Large Negative Magnetic Anisotropy in Epitaxially Grown Fe/Co Multilayer Films

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We have investigated the relationship between the crystal structure and the magnetic anisotropy constant of epitaxially grown Fe/Co multilayers (ML). The Fe/Co ML grown on a Ag (100) underlayer at ambient temperature exhibits body centered cubic (bcc) structure with (100) orientation for the stacking period \( \lambda \) less than 15 nm. The saturation field along the film normal is much larger than the demagnetization field \( 4\pi M_s \), indicating negatively large magnetic anisotropy \( K_u \). The negative \( K_u \) takes the peak value as large as \( 1 \times 10^7 \) erg/cc at \( \lambda = 0.75 \) nm. This behavior of \( K_u \) against \( \lambda \) is very similar to that reported for Fe/Co (110) polycrystalline ML [Vas’ko et al., Appl. Phys. Lett. 89, 092592 (2006)]. Moreover, slight lattice distortion is observed, but the magnetoelastic contribution is much smaller than the negative \( K_u \). These results lead to the conclusion that the negative \( K_u \) of Fe/Co ML is attributable to the interfacial magnetic anisotropy with weak crystal orientation dependence.

Key words: Fe/Co multilayer, magnetic anisotropy, epitaxial growth

1. Introduction

A magnetic material with large uniaxial magnetic anisotropy energy (MAE), \( K_u \), is technologically important because of its application to magnetic recording media and spintronic random access memory cells. On the other hand, the negative \( K_u \) is also technologically useful, which can be utilized for soft underlayers of perpendicular recording media and high frequency magnetic devices.

It is well known that CoIr alloy exhibits a large negative \( K_u \) of \(-8 \times 10^6\) erg/cc\(^1,2\). Recently, the negative \( K_u \) comparable to that of CoIr has been observed in polycrystalline (110) oriented Fe/Co multilayers (ML) prepared by sputtering [3]. It is noteworthy that large negative \( K_u \) is realized in Fe/Co ML even in the absence of expensive 4d and 5d elements which are in general very effective to enhance the MAE due to its large spin-orbit coupling. This feature of Fe/Co ML is favorable from practical viewpoints and may give a hint for designing new magnetic materials.

Vas’ko et al. speculated that the large negative \( K_u \) of Fe/Co ML is mainly due to interface anisotropy\(^3\). However, there still remain some open questions about the origin of the large negative \( K_u \). The first is how the crystal orientation and lattice strain affect the MAE. The second is how interdiffusion at the Fe/Co interfaces influences the MAE. Recent \( \textit{ab initio} \) calculations\(^4,5\) have revealed that the lattice distortion significantly affects the MAE of Fe-Co alloy which may be formed due to interdiffusion at the interfaces of Fe/Co ML. Therefore, we have to be careful about the interdiffusion at the Fe/Co interfaces. As mentioned above, very detailed structural characterizations are necessary to clarify the origin of the large negative \( K_u \) of Fe/Co ML. For such purposes, epitaxially grown Fe/Co MLs are most suitable rather than the polycrystalline samples used in the previous study\(^3\).

In this study, to clarify the origin of the large negative \( K_u \) of Fe/Co ML, we have used the epitaxially grown samples and elaborately examined the correlation between the MAE and the crystal structure.

2. Experiment

The Fe/Co ML was deposited on a Ag (100) single crystal underlayer using the evaporator cells built in a molecular beam epitaxy (MBE) system with the base pressure \(< 2 \times 10^{-7} \) Pa. The Ag underlayer was fabricated on a HF-treated Si (100) substrate by using an e-beam evaporator. Firstly a 1 nm-thick Ag layer was deposited at 350 °C, and after cooling it down to ambient temperature, a 40 nm-thick Ag layer was subsequently deposited. Finally, heat treatment at 350 °C for 30 min was carried out. The above procedure can produce a high quality Ag single crystal film with the mean square surface roughness \(< 0.2 \) nm\(^6\). An Fe/Co ML film was epitaxially grown on the Ag underlayer at

![Fig. 1 XRD profile and RHEED image of Fe/Co ML with stacking period \( \lambda = 0.75 \) nm.](image)
room temperature. The layer thicknesses of Fe ($t_{Fe}$) and Co ($t_{Co}$) were equal, and the stacking period $\lambda$ ($= t_{Fe} + t_{Co}$) was varied from 0.3 to 15 nm. The total thickness of Fe/Co ML was fixed at 30 nm for each $\lambda$ by adjusting the stacking number. After the deposition of Fe/Co ML, a 10 nm-thick Ag capping layer was deposited as a protective layer.

The crystal structure was studied by reflection high energy electron diffraction (RHEED) and x-ray diffraction (XRD) with Cu-Kα line. The in-plane and out-of-plane lattice constants $a$ and $c$ were determined from the several lattice spacing measured under off-normal and normal scattering vector configurations using a 4-axis goniometer7).

The effective anisotropy field $H_{keff}$ including the demagnetization field $4\pi M_s$ was evaluated from the saturation field of the vertical magnetization curve measured at 300 K by means of 4 terminal anomalous Hall effect (AHE) technique under the maximum field of 70 kOe. The magnetization $M_s$ was determined from the in-plane magnetization curves measured using a vibrating sample magnetometer (VSM).

3. Results and Discussions

Figure 1 shows the example of a XRD profile and a RHEED image of Fe/Co ML film. The streak pattern is clearly observed in the RHEED image, indicating that the Fe/Co ML is epitaxially grown on Ag (100). In the XRD profile, only one sharp diffraction line is observed around at 66 deg, indicating that the whole Fe and Co layers are coherently grown as a body centered cubic (bcc) structure. For $\lambda \geq 15$ nm, however, some diffractions due to hcp-Co were detected. Therefore, it is concluded that the crystal uniformity becomes deteriorated for $\lambda \geq 15$ nm.

Figure 2 shows the magnetization $M_s$ as a function of $\lambda$. The $M_s$ monotonically increases from the simple sum of Fe and Co moments toward the value of bulk FeCo alloy8) with the decrease of $\lambda$. The large $M_s$ for $\lambda \leq 1$ nm suggests two possibilities: one is the alloying due to interdiffusion at the interfaces, and the other is the enhanced magnetic moments of Fe and Co at the interfaces due to the proximity effect. In our deposition method, the former possibility is eliminated because of low particle energy and low substrate temperature in the present deposition conditions. Therefore, we believe the layered structure in our Fe/Co ML remains even for very thin $\lambda$, yielding enhanced magnetization. This speculation is supported by another experimental result as will be mentioned later.

Figures 3 (a) and 3 (b), respectively, show the axial ratio $c/a$ and lattice constants $a$ and $c$ as functions of $\lambda$. The $c/a$ is always smaller than 1 over the whole range of $\lambda$, indicating tetragonal distortion of Fe/Co ML. The simply averaged lattice constant of bcc-Fe (0.2866 nm) and bcc-Co (0.2819 nm)9) is 0.2843 nm, which is smaller than that of Ag ($a/\sqrt{2} = 0.2889$ nm). Therefore, the in-plane lattice matching of Fe/Co with Ag may give rise to slight $c$-axis contraction, as shown in Fig. 3 (b). It should be also noted that the $a$ is about 0.288 ~ 0.286 nm which agrees with that of Ag. Using the Poisson’s ratio of bcc-Fe, the $c$ of Fe/Co lattice is calculated as 0.280 nm based on the simple elastic theory, which is in good agreement with the experiment.

Figure 4 (a) shows the vertical AHE curves for $\lambda = 0.5$, 0.75, and 15 nm, respectively. The all magnetization curves exhibit linear shapes typical of hard axis magnetization curves for uniaxial anisotropy. Note that the saturation fields for $\lambda = 0.5$ and 0.75 nm are obviously larger than that for $\lambda = 15$ nm. The effective anisotropy field $H_{keff} (= saturation field)$ is
plotted as a function of $\lambda$, as shown in Fig. 4 (b). The $H_{keff}^{\text{eff}}$ is much larger than the demagnetization field $(-4\pi M_s)$ and takes a peak value of $-35$ kOe at $\lambda = 0.75$ nm. The magnetic anisotropy constant $K_o$ derived from the $H_{keff}^{\text{eff}}$ is plotted in Fig. 5. The $K_o$ reaches $-1\times10^7$ erg/cc, which is negatively larger than that of CoIr$^{11}$. Note that the lattice constants of Fe/Co ML are almost constant in this range of $\lambda$, as seen in Fig. 3 (a). This fact strongly suggests that the magnetoelastic anisotropy $K_{el}$ is not a major origin of the large negative $K_o$ of Fe/Co ML. Let us estimate the $K_{el}$ in the presence of bi-axial strain. In this case, $K_{el}$ is given by

$$K_{el} = -\frac{3}{2} \lambda_{100} \frac{E(1-c/a)}{2\sigma+(1-\sigma)c/a},$$

where $\lambda_{100}$ is the magnetostriction coefficient along $<100>$ direction, $E$ is the Young’s modulus, and $\sigma$ is the Poisson’s ratio, respectively. The calculated $K_{el}$ for the experimentally determined $c/a$ and the $\lambda_{100}$ of FeCo alloy$^{11}$ is much smaller than the experimental $K_o$ except for $\lambda \sim 0.3$ nm and 1.5 nm, as shown in Fig. 5. Considering the excellent crystal uniformity over this range of $\lambda$, it can be deduced that the interface magnetic anisotropy due to the layered structure of Fe/Co ML is the main origin of the large negative $K_o$. The decrease of $K_o$ for $\lambda \leq 0.5$ nm is probably attributed to the deterioration of layered structure.

Finally, we compare our result for (100) oriented epitaxial Fe/Co MLs to that for (110) polycrystalline ones$^{9}$. In Fig. 5, the $K_o$ reported in Ref. [3] is also plotted. Although the $K_o$ in this study is somewhat larger than that of Ref. [3], these two data exhibit similar behavior. This fact suggests that the interface anisotropy of Fe/Co ML has weak dependence on the crystal orientation.

4. Summary

We have investigated the relationship between the crystal structure and the magnetic anisotropy energy (MAE) of epitaxially grown Fe/Co multilayers (ML). The crystal structure of Fe/Co ML is the (100) oriented bcc structure with slight tetragonal distortion along the film normal. The large negative perpendicular magnetic anisotropy $K_o$ is observed in vertical magnetization curve measurements. The $K_o$ takes a peak value as large as $-1\times10^7$ erg/cc at the stacking period of 0.75 nm. This behavior of $K_o$ is very similar to that previously reported for the (110) oriented polycrystalline Fe/Co ML, and the value of $K_o$ cannot be explained by the magnetoelastic effect. Thus, it is concluded that the large negative $K_o$ of Fe/Co multilayer is attributed to the interface magnetic anisotropy with weak orientation dependence.

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