can be calculated by using in-plane XRD data. Reflection intensity \( (I) \) is proportional to \( F \) and the complex conjugate \( (F^*) \), Lorentz-polarization factor \( (L) \), and absorption factor \( (A) \). In the \( \text{L10} \) structure shown in Fig. 1(a), the positions of Fe atom are \((0, 0, 0)\) and \((1/2, 1/2, 0)\), while those of Pt atom are \((1/2, 0, 1/2)\) and \((0, 1/2, 1/2)\). The intensity ratio of FePt(001) superlattice to \((002)\) fundamental in-plane reflection is expressed as

\[
\frac{I(001)}{I(200)+(020)+(002)} = (FF^* L)^{1/2},
\]

where \( k \) and \( W-L \) are due to Cu-K\(\alpha \) and W-La radiations included in the X-ray source, respectively.

The \( \text{SL10}(100) \) value of 40-nm-thick FePt film without cap layer is calculated to be 0.11.

\[
\text{SL10}(100) = \left( \frac{I(001)}{I(200)+(020)+(002)} \right)^{1/2}.
\]

The scattering vector of in-plane XRD is parallel to MgO[100] substrate. The small reflections noted as \( k \) and \( w \) are due to Cu-K\(\alpha \) and W-La radiations included in the X-ray source, respectively.
Figure 9 shows the out-of-plane and in-plane XRD patterns. The films with cap layers show diffraction patterns similar to the case of films without cap layers for the same thicknesses. The 2- and 10-nm-thick films with cap layers consist of A1(001) crystal, whereas the 40-nm-thick film involves a small volume of L10(100) crystal in addition to A1(001) crystal. The $S_{L10[100]}$ value is estimated to be 0.07.

### 3.3 Strain of disordered FePt films

Figures 10 and 11, respectively, show the lattice parameters of $a$, $c$, and $c/a$ of FePt films without and with cap layers before annealing, which are calculated from the XRD data by using the relations of

$$a = 2\lambda_{X_{cap}} / [2\sin(\theta_{FePt[002]})] ,$$  

$$c = 2\lambda_{X_{cap}} / [2\sin(\theta_{FePt[200]})].$$

The $a$ value is larger than the $c$ value for the 2- and the 10-nm-thick films both without and with cap layers. The in-plane and out-of-plane lattices are respectively expanded and shrunk due to accommodation of the lattice misfit of film with respect to the substrate. With increasing the thickness from 10 to 40 nm, the $c/a$ ratios of films without and with cap layers increase from 0.9768 to 1.0079 and from 0.9737 to 1.0067, respectively. The strain decreases with increasing the thickness, as expected by the RHEED observation.

From the RHEED and the XRD results, strain distributions in FePt films without and with cap layers along the thickness are considered and possible schematic illustrations are depicted in Fig. 12. Larger strain is introduced in an FePt film around the FePt/MgO interface, similar to the cases of Fe$_{50}$Co$_{50}$/MgO(001), Ni$_{50}$Fe$_{20}$/MgO(001) systems, and Co/MgO(001) systems. A small volume of L10(100) variant is thus considered to be formed in the thickness region from 10 to 40 nm, where the strain caused by accommodation of
the lattice misfit between film and substrate seems to be released.

### 3.4 c-axis orientation of L1₀ ordered FePt films

Figure 13(a) shows the RHEED pattern observed for a 2-nm-thick A1-FePt(001) film without cap layer after annealing at 600 °C. The diffraction pattern apparently involves reflections from both L1₀(001) and L1₀(100) variants, as shown in the schematic diagrams of Figs. 6(b) and (c-3). Phase transformation from A1 to L1₀ is taking place in the 2-nm-thick FePt film. The crystallographic orientation relationships are determined as

\[
\begin{align*}
\text{A1-FePt(001)[001]} & \parallel \text{L1₀-FePt(001)[001]}, \\
\text{A1-FePt(001)[100]} & \parallel \text{L1₀-FePt(100)[001]}, \\
\text{A1-FePt(001)[010]} & \parallel \text{L1₀-FePt(100)[001]}. 
\end{align*}
\]

Figures 13(b) and (c) show the RHEED patterns observed for FePt films of 10 and 40 nm thicknesses without cap layers, respectively. The diffraction patterns also include the reflections from L1₀(100) variant, as shown by the arrows in the intensity profiles. Figure 14 shows the XRD patterns of films without cap layers after annealing. Strong out-of-plane and weak in-plane FePt(001) superlattice reflections are observed for all the films. The RHEED and the XRD results show that the films involve L1₀(100) variant in addition to L1₀(001) variant. Although the films of 2 and 10 nm thicknesses before annealing are strained, the phase transformation occurs not only along the perpendicular direction of A1[001] but also along the in-plane directions of A1[100] and A1[010].

Figure 15 shows the a, c, and cla values of films without cap layers after annealing. The cla ratio of 10-nm-thick film after annealing (0.9781) is slightly larger than that of 10-nm-thick film before annealing.
observed for all the films. The MgO(001) cap layer FePt films with MgO cap layers after annealing. are relaxed by the high-temperature annealing. in the films of 2 and 10 nm thicknesses, whose strains more enhanced for the 2-nm-thick film as shown by the thinner film. The reason is because atomic diffusion is indicating that the strain is more released for the (0.9906) than that of 10-nm-thick film (0.9781), 2-nm-thick film after annealing shows a larger c/a ratio 0.375 than that of 10-nm-thick film 0.380, 2-nm-thick film after annealing shows a larger c/a ratio 0.9906 than that of 10-nm-thick film 0.9781, indicating that the strain is more released for the thinner film. The reason is because atomic diffusion is more enhanced for the 2-nm-thick film as shown by the AFM images of films before and after annealing [Fig. 3(a)]. \( L_{10}(100) \) variant is thus considered to be included in the films of 2 and 10 nm thicknesses, whose strains are relaxed by the high-temperature annealing.

Figure 16 shows the RHEED patterns observed for FePt films with MgO cap layers after annealing. Reflections from MgO(001) single-crystal surfaces are observed for all the films. The \( \text{MgO}(001)_{\text{cap layer}} /\text{FePt}(001)/\text{MgO}(001)_{\text{substrate}} \) hetero-epitaxial structure is kept after annealing. Figures 17(a) and (b) show the XRD patterns of films of 2 and 10 nm thicknesses after annealing. FePt(001) superlattice reflections are observed only in the out-of-plane patterns, indicating that the films consist of only \( L_{10}(001) \) variant. The c-axis orientation is controlled to be perpendicular by using the cap layer. Figure 17(c) shows the XRD patterns of 40 nm-thick film. Out-of-plane and in-plane FePt(001) superlattice reflections are observed, similar to the case of 40 nm-thick film without cap layer. The film consists of a mixture of \( L_{10}(001) \) and \( L_{10}(100) \) variants, since the \( L_{10}(100) \) variant is included in the film before annealing.

Figure 18 shows the \( a, c, \) and \( c/a \) values of films with cap layers after annealing. The \( c/a \) ratios of 2- and 10 nm-thick films are 0.9571 and 0.9599, respectively. The cap layer prevents the relaxation of strain caused by accommodation of the lattice misfit of film with respect to the MgO substrate. Therefore, the phase
transformation preferentially occurs along the perpendicular direction in the MgO/FePt/MgO structure.

### 3.5 Order degree

The order degree of \( L_{10}(001) \) variant, \( S_{L10(001)} \), can be calculated by using out-of-plane XRD data. The intensity ratio of FePt(001) superlattice to FePt(002)+(200) fundamental out-of-plane reflection is expressed as

\[
S_{L10(001)} = \left[ \frac{I(001)}{I(002) + I(200)} \right]^{1/2},
\]

where \( I(001) \) and \( I(100) \) are the volume ratios of \( L_{10}(001) \) and \( L_{10}(100) \) variants. Here,

\[
V(001) + V(100) = 1.
\]

Since diffraction angles of FePt(001) and FePt(200) reflections are generally close each other, \( F(002) \approx F(001) \). Therefore,

\[
S_{L10(001)} = \left[ \frac{F(001)/(FF*LA)(001)}{F(002)/(FF*LA)(002)} \right]^{1/2}.\]

In the case of \( V(001) \gg V(100) \), the equation (16) is expressed as

\[
S_{L10(001)} = \left[ \frac{1}{1 + V(100)/V(001)} \right] \times \frac{(FF*LA)(001)}{(FF*LA)(002)}.\]

Therefore, \( S_{L10(001)} \) is given as

\[
S_{L10(001)} = \left[ \frac{F(002)}{F(001)} \right] \times \left[ \frac{V(001)/(FF*LA)(001)}{V(100)/(FF*LA)(002)} \right].
\]

Figure 19(a) shows the order degrees, \( S_{L10(001)} \), of FePt films (a) without and (b) with cap layers after annealing.

### 3.6 Magnetic properties

Figure 20(a) shows the magnetization curves of 40-nm-thick FePt films without and with cap layer after annealing, whose \( S_{L10(001)} \) values are respectively 0.21 and 0.34. The out-of-plane and in-plane magnetizations do not saturate at an applied field of 10 kOe. The reason is due to that the films consist of a mixture of \( L_{10}(001) \) and \( L_{10}(100) \) variants whose easy magnetization axes are respectively perpendicular and parallel to the substrate surface. The film with cap layer shows a
stronger perpendicular magnetic anisotropy than that without cap layer, since a higher $S_{L10(001)}$ value is obtained.

Figure 20(b) shows the magnetic properties of 10-nm-thick FePt films without and with cap layer after annealing, whose $S_{L10(001)}$ values are respectively 0.58 and 0.82. The films show strong perpendicular anisotropies. The out-of-plane coercivities of films without and with cap layers are 1.7 and 2.1 kOe, respectively. The magnetization seems to reverse by domain wall motion\(^{49}\), because magnetic continuous structures with flat surfaces are formed. The film with cap layer shows a very low in-plane coercivity due to that the $c$-axis orientation is controlled to be perpendicular to the substrate surface and the $L1_0$ ordering is promoted by employing the cap layer.

4. Conclusion

FePt films without and with thin MgO cap layers are prepared by varying the thickness in a range between 2 and 40 nm. The influences of lattice strain on the $c$-axis orientation, the order degree, and the magnetic property are investigated. The films of 2 and 10 nm thicknesses with cap layers consist of only $L1_0(001)$ variant with the $c$-axis normal to the substrate surface, whereas those without cap layers involve $L1_0(100)$ variant with the $c$-axis lying in the film plane. The $c$-axis orientation is controlled to be perpendicular by introducing the cap layer for the films thinner than 10 nm. The cap layer prevents the relaxation of strain caused by accommodation of the lattice misfit between FePt film and MgO substrate and to effectively expand the in-plane lattice of FePt film. The phase transformation along the perpendicular direction preferentially occurs in the strained film with cap layer. The $S_{L10(001)}$ values of films with cap layers are larger than those of films without cap layers for the same film thicknesses. Furthermore, the cap layer prevents condensation of deposited FePt atoms on MgO substrate caused by high-temperature annealing after deposition. The film with cap layer shows a strong perpendicular magnetic anisotropy and a very low in-plane coercivity. Introduction of MgO cap layer is useful for aligning the $c$-axis perpendicular to the substrate surface and achieving a higher order degree, a flat film surface, and a strong perpendicular magnetic anisotropy.

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References

1) O. A. Ivanov, L. V. Solina, V. A. Demishina, and L. M. Magat: *Fiz. Metal. Metalloved.*, 35, 81 (1973).
2) Y. S. Shur, L. M. Magat, G. V. Ivanova, A. I. Mitsek, A. S. Yermolenko, and O. A. Ivanov: *Fiz. Met. Metalloved.*, 26, 241 (1968).
3) A. Ye. Yermakov and V. V. Maykov: *Phys. Met. Metalloved.*, 69, 198 (1990).
4) T. Suzuki, K. Harada, N. Honda, and K. Ouchi: *J. Magn. Magn. Mater.*, 193, 85 (1999).
5) N. Li and B. M. Lairson: *IEEE Trans. Magn.*, 35, 1077 (1999).
6) R. A. Ristau, K. Barmak, L. Henderson-Lewis, K. R. Coffey, and J. K. Howard: *J. Appl. Phys.*, 86, 4527 (1999).
7) C. P. Luo, S. H. Liou, and D. J. Sellmyer: *J. Appl. Phys.*, 87, 6941 (2000).
8) M. Watanabe, T. Masumoto, D. H. Ping, and K. Hono: *Appl. Phys. Lett.*, 76, 3971 (2000).
9) P. C. Kuo, Y. D. Yao, C. M. Kuo, and H. C. Wu: *J. Appl. Phys.*, 87, 6146 (2000).
10) J. A. Christodoulides, Y. Huang, Y. Zhang, G. C. Hadjipanayis, I. Panagiotopoulos, and D. Niarchos: *J. Appl. Phys.*, 87, 6938 (2000).
11) C. Chen, O. Kitakami, S. Okamoto, and Y. Shimada: *Appl. Phys. Lett.*, 76, 3218 (2000).
12) S. Jeong, Y. N. Hsu, M. E. McHenry, and D. E. Laughlin: *Appl. Phys. Lett.*, 87, 6950 (2000).
13) Y. Endo, N. Kikuchi, O. Kitakami, and Y. Shimada: *J. Appl. Phys.*, 89, 7065 (2001).
14) S. Jeong, M. E. McHenry, and D. E. Laughlin: *IEEE Trans. Magn.*, 37, 1309 (2001).
15) O. Kitakami, Y. Shimida, K. Oikawa, H. Daimon, and K. Fukamiuchi: *Appl. Phys. Lett.*, 78, 1104 (2001).
16) Y. Xu, J. S. Chen, D. Dai, and J. P. Wang: *IEEE Trans. Magn.*, 38, 2042 (2002).
17) T. Maeda, A. Kikitsu, T. Kai, T. Nagase, H. Aikawa, and J. Akiyama: *IEEE Trans. Magn.*, 38, 2796 (2002).
18) Y. K. Takahashi, M. Ohnuma, and K. Hono: *J. Magn. Magn. Mater.*, 246, 259 (2002).
19) S. Jeong, T. Ohkubo, A. G. Roy, D. E. Laughlin, and M. E. McHenry: *J. Appl. Phys.*, 91, 6863 (2002).
20) H. Ito, T. Kununoki, H. Saito, and S. Ishiw: *J. Magn. Soc. Jpn.*, 27, 1083 (2003).
21) M. F. Toney, W. Y. Lee, J. A. Hedstrom, and A. Kellok: *J. Appl. Phys.*, 93, 9902 (2003).
22) H. Kodama, S. Momose, T. Uzumaki, and A. Tanaka: *J. Magn. Soc. Jpn.*, 28, 372 (2004).
23) A. C. Sun, P. C. Kuo, J. H. Hsu, H. L. Huang, and J. M. Sun: *J. Appl. Phys.*, 98, 076109 (2005).
24) S. Nakagawa and T. Kamiki: *J. Magn. Magn. Mater.*, 287, 204 (2005).
25) N. Katayama, T. Maekawa, X. Liu, A. Morisako, and M. Matsumoto: *J. Magn. Soc. Jpn.*, 29, 235 (2005).
26) T. Hasegawa, G. Q. Li, W. Pei, H. Saito, S. Ishio, K. Taguchi, K. Yamakawa, N. Honda, K. Ouchi, T. Aoyama, and I. Sato: *J. Appl. Phys.*, 99, 053505 (2006).
27) N. Murayama, S. Soeya, Y. Takahashi, and M. Futamoto: *J. Magn. Magn. Mater.*, 320, 3057 (2008).
28) A. Itoh and A. Tsukamoto: *J. Magn. Soc. Jpn.*, 33, 507 (2009).
29) H. Ho, J. Wu, A. Kulovits, D. E. Laughlin, and J. -G. Zhu: *Dig. IEEE Internmag*, Dresden, Germany, 2014, pp. 581–582 (BB-06).
30) B. M. Lairson, M. R. Visokay, R. Sinclair, and B. M. Clemens: *Appl. Phys. Lett.*, 62, 639 (1993).
31) A. Cebollada, D. Weller, J. Sticht, G. R. Harp, R. F. C. Farrow, R. F. Marks, R. Savoy, and J. C. Scott: *Phys. Rev. B*, 50, 3419 (1994).
32) R. F. C. Farrow, D. Weller, R. F. Marks, M. F. Toney, S. Hom, G. R. Harp, and A. Cebollada: *Appl. Phys. Lett.*, 69, 1166 (1996).
33) M. Watanabe and M. Homma: *Jpn. J. Appl. Phys.*, 35, L1264 (1996).
34) R. F. C. Farrow, D. Weller, R. F. Marks, M. F. Toney, A. Cebollada, and G. R. Harp: *J. Appl. Phys.*, 79, 5967 (1996).
35) M. H. Hong, K. Hono, and M. Watanabe: *J. Appl. Phys.*, 84, 4403 (1998).
36) J. U. Thiele, L. Folks, M. F. Toney, and D. K. Weller: *J. Appl. Phys.*, 84, 5686 (1998).
37) T. Goto, Y. Ide, K. Watanabe, and H. Yoshida: *J. Magn. Soc. Jpn.*, 87, 175 (2015).
