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Formation of Co nanodisc with enhanced perpendicular magnetic anisotropy driven by Ga$^+$ ion irradiation on Pt/Co/Pt films

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The origin of magnetic phase transition from in-plane to perpendicular magnetic anisotropy (PMA) of Pt/Co/Pt thin film by Ga$^+$ ion irradiation at fluences of 1–5 × 10$^{15}$ ions/cm$^2$ is investigated by means of x-ray magnetic circular dichroism (XMCD) and extended x-ray absorption fine structure (EXAFS) analyses. We find that Pt and Co atoms are mixed with each other and that Co is oxidized near the surface due to removal of the Pt overlayer. Furthermore, polarization-dependent EXAFS analysis shows that Co is firstly dispersed as separated single-atom-thick sheets in a Pt matrix at 1 × 10$^{15}$ ions/cm$^2$, then the Co sheets are divided into a few Å clusters at 5 × 10$^{15}$ ions/cm$^2$, which are regarded as nanodiscs parallel to the film plane. This process is accompanied by the appearance of an out-of-plane magnetization component and a remanence peak is observed. Because we do not observe an enhancement in anisotropy of Co orbital moment which leads to change in magnetic anisotropy through the transition at about 5 × 10$^{15}$ ions/cm$^2$, it might be possible that such nanodisc formation induces increase of magnetic anisotropy via a shape effect. By comparing with the phase transition observed at lower fluence [Phys. Rev. B 86, 024418 (2012)], we find that the mechanism of two transitions is different, i.e., the transition at lower fluence is caused by anisotropy of orbital moment due to structural strain, while the present transition is possibly by shape effect due to nanodisc formation.

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I. INTRODUCTION

Controlled modification of the magnetic properties of nanoscale magnetic materials by focused beams, such as ions and laser, is an important technique for fabricating high-density magnetic recording media. Among such techniques, ion irradiation has attracted much interest because it shows fine beam convergence, controllability of beam intensity, and homogeneity of the beam spot. In fact, ion irradiation-induced changes of magnetic anisotropy of magnetic thin films have been reported by using various ion species [1,2]. To understand the mechanism of the ion irradiation-induced magnetic change of the films, impact of the collision between ions and concerned atoms has to be described from a physical viewpoint, e.g., energy and momentum transfer, penetrability of ions, and structural change. It is important to choose proper ion species to realize desired magnetic modifications. Ga$^+$ ion irradiation for a Pt/Co/Pt thin film is a promising candidate to realize perpendicularly magnetized thin films because irradiation-induced effect on the magnetic anisotropy is larger than that by other lighter ions [3]. Ga$^+$ ion irradiation is especially important because of application in focused ion beam (FIB) technique enabling local modifications of magnetic and structural properties of continuous films at the nanometer scale [1,2]. Pt/Co/Pt FIB patterning can be used for different applications, e.g., Refs. [4,5]. Recently, it has been found that Ga$^+$ ion irradiated Pt/Co/Pt thin film shows two magnetic phase transitions from in-plane to perpendicular magnetization with increasing ion fluence [6,7]. The first transition at the lower fluence is found to be caused by anisotropy of Co orbital moment due to structural strain [8], while it is not clear what a factor induces the second transition at the higher fluence. Jakubisova-Liskova et al. experimentally revealed heavy intermixing between the top Pt and Co layers, and the rest of cobalt can be interpreted as Co-Pt alloy at the second perpendicular magnetic anisotropy (PMA) phase [9]. In addition, Avchaciov et al. predicted preservation of the fcc long-range order in the irradiated structures from molecular dynamics simulations [10]. On the other hand, they found a negligible fraction of a specific chemical short-range order which could be associated with the possible magnetic anisotropy of the second transition. Therefore, we focus on the short-range order of Co, which could give rise to the PMA, and investigate its magnetic state and local crystal structure by x-ray magnetic circular dichroism (XMCD) and extended x-ray absorption fine structure (EXAFS) analyses, respectively. Here, we show from the polarization-dependent Co K-edge EXAFS analysis that Co is dispersed as a separated sheet in a Pt matrix before the second transition, then the Co sheet is divided into a few Å clusters, which are regarded as nanodiscs parallel to the film plane and associated with the appearance of the second PMA phase. On the other hand, from the Co L-edge XMCD analysis, we do not observe an enhancement in anisotropy of Co orbital moment which might lead to change in magnetic anisotropy through the second transition. From these findings, we assume that such nanodisc formation induces out-of-plane magnetization component via the shape effect, and that the mechanism of two transitions is different, i.e., the transition at lower fluence is caused by

PHYSICAL REVIEW B 94, 174422 (2016)
anisotropy of orbital moment due to structural strain, while the second transition is possibly by nanodisc formation.

II. EXPERIMENTAL

Epitaxial films were grown by molecular beam epitaxy system under a base pressure of $10^{-10}$ Torr. A 20-nm-thick Mo(110) buffer layer was grown on a Al$_2$O$_3$(1120) substrate at 1000°C, then a 3.3 nm Co film sandwiched by 20 nm Pt underlayer and 5 nm Pt overlayer was prepared at room temperature. The films were irradiated orthogonal to the sample surface with Ga$^+$ ions at the ion energy of 30 keV. The set of 21 irradiation fluences sequentially increasing from 0 to $1 \times 10^{16}$ ions/cm$^2$ steplike changed over contiguous stripes each 0.9 mm wide. Polar magneto-optical Kerr effect (PMOKE) was used to: (i) perform magnetization distribution imaging applying the CCD camera and special digital image processing procedures; (ii) determine local magnetic hysteresis loops. PMOKE remanent image was calculated as $[I_+(i,j) - I_-(i,j)]/[I_+(i,j) + I_-(i,j)]$, where $I_+(i,j)$ and $I_-(i,j)$ denote images recorded at zero field after sample saturation in $+H$ and $-H$, respectively; $H$ fields amplitude was larger than the saturation field.

The films were then investigated by means of the Co L-edge depth-resolved x-ray absorption spectroscopy (XAS) technique [11], at the soft x-ray undulator beamline, BL-16A, of the Photon Factory in the Institute of Materials Structure Science, High Energy Accelerator Research Organization (KEK-PF). In this technique, the electrons emitted after x-ray absorption were separately collected at different detection angles by using an imaging-type detector, which consists of a microchannel plate, a phosphor screen, and a CCD camera. The probing depth $\lambda_e$ of the XAS spectrum is determined by the effective escape depth of the electrons which depends on the electron emission angle $\theta_d$. The partial electron yield (PEY) mode with a retarding voltage of 500 V was adopted, so the Co LMM Auger electrons were mainly collected. The probing depth $\lambda_e$ was experimentally determined at each $\theta_d$ from the thickness dependence of the edge-jump intensity of the films [12]. We also adopted the total-electron-yield (TEY) mode, by which one can obtain the spectra with deeper $\lambda_e$ than that for PEY mode, i.e., $\lambda_e \sim 3.5$ nm for TEY, $\lambda_e \sim 0.5 - 1.5$ nm for PEY.

On the other hand, the Co L-edge XMCD spectra were recorded at room temperature in the TEY mode, in which the drain current from the sample was measured. All the XMCD measurements were performed in an applied magnetic field of 1 T. The applied magnetic field was parallel or antiparallel to the x-ray propagation direction. We adopted the normal and grazing incidence configurations for the XMCD measurements, in which the angle between the surface normal and the x-ray beam $\theta$ was 0 and $60^\circ$, respectively, in order to examine magnetic anisotropy of the film.

The Co K-edge EXAFS spectra were measured at the bending magnet beamline, BL-12C, of KEK-PF. All the spectra were recorded at 20 K in the fluorescence-yield mode with a probing depth $\lambda_e$ of 2–3 nm. The probing depth $\lambda_e$ was experimentally determined at each $\theta_d$ from the thickness dependence of the edge-jump intensity of the films [12]. We also adopted the total-electron-yield (TEY) mode, by which one can obtain the spectra with deeper $\lambda_e$ than that for PEY mode, i.e., $\lambda_e \sim 3.5$ nm for TEY, $\lambda_e \sim 0.5 - 1.5$ nm for PEY.

![FIG. 1. (a) Polar magneto-optic Kerr effect (MOKE) image for Al$_2$O$_3$(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga$^+$ ion fluences (ions/cm$^2$) taken at the remanent magnetization condition. The bright area corresponds to perpendicular magnetization. (b)–(f) Exemplary PMOKE magnetization loops measured at fluences $5 \times 10^{13}$, $1.5 \times 10^{14}$, $2.8 \times 10^{14}$, $1 \times 10^{15}$, $5 \times 10^{15}$ ions/cm$^2$. (g) Fluence dependence of maximal PMOKE rotation $\theta_{\text{MAX}}$ (■) and remanence of PMOKE rotation $\theta_{\text{REM}}$ (●). Dashed lines are the guides to the eye.](image-url)
19-element solid-state detector. To examine crystallographic anisotropy in the Co thin film, the EXAFS spectra were taken at normal and grazing incidence configurations at $\theta = 0$ and 45°, respectively.

III. RESULT AND DISCUSSION

The grayscale image in Fig. 1(a) is a PMOKE remanent image of the steplike Ga$^+$ irradiated sample after saturation in perpendicular direction to the sample surface. Bright areas indicate regions with out-of-plane magnetization component, while the dark areas show in-plane magnetic anisotropy, paramagnetic state, or multidomain structure with out-of-plane magnetization component. Exemplary PMOKE magnetization loops measured for selected fluences are presented in Figs. 1(b)–1(f). Two parameters were distinguished from these loops: maximal PMOKE rotation $\theta_{\text{MAX}}$ and remanence PMOKE rotation $\theta_{\text{REM}}$. Profiles of these parameters are shown below the grayscale remanent image in accordance with irradiation fluences, marked on the $x$ axis. Two regions with increased remanence are observed at Ga$^+$ ion fluences of $1$–$1.5 \times 10^{14}$ and $2$–$4 \times 10^{15}$ ions/cm$^2$. These regions are connected with regions of increased magnetic anisotropy. Although the fluences are slightly different from those observed in Ref. [7], in which the sample configuration is different from the present one, essentially the same tendency was observed in both cases. Moreover, we also measured longitudinal magneto-optical Kerr effect, and hysteresis behavior and remanent magnetization were observed for the nonirradiated film and the film at $2.8 \times 10^{14}$–$1 \times 10^{15}$ ions/cm$^2$ (not shown), which indicates that in-plane magnetic anisotropy is realized in these regions. Our previous EXAFS analysis shows that the transition at lower fluences is caused by anisotropy of Co orbital moment due to structural strain [8]. To reveal the origin of the transition at higher fluences, we look at the chemical state of Co.

Figure 2 shows Co L-edge XAS data for Al$_2$O$_3$ (1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films taken in the TEY mode at normal incidence configuration. At Ga$^+$ ion fluence below $1 \times 10^{15}$ ions/cm$^2$, absorption edge jump is smaller than that of the others, indicating that Co layers are buried under the Pt overlayer. While in the medium range of $1 \times 10^{15}$ ions/cm$^2$, absorption intensity increases due to a removal of the overlayer, and again the intensity gets smaller because the amount of Co decreases due to resputtering by heavy ion irradiation more than $1 \times 10^{15}$ ions/cm$^2$. We then look at $\lambda_e$ dependence of the XAS spectra to see depth contrast. Co L-edge XAS for the film with characteristic Ga$^+$ ion fluences at $\lambda_e$ of 0.5 and 3.5 nm are displayed in Fig. 3. The spectrum at $1 \times 10^{15}$ ions/cm$^2$ has a Co metal-like feature at $\sim 781$ eV, as indicated by the arrow, and shows no $\lambda_e$ dependence. At higher fluences, two shoulders are seen at $\lambda_e = 0.5$ nm, which are characteristic of Co oxide, while the spectra at $\lambda_e = 3.5$ nm keep metal-like features. From these results, Co atoms are found to be in the metallic state beneath the cover layer at lower fluences, and with increasing the fluences, oxidation of Co proceeds from the surface due to the removal of the Pt cover layer. However, we do not find any changes between $5 \times 10^{15}$ and $1 \times 10^{16}$ ions/cm$^2$. 

![Co L-edge XAS (Normal Incidence)](image-url)

**FIG. 2.** Total-electron-yield Co L-edge XAS for Al$_2$O$_3$(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga$^+$ ion fluences (ions/cm$^2$) taken at the normal incidence configuration.

![Co L-edge XAS (Normal Incidence)](image-url)

**FIG. 3.** Probing depth ($\lambda_e$) dependence of Co L-edge XAS for Al$_2$O$_3$(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga$^+$ ion fluences (ions/cm$^2$) taken at the normal incidence configuration.
although magnetic state is different from each other, i.e., ferromagnetism at $5 \times 10^{15}$ and paramagnetism at $1 \times 10^{16}$ ions/cm$^2$. Then we perform XMCD analysis to investigate local magnetic state of Co.

A. XMCD analysis

Figure 4 shows Co L-edge XAS spectra taken with opposite circular polarizations, $\mu_+ \text{ and } \mu_-$, and XMCD spectra, $\mu_+ - \mu_-$, for Al$_2$O$_3$(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with characteristic Ga$^+$ ion fluences taken under a magnetic field of 1 T. XMCD signal was observed up to $5 \times 10^{15}$ ions/cm$^2$, while for the highest fluence the film shows paramagnetic state. In addition, almost the same intensity between two incidence angles is observed for all the films. This indicates that the films are magnetically saturated in both the normal and in-plane directions at 1 T, thus we can determine anisotropy of the orbital magnetic moment. We estimate the spin magnetic moment $m_s$, and the perpendicular and in-plane components of the orbital magnetic moments $m_l$ from the angle dependent XMCD analysis [13,14]. According to the angle-dependent XMCD spin and orbital sum rules [15,16], we can obtain both the spin and orbital magnetic moments, as described in Ref. [8].

The estimated $m_s$, as well as the perpendicular and in-plane components of $m_l$ are shown in Fig. 5. Here we show the ratio $m_l/m_s$ to make quantitative discussion about ion fluence dependence. We also estimate the difference between the perpendicular and in-plane components of $m_l/m_s$ as functions of ion fluence. In the shadowed area, the film shows PMA. $m_s$ slightly decreases from $\sim 1.2$ to $\sim 1.0 \mu_B$ with increasing the ion fluences. $m_s$ at $1.5 \times 10^{15}$ ions/cm$^2$ is $\sim 25\%$ smaller than that observed in Ref. [8], and this could be due to the difference of the sample configuration, e.g., Co thickness and structural disorder. For $m_l/m_s$, larger anisotropy is observed at $1.5 \times 10^{15}$ ions/cm$^2$, and the anisotropy decreases with increasing the ion fluences. We suppose that larger anisotropy at $1.5 \times 10^{14}$ ions/cm$^2$ is induced by structural strain in Co and that the first PMA phase is enhanced by the strain, as discussed in our previous report [8]. On the other hand, although the orbital moment anisotropy is smaller at $5 \times 10^{15}$ ions/cm$^2$ compared with $1 \times 10^{15}$ ions/cm$^2$, as indicated in Fig. 5, the magnetic anisotropy changes from in-plane to perpendicular between these fluences. This means that the emergence of the second PMA is not attributed to the orbital moment anisotropy. To confirm the mechanism of the second PMA phase, we perform the structural analysis by EXAFS.

B. EXAFS analysis

Figure 6 shows Fourier transforms of the Co K-edge EXAFS function, $k\chi(k)$ (inset), for Al$_2$O$_3$(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga$^+$ ion fluences (ions/cm$^2$), taken at the normal and grazing incidence configurations. It should be noted that the phase shift effect in EXAFS is not corrected, so that the peak appears at a shorter distance than the actual one. Horizontal shift of the spectra by ion irradiation is not observed between $1 \times 10^{15}$ ions/cm$^2$ and $1 \times 10^{16}$ ions/cm$^2$, which indicates that there is no big change in the bond lengths. The broad peak at $\sim 2.1$ Å is composed of the first nearest neighbor Co-Co and Co-Pt components, and its amplitude becomes smaller with increasing the Ga$^+$ ion fluences. Because any Co-O components are not observed here, the oxidation degree of Co is small. From this finding, we suppose that two shoulders observed in Fig. 2 are attributed to the surface oxidation of Co, while the inside of the film keeps a metallic Co state. On the other hand, the sharp peak at $\sim 2.8$ Å is attributed to the Co-Pt bond, and it gets larger with increasing the ion fluences due to the intense intermixing between Co and Pt atoms. Surprisingly, at a distance of 3 Å to 5.5 Å, a clear peak structure is observed.
and it does not largely change by ion irradiation. We suppose that Co and Pt keep crystalline fcc and/or hcp framework.

We further analyze the EXAFS data including the second and third coordinated atoms. EXAFS curve fitting is performed by IFEFFIT package [17,18] in \( k \) space after the inverse Fourier transformation for the nearest Co-Co and Co-Pt contributions (<5.1 Å). The EXAFS spectra for the Co-Pt alloy [19] calculated by FEFF8 [20] code are used as a theoretical standard. In-plane and out-of-plane bond lengths are separately determined by simultaneous fitting of the EXAFS data taken both at 0° and 45°. We separately analyze the parameters of fcc (111) and/or hcp (0001) in-plane and out-of-plane bonds for the first coordination atoms, while the second coordination atoms only consist of the out-of-plane bonds. For the third coordination atoms, in-plane and out-of-plane parameters have to be separately considered, but we approximately treat them as a single parameter. Experimental and fitted Co K-edge EXAFS functions, \( k^2 \chi (k) \), are displayed in Fig. 7, and obtained structural parameters are showed in Table I. The Co-Co and Co-Pt bond lengths are almost isotropic and slightly expand with increasing the Ga⁺ ion fluence. The coordination number (CN) of Co-Co shows rapid decrease with increasing the ion fluence, and the nearest Co-Co almost vanishes at \( 1 \times 10^{15} \) ions/cm², while the CN of Co-Pt increases with increasing the ion fluence due to strong intermixing between Co and Pt atoms. \( f \) is a parameter for the ratio of fcc phase in the films, and we find that the fcc structure dominates in the films at all the ion fluences.

Now we carefully look at the CN analysis. Figure 8 displays first, second, and third nearest-neighbor CNs, \( N_1 \) (a), \( N_2 \) (b), and \( N_3 \) (c), respectively. The in-plane and out-of-plane coordinated atoms are separately obtained from the polarization-dependent EXAFS data for the first nearest-neighbor atoms (a). At \( 1 \times 10^{15} \) ions/cm², the out-of-plane \( N_1 \) of Co-Co is much smaller than the in-plane one. In contrast, in-plane \( N_1 \) of Co-Co rapidly decreases, while that of Co-Pt increases at \( 5 \times 10^{15} \) ions/cm², and the estimated ratio of Co and Pt is \( \sim 1 : 3 \). Please note that the obtained \( N_1 \) is smaller than that estimated for the ideal fcc and/or hcp structure. We suppose that this is due to structural disorder of Co. Next, if we look at \( N_2 \), there is no Co-Co contribution and only Co-Pt contribution is seen at all the ion fluences, while for \( N_3 \), Co-Co contribution is only observed at \( 1 \times 10^{15} \) ions/cm², and it vanishes at higher fluence. At \( 1 \times 10^{16} \) ions/cm², there

![Fourier transforms of the Co K-edge EXAFS function, \( k^2 \chi (k) \), for Al₂O₃(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga⁺ ion fluences (ions/cm²), taken at the normal (a) and grazing (b) incidence configurations.](image)

FIG. 6. Fourier transforms of the Co K-edge EXAFS function, \( k^2 \chi (k) \) (inset), for Al₂O₃(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with indicated Ga⁺ ion fluences (ions/cm²), taken at the normal (a) and grazing (b) incidence configurations.

![Comparison between the experimental (Exp.) and fitted (Fit) Co K-edge EXAFS functions, \( k^2 \chi (k) \), for Al₂O₃(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with different Ga⁺ ion fluences, 1 \times 10^{15} \) (a), \( 5 \times 10^{15} \) (b), and \( 1 \times 10^{16} \) ions/cm², taken at the grazing incidence configuration.](image)

FIG. 7. Comparison between the experimental (Exp.) and fitted (Fit) Co K-edge EXAFS functions, \( k^2 \chi (k) \), for Al₂O₃(1120)/Mo(20 nm)/Pt(20 nm)/Co(3.3 nm)/Pt(5 nm) thin films with different Ga⁺ ion fluences, 1 \times 10^{15} \) (a), \( 5 \times 10^{15} \) (b), and \( 1 \times 10^{16} \) (c) ions/cm², taken at the grazing incidence configuration.
TABLE I. Structural parameters for a Pt/Co/Pt film with indicated Ga+ ion fluences (ions/cm²) obtained by fitting the Co K-edge EXAFS data after the inverse Fourier transform for the nearest Co-Co and Co-Pt contributions. \( r, N, \) and \( \sigma^2 \) represent bond length, coordination number, and Debye-Waller factor, respectively. We assume that the energy shift of the absorption edge \( \Delta E_0 \) of the Co-Co shell is equivalent to that of the Co-Pt shell. \( f \) is a ratio of fcc phase in Co films.

<table>
<thead>
<tr>
<th>Ion Fluence (ions/cm²)</th>
<th>Co-Co bond</th>
<th>Co-Pt bond</th>
<th>( \Delta E_0 ) (eV)</th>
<th>( f )</th>
</tr>
</thead>
<tbody>
<tr>
<td>( 1 \times 10^{15} )</td>
<td>r(Å) 2.656(45) N 2.60(40) ( \sigma^2 ) 0.0163(47)</td>
<td>r(Å) 2.692(11) N 2.33(20) ( \sigma^2 ) 0.0030(11)</td>
<td>(-2.3(2))</td>
<td>0.93(29)</td>
</tr>
<tr>
<td></td>
<td>In-plane</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>Out-of-plane</td>
<td>2.648(45)</td>
<td>1.22(18)</td>
<td>0.0040(47)</td>
</tr>
<tr>
<td>( 5 \times 10^{15} )</td>
<td>In-plane</td>
<td>2.682(70)</td>
<td>1.25(32)</td>
<td>0.0146(52)</td>
</tr>
<tr>
<td></td>
<td>Out-of-plane</td>
<td>2.651(70)</td>
<td>0.77(25)</td>
<td>0.0084(52)</td>
</tr>
<tr>
<td>( 1 \times 10^{16} )</td>
<td>In-plane</td>
<td>2.690(72)</td>
<td>0.11(10)</td>
<td>0.012(12)</td>
</tr>
<tr>
<td></td>
<td>Out-of-plane</td>
<td>2.604(93)</td>
<td>0.27(40)</td>
<td>0.006(12)</td>
</tr>
</tbody>
</table>

is no Co-Co contribution but Co-Pt bond is dominating, which indicates that Co exists as a localized atom dispersed in Pt. Such local environment cannot keep the magnetic interaction, so the paramagnetic phase appears at the highest fluence.

Schematic illustration of the film structure is depicted in Fig. 9. We suppose that the Co layer is dispersed as single-atom-thick sheets in the Pt matrix at \( 1 \times 10^{15} \) ions/cm², in which the film shows in-plane magnetic anisotropy. At \( 5 \times 10^{15} \) ions/cm², Co sheets are transformed to the nanodiscs with several atoms in the in-plane direction, and such nanodisc formation would enhance out-of-plane magnetic component. Actually, the appearance of two-dimensional Co nanoclusters (~10 Å) parallel to the surface was reported in CoPt₃ disordered alloy [21], and the authors assumed that such clustering is responsible for PMA. Furthermore, Maret et al. reported the formation of (111) alternate Pt-rich and Pt-poor planar regions in the CoPt₃ film, and they propose that such microstructure enhances PMA [22]. In both cases, magnetic properties were controlled by changing the growth temperature of the films. We suppose that the ion irradiation and heating exhibit a similar impact to the CoPt alloy, and the derived nanostructure, i.e., Co nanodisc, induces PMA. If we assume that the classical electromagnetic theory is valid for nm-scale magnets, it is estimated from the obtained structural parameter that the effect of the shape anisotropy is reduced by ~30% due to the nanodisc formation compared to that of the layered structure. The weakening of the shape anisotropy assists the enhancement of the second PMA. In fact, the effect of decrease in the shape anisotropy is larger than that in the anisotropy of \( m_I \) from the first (\( 1.5 \times 10^{14} \) ions/cm²) to the second (\( 5 \times 10^{15} \) ions/cm²) PMA phases. Strictly speaking, such small nanodiscs cannot be in the ferromagnetic state because

![FIG. 8](image_url)

FIG. 8. First, second, and third nearest-neighbor coordination numbers, \( N_1 \) (a), \( N_2 \) (b), and \( N_3 \) (c), respectively, around the Co atom. The in-plane and out-of-plane coordinated atoms are separately analyzed from the polarization-dependent EXAFS data for the first nearest-neighbor atoms (a).
of lower Curie temperature if they are magnetically isolated with each other. Therefore, we suppose that magnetic coupling between Co nanodiscs and Pt atoms causes ferromagnetism in the film. Although a simple estimation of the shape anisotropy based on the classical electromagnetic theory is not valid in the present case, similar shape effect can affect the magnetic anisotropy of the film. Therefore it might be possible that PMA is realized by the shape effect due to Co nanodisc formation in the Pt matrix. We thus find that the origin of the first and second PMA phases is different, i.e., strain and nanodisc formation, respectively. Although the details of the electronic state of Co are not discussed in this paper, a size effect, hybridization between Co and Pt, and surface state have to be considered for the overall understanding of the ion irradiation-induced PMA.

IV. SUMMARY

We have investigated the origin of magnetic phase transition from in-plane to perpendicular magnetic anisotropy of Pt/Co/Pt thin film by Ga\(^{+}\) ion irradiation at fluences of 1–5 \times 10^{15} ions/cm\(^2\) by means of XMCD and EXAFS analyses. Pt and Co atoms were found to be mixed with each other, and Co was oxidized near the surface due to Pt overlayer removal. Moreover, polarization-dependent EXAFS analysis revealed that first Co was dispersed as separated sheets in a Pt matrix at 1 \times 10^{15} ions/cm\(^2\), and then the Co sheets were divided into clusters with a few Å in diameter at 5 \times 10^{15} ions/cm\(^2\), which were regarded as nanodiscs, associated with the appearance of out-of-plane magnetization component. Because we did not observe an enhancement in anisotropy of Co orbital moment through the transition at about 5 \times 10^{15} ions/cm\(^2\), it might be possible that such nanodisc formation induces PMA via the shape effect. By comparing with the phase transition observed at lower fluence, we found that the mechanism of two transitions is different, i.e., the transition at lower fluence is caused by anisotropy of orbital moment due to structural strain of Co, while the second transition is possibly by shape effect due to Co nanodisc formation.

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