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ABSTRACT

Magnetic properties of Co film in Pt/Co/α-Cr_2O_3/Pt/α-Al_2O_3 structure were investigated. Co layer thickness t_{Co} dependence of perpendicular magnetic anisotropy energy density K reveals that the bulk magnetic anisotropy plays an important role in the system in addition to the interfacial anisotropy. Damping constant α monotonically increases with the decrease of t_{Co} but not proportionally to $1/t_{Co}$. Both K and α increase with the increase of Pt layer thickness t_{Pt} from 3 nm to 5 nm and keeps almost constant in the t_{Pt} range between 5 nm to 20 nm. These results are of importance to understand the magnetization switching behavior driven by the magneto-electric (ME) effect as well as to design the spintronics device using the ME effect.

I. INTRODUCTION

Modulation of magnetic properties by means of electric-field is attracting much attention for application to low power electronics. The electric-field effect on the magnetic properties has been first observed in magnetic semiconductors, followed by in metallic systems.[1–12] The electric-field induced magnetic anisotropy has been studied most intensively owing to its possibility for non-volatile memory application as it can switch the magnetization direction in nanomagnet.[10,11] Recently, the other scheme to manipulate the magnetization direction has been also studied, that is, so called magneto-electric (ME) switching.[13–22] The ME switching of magnetization enables the direct coupling between the magnetization and the electric field using ME materials whose magnetization can be controlled by the electric field, and vice versa, the electric polarization can be controlled by the magnetic field. This scheme enables one to achieve the electric-field-induced switching the magnetization of a ferromagnetic (FM) layer by the exchange coupling with an antiferromagnetic (AFM) layer at the FM/AFM interface through the change in the polarity of exchange bias. Among variety of ME materials, Cr_2O_3 is one of the most intensively studied systems showing the global magnetization switching up to room temperature as demonstrated in the bulk system.[14] In particular, the demonstrations in all-thin-film Pt/Co/α-Cr_2O_3 systems with the perpendicular magnetic anisotropy (PMA) have recently achieved opening the potential for device application as well as understanding their rich fundamental physics regarding ME effect.[15–22] Since the PMA plays dominant role in determining the switching behavior using the ME effect in the system, a detail investigation of magnetic properties for the Co layer in this system is required. So far there are some papers reporting on the magnetic properties of the Co layer not only in Pt/Co stack films,[23] but also in Pt/Co multilayers[24] with the PMA. However, since the magnetic properties, especially the dynamic magnetic properties such as the damping constant is very sensitive with the

stacking structure and the interfacial condition. Therefore, a detailed investigation is important to realize the application of Cr_2O_3-based ME switching. As it is well known, both of the interfacial anisotropy and the bulk anisotropy are present in the Pt/Co system. Previously, those anisotropies in the Pt/Co/α-Cr_2O_3 system have been quantitatively evaluated.[25] However, the relation with the damping constant, which characterizes magnetization dynamics,[26] has been lacked. For practical applications, high speed switching characteristics is of importance and the damping is one of factors governing it. In the previous report, it was shown that the ME-driven magnetic domain wall (DW) mobility in the Pt/Co/Au/α-Cr_2O_3 system was quite similar to the magnetic-field DW mobility of the Pt/Co/Pt system despite that the very small damping constant was predicted for the α-Cr_2O_3, which suggests that the switching dynamics in the Pt/Co/α-Cr_2O_3 system was governed by the magnetization dynamics of the Pt/Co. In the present study, to deepen understating those important magnetic properties for the switching using the ME effect with α-Cr_2O_3, we investigated the magnetic properties of the Co layer in the Pt/Co/α-Cr_2O_3 system.

II. EXPERIMENTAL METHODS

Pt-cap(t_{Pt})/Co(t_{Co})/α-Cr_2O_3/Pt-buffer layer stacked films were deposited by DC magnetron sputtering onto α-Al_2O_3 (0001) substrates. The base pressure of the sputtering chamber was below 5×10^{-7} Pa. First, the 20-nm thick Pt buffer layer was deposited on the α-Al_2O_3 (0001) substrate at 873 K. Then, the 200-nm thick Cr_2O_3 layer was formed on the Pt buffer layer by the reactive sputtering in the flow of an argon-oxygen gas mixture with the substrate temperature of 773 K. The Co layer and the Pt capping layer were subsequently deposited at room temperature. t_{Co} was varied from 0.6 nm to 1.1 nm with the fixed t_{Pt} = 3 nm, and t_{Pt} was varied from 3 nm to 20 nm with the fixed t_{Co} = 1 nm. More details on the sample preparation and the structural characterization were described elsewhere.[27]

The crystalline orientations of each layer were characterized by reflection high-energy electron diffraction (RHEED) with the acceleration voltage of 25 kV and the filament current of typically ~50 μA. As for the magnetic properties of the Co layer in this system, their magnetization measurements were investigated by a vibrating sample magnetometer (VSM), which provided the saturation magnetization M_S. An effective perpendicular magnetic anisotropy field H_K^{eff} and the damping constant α were evaluated by a broadband ferromagnetic resonance (B-FMR) measurement technique composing of a vector network analyzer and a coplanar waveguide. In the B-FMR experiment, the frequency range from 10 MHz to 50 GHz was used to excite FMR under the application of various external magnetic fields along the film normal direction. The FMR spectra were collected in the frequency domain at various fixed magnetic fields, after subtracting the background signal. Details for the FMR experiment was described elsewhere.[26] All the measurements were performed at room temperature.

III. RESULTS AND DISCUSSION

First, we show structural analysis results on the Pt-cap/Co/Cr_2O_3/Pt-buffer layer stacked on α-Al_2O_3 (0001) substrate. Typical RHEED images with the $[11\bar{2}0]_{substrate}$-azimuth obtained for the film with t_{Co} = 1.0 nm and t_{Pt} = 10 nm are shown in Fig. 1 with the clear streaks/spots. The RHEED patterns showing the crystal quality of each layers depend strongly on the deposition condition and the stacking structure as shown in our previous paper.[27] In this research, by using the deposition condition as described in the experimental part, the clear RHEED pattern was observed as a result of the good epitaxial structure between all layers. Also for the Co layer and Pt cap layer, the clear RHEED pattern could still be obtained even though they were deposited at room temperature at which the temperature dependence of the magnetic properties of the Co layer can be ignored. In addition, the good epitaxial relationship between Co (a = 5.057 Å)[28] and Cr_2O_3 (a = 4.958 Å; c = 13.586 Å)[29] would be attributed to their small lattice mismatch (1.96%), while (111) oriented Pt (a = 5.532 Å)[30] layer still could be obtained on Co, despite of their high lattice mismatch (9%). The diffraction pattern can be indexed by fcc(111) for Pt and Co, and corundum(0001) for Cr_2O_3, respectively.[27,31] According to the structural factor of the corundum structure, the $[11\bar{2}0]_{substrate}$-azimuthal diffraction pattern is asymmetric with respect to (0,0) streak. The observed symmetric diffraction pattern indicates that the twin boundaries along $[11\bar{2}0]$ are included in the film.[32]

Next, we show Co layer thickness dependence of magnetic properties. Figure 2(a) shows saturation magnetic moment per unit area as a function of Co layer thickness t_{Co}. We fit a linear function to the experimental results and obtained the saturation magnetization M_S of Co from its slope. M_S of the Co layer was determined to be 1503 emu/cm^3, which was very close to the bulk value, indicating no significant mutual diffusion between Co and Cr_2O_3 or Pt capping layer. As shown in Fig. 2(a), the intercept of the linear fit was positive. This can be ascribed to the proximity effect at the Pt/Co interface, which induces a finite magnetic moment in Pt at the interface. The value of induced moment of Pt was determined to be 3.5×10^{-5} emu/cm^2, which is comparable to the value in the other reports.[33,34] Someone might wonder at the non-continuity of the Co

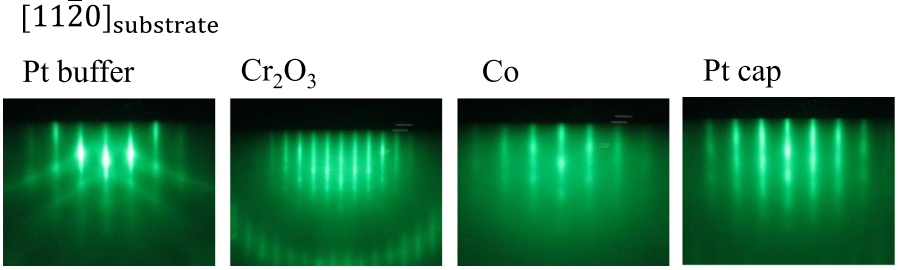

FIG. 1. RHEED images during deposition of a Pt(10 nm)/Co(1 nm)/Cr_2O_3/Pt stacked film with electron azimuths $[11\bar{2}0]$ of the α-Al_2O_3(0001) substrate.

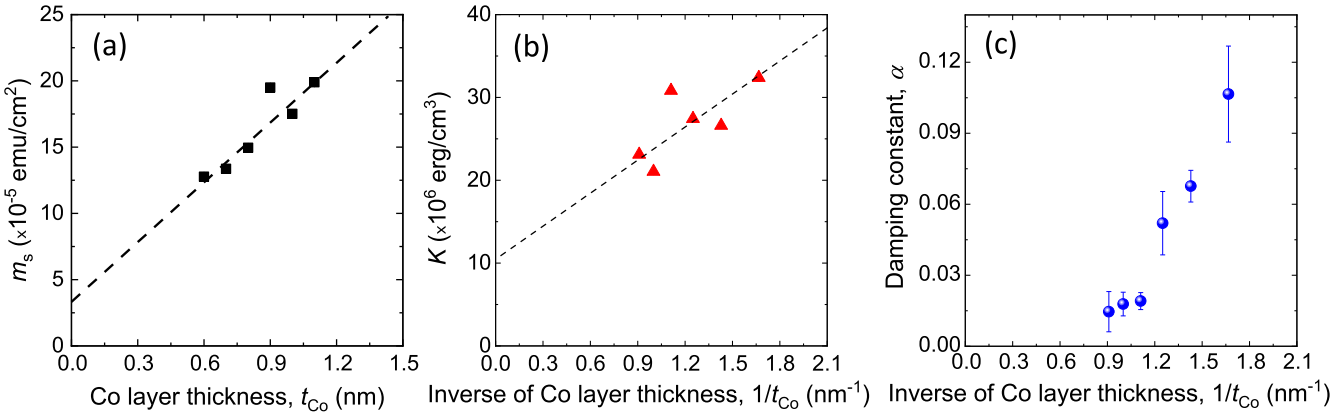

FIG. 2. (a) Co layer thickness dependence of (a) saturation magnetic moment per unit area m_S. Inverse of Co layer thickness dependence of (b) perpendicular magnetic anisotropy energy density K and (c) damping constant α.

layer in the low thickness region. If that is indeed a case in this study, magnetic anisotropy should be significantly reduced. However, as shown in Fig. 2(a) and Fig. 2(b), we did not see any anomaly in the Co thickness dependence of M_s and magnetic anisotropy. Thus, we infer that the Co film is still continuous at the studied thickness range. The intermixing sometime appeared in the system with small FM layer thickness[35] would also be exclude in these samples because we could obtain well-defined RHEED patterns and did not find any significant change in the VSM data of the samples with small t_{Co} which is similar with our previous reports.[16–22]

Figure 2(b) shows inverse of Co layer thickness dependence of perpendicular magnetic anisotropy energy density K. In the Pt/Co film with the interfacial anisotropy, three different components contribute to the magnetic anisotropy:[36,37]

$$K_{eff} = K_i/t_{Co} + K_b - 2\pi M_S^2$$

The first term is the interfacial anisotropy K_i, the second term is bulk magnetic anisotropy K_b originating from either crystalline anisotropy or magnetic elastic anisotropy), and the last term is the shape anisotropy. Thus, the magnetic anisotropy of the Pt/Co can be expressed as follows;

$$K = K_{eff} + 2\pi M_S^2 = K_i/t_{Co} + K_b$$

where K_{eff} is effective magnetic anisotropy energy density, M_S is the saturation magnetization taking into account the induced moment of Pt. K_{eff} was determined by the following relationship $M_S H_K^{eff}/2$, where M_S was measured by VSM and H_K^{eff} was measured by FMR. As shown in Fig. 2(b), perpendicular magnetic anisotropy energy density (including both bulk and interfacial contributions) increases with the increase of $1/t_{Co}$, indicating the presence of interfacial anisotropy. From a linear fit to the experimental results, K_i and K_b were obtained from the slope and intercept, respectively. The value of K_i was about 1.1 erg/cm^2, which was very close to the reported values with the Pt/Co on a-Cr_2O_3[25] and the fcc(111)-textured Co/Pt multilayer.[37] On the other hand, the value of K_b obtained from the intercept was 13.3×10^6 erg/cm^3, which was higher than the reported value for the Pt/Co/α-Cr_2O_3 system (12.7×10^6 erg/cm^3),[25] the Co/Pt multilayer and another system with oxide/FM/NM structure.[25,36,37] In the previous report on Pt/Co/α-Cr_2O_3 system,[25] the α-Cr_2O_3 layer thickness was 50 nm whereas 200 nm in the present study. The thick α-Cr_2O_3 can be robust and resultantly, the much more strain would be stored in the Co layer. Another possible explanation for the discrepancy is the difference in the crystal structure. In this study, we deposited the single Co film on α-Cr_2O_3 film using sputtering, whereas the Co/Pt multilayer was deposited on the Pt buffer layer using e-beam evaporation. Those difference could yield in the difference of the crystal structure of the Co layer, which could result in the difference in the bulk anisotropy as it originates from either the crystalline anisotropy or the magnetic elastic anisotropy.

Figure 2(c) shows the dependence of the damping constant α on the inverse of the Co layer thickness. α takes the same trend as that observed in K, i.e., α increases with the decreases of t_{Co} similar to the behavior of K. The α enhancement in a PMA system would be attributed to various origins such as the spin orbit coupling (SOC),[38] the spin-pumping.[39] The spin pumping effect should lead to a monotonous decrease of α with t_{Co}, but it was not observed in our system, i.e., α increased significantly in the low t_{Co} range whereas α only slightly changed in the high t_{Co} range. This result implied that the spin pumping is not the main contribution to the origin of α. The similar behavior was also seen in the previous report; however, the change in increasing rate of α below 1 nm was smaller than that in the present study.[23] Because the Pt/Co/Pt structure was used in the previous study, the difference might be related to the crystal structure and/or electronic state of the Co layer. Further study is required to elucidate the anomaly in the t_{Co} dependence of α.

Finally, we show capping layer thickness t_{Pt} dependence of the perpendicular magnetic anisotropy energy density K and α in the Figures 3(a) and (b), respectively. Both K and α increase with the increase of t_{Pt} from 3 nm to 5 nm and keeps almost constant in the t_{Pt} range between 5 nm to 20 nm. While the increase of K could originate from increase of bulk anisotropy as the increased Pt layer thickness is much larger than monolayer, it is yet to be elucidate.

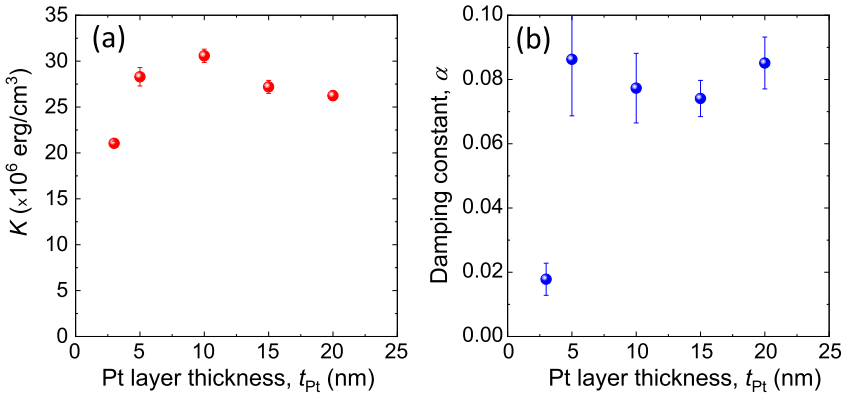

FIG. 3. Pt capping layer thickness dependence of (a) perpendicular magnetic anisotropy energy density K and (b) damping constant α.

Although no good correlation between K and α was observed in Co layer thickness dependence, it was observed in the Pt layer thickness dependence. This is similar with the result calculated in the previous report[40] for (001) fcc-Co/Pt bilayers in which the SOC has been taken into account as the origin of the magnetization dynamics. The result again suggests that the α enhancement in the thin Co layer thickness region does not originate solely from the spin-pumping, because the α enhancement with the increase of t_{Pt} was observed in much thinner region.[40,41]

IV. SUMMARY

The magnetic properties of the Co layer with the structure of Pt/Co/Cr_2O_3/Pt deposited on Al_2O_3 substrate were investigated. From the magnetization measurement, we found that the magnetic moment is induced in Pt at vicinity of the Pt/Co interface. Co layer thickness dependence reveals that the perpendicular magnetic anisotropy of the Co layer in the stack has its origin in the bulk anisotropy in addition to the interfacial anisotropy. Damping constant α increased with the decrease of the Co layer thickness, however, it was not in proportion to the inverse of Co layer thickness, indicating that the effects other than the spin-pumping is related to the α enhancement. Through the Pt layer thickness dependence of the magnetic properties, we found that the Pt capping layer also influences the magnetic anisotropy and damping constant. Both of the magnetic anisotropy and the damping constant increase with the increase of Pt layer thickness from 3 nm to 5 nm. In contrast to the Co layer thickness dependence, the good correlation between the magnetic anisotropy and damping constant was observed in the Pt layer thickness dependence. The results obtained in this study are of importance to understand the switching behavior using ME effect in α-Cr_2O_3 as well as to design the spintronics device using it.

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