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Temperature dependence of the magnetization reversal in Co(fcc)–BN–Co(poly hcp) structures

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The magnetic properties of multilayer structures with two magnetic layers of the same metal (Co) but with different crystallographic structures separated by an insulating BN layer have been studied. These structures were prepared on Si (001) substrates by a combination of molecular beam epitaxy (metallic layers) and electron cyclotron resonance-assisted sputtering (BN layer). An fcc Co single-crystal layer (60 Å) was first stabilized by growing it on a copper fcc buffer layer and subsequently a polycrystalline Co layer (70 Å) with hcp structure was grown on top of the insulating BN layer. A CoO antiferromagnetic layer, formed adjacent to this hcp Co layer, significantly influenced the magnetic behavior of the polycrystalline hcp Co layer. The magnetic hysteresis loops for these structures were measured at temperatures ranging from 5 to 350 K with the magnetic field applied along the easy (110) in-plane axis of the fcc Co. A very sharp flipping of the magnetization was found for the fcc Co layer with a nearly temperature-independent coercive field that increased from 14 mT below 100 K to 16 mT at 300 K. In contrast, the magnetization reversal in the hcp Co layer was smoother and its coercivity varied significantly with temperature depending on the strength of the exchange coupling with the adjacent CoO layer. At 5 K the coercivity was greater than 0.2 T and decreased with increasing temperature, becoming essentially zero above room temperature. When cooling in a magnetic field, an exchange offset was observed below 150 K that increased to about 0.1 T at 5 K. © *1999 American Institute of Physics.*
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I. INTRODUCTION

Magnetic structures intended for devices using spin-dependent tunneling junctions have two magnetic electrodes with different coercivities separated by a very thin insulating layer. According to standard theories,[1,2] a range of fields exists where the spins in both electrodes are antiparallel and the tunneling resistance as a function of the magnetic field will be larger. For all other field values, the spins in both electrodes are parallel and the resistance will have a lower value. Most researchers in this field use different magnetic materials for the two electrodes and/or the shape anisotropy to create the two different coercivities[3] or they use an antiferromagnetic layer to pin one of the ferromagnetic layers.[4,5]

In this article we report results on multilayer structures which were prepared[6] with two magnetic layers with different magnetic properties: a structure consisting of two magnetic layers from the same metal (Co) but with different crystallographic structure. In addition to the intrinsic difference between these layers, the magnetic behavior, in particular the coercive field of one of the layers, can be significantly modified by the exchange coupling with an adjacent antiferromagnetic layer.[7] In our case, an antiferromagnetic CoO (Néel temperature 292 K and blocking temperature around 150 K) layer is formed adjacent to the hcp Co layer. In this way we were able to study the magnetic behavior both at room temperature and at low temperatures below the antiferromagnetic transition where a strong exchange interaction between CoO and Co exists. At temperatures above the Néel temperature, the coercivity of the two ferromagnetic layers is still quite different even though the effect of the CoO is negligible. On the other hand, it was difficult to distinguish two different coercivities at low temperatures on samples without an antiferromagnetic layer and this would adversely affect our study of the spin-dependent tunneling at low temperatures.

To better understand and control the magnetic properties of such a multilayer structure, the process of magnetization reversal has been studied in detail over a wide temperature range where the strength of the interaction with the adjacent antiferromagnetic layer varies dramatically.

II. EXPERIMENT

The Co multilayer system essentially consisted of two ferromagnetic films with different crystalline structures separated by an insulating layer prepared by a combination of molecular beam epitaxy (metallic layers) and electron cyclotron resonance-assisted sputtering (BN layers). A Cu (3000 Å) single-crystalline seed layer was first grown on a Si (001) substrate at a rate of ~0.4 Å/s in several stages with intermediate annealings. This Cu single-crystal film grew epitaxially with an in-plane rotation angle of 45° with respect to the

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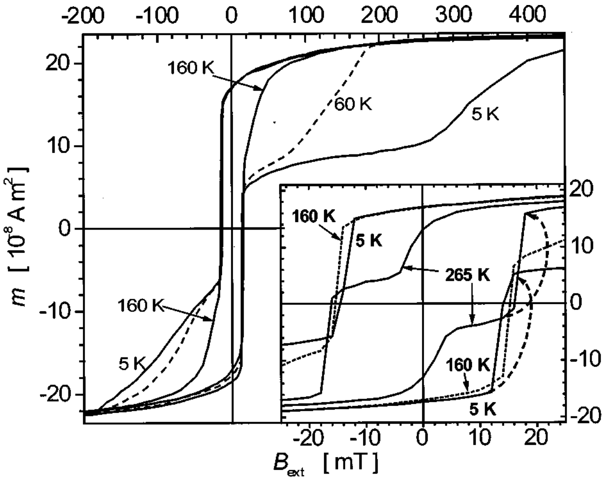

FIG. 1. Magnetic hysteresis loops at 5, 60, and 160 K after cooling from 390 K down to 5 K in a field of 2 T. Inset: details of the low-field MHLs at 5, 160, and 265 K. The curved dashed arrows mark the flip of the magnetization in the fcc Co layer.

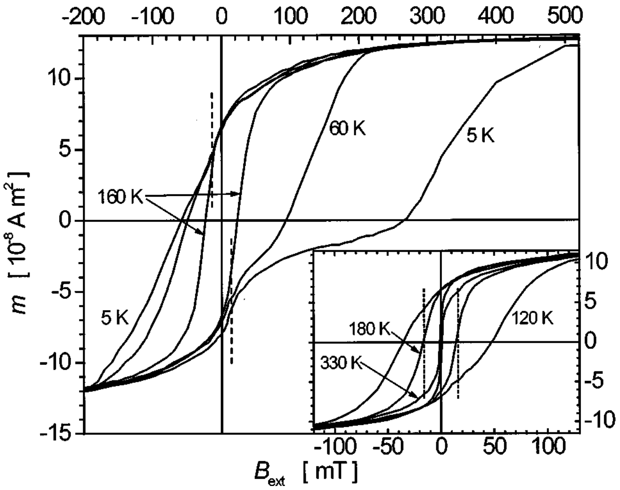

FIG. 2. Magnetic hysteresis loops associated with the hcp Co layer at 5, 60, 120, 160, 180, and 330 K. Vertical dashed lines indicate where these MHLs were ''sewed up'' after the contribution from the fcc Co layer was removed.

Si lattice[8] and served as a template to stabilize the subsequent growth of the fcc Co layer.[9] Reflection high-energy electron diffraction (RHEED) oscillations were used to monitor the growth of the single-crystalline fcc Co layer (60 Å). An insulating layer of polycrystalline hcp boron nitride (BN) (50 Å) was grown using ion-assisted sputtering. A second Co layer (70 Å) was then deposited in the same manner as the first Co layer, but in the absence of a single-crystal Cu template this Co layer grew polycrystalline and with an hcp structure. The top surface of this hcp Co layer was oxidized forming a thin CoO layer[10,11] before being capped by 30 Å of Cu.

An *ex situ* transmission electron microscopy (TEM) image of this multilayer structure showed that the insulating layer was continuous with no observable pinholes. High resolution TEM (HRTEM) confirmed the crystallographic structure of the various layers as previously determined by RHEED. Fourier transform analysis of the HRTEM images indicated an fcc structure for the initial Cu and Co layers, polycrystalline columnar growth for the BN layer, and an hcp polycrystalline structure with some degree of texture for the second Co layer.[6]

III. RESULTS AND DISCUSSION

The process of magnetization reversal in this multilayered system was studied using a Quantum Design superconducting quantum interference device magnetometer at temperatures ranging from 5 to 350 K with the magnetic-field B_{ext} applied along the easy (110) in-plane axis of the fcc Co. The typical sample size was $5\times 5\ \mathrm{mm}^2$. Figure 1 displays several magnetic hysteresis loops (MHLs) measured after cooling from 390 to 5 K in a magnetic field of 2 T. These MHLs exhibit a complex shape, which is strongly dependent on temperature with the hysteresis becoming wider and more nonsymmetrical at lower temperatures. The nonsymmetric MHL characteristic is more noticeable after the sharp flipping of the magnetization which occurs at fields $B_{\mathrm{ext}} = \pm B_{\mathrm{flip}}$. In addition, this nonsymmetric behavior is strongly dependent upon the cooling field, and could be due to an exchange bias from an antiferromagnetic layer pinning a ferromagnetic layer.[4] Since the antiferromagnetic CoO layer is adjacent to the hcp Co layer, it is reasonable to assume that this part of the MHL is associated with the hcp Co layer. In contrast, the sharp flipping of the magnetization which is more clearly seen in the inset of Fig. 1 (see the curved dashed arrows), is nearly temperature independent. B_{flip} is approximately 14 mT for temperatures below 100 K and increases to about 16 mT at 300 K. MHLs measured at higher temperatures, up to 350 K, are very similar to the loop at 265 K except for a slight decrease in the saturation magnetic moment m_s with increasing temperature. The m_s decreased by 4% from 5 to 250 K and 8.8% up to 350 K. The abrupt jump of the magnetic moment at B_{flip} is especially sharp at temperatures below 200 K with the magnitude of the jump being about 45% of m_s. Since the nominal thickness of the fcc Co layer is 60 Å out of a total Co thickness of 130 Å, or 46% of the total, these jumps correspond to the switching of the magnetization vector in the fcc single-crystalline Co layer.

The observation of two distinct, characteristic behaviors in the magnetization reversal permit the identification of the contributions to the MHLs from each individual Co layer. The sharp flip of the magnetization in the fcc Co layer and the smooth magnetization reversal in the hcp Co layer can be easily distinguished, particularly at lower temperatures. By shifting the MHLs on either side of the jumps in the magnetization so that the MHL branches below and above the jumps point towards each other, the resulting MHLs can be attributed to the hcp Co layer only. The magnitude of these shifts were scaled with $m_s(T)$. The resulting MHLs exhibit a nearly archetypal shape with very low coercivity above 200 K while the MHLs at lower temperatures are wide and asymmetrical as shown in Fig. 2. Vertical dashed lines indicate the location where these MHLs were ''sewed up'' after subtraction of the contribution from the fcc Co layer. The smooth shape of these MHLs is also consistent with properties expected for the polycrystalline structure of this Co layer and that the subtraction of the sharp contribution from the

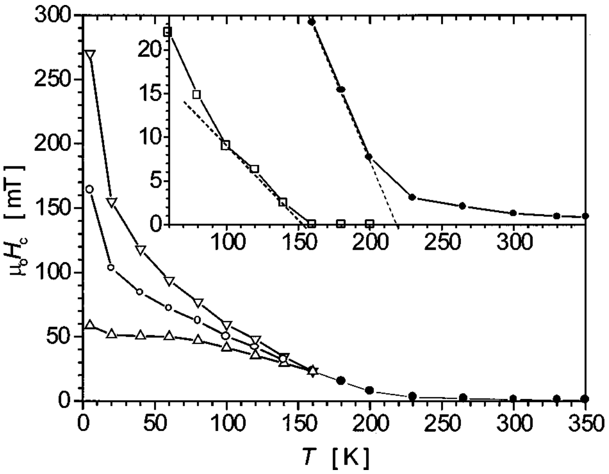

FIG. 3. Temperature dependence of the coercive field μ_0H_c (●). Below 150 K the coercive field for each MHL branch $\mu_0H_c^-$ (△) and $\mu_0H_c^+$ (▽) is plotted along with the average coercive field, $(\mu_0H_c^- + \mu_0H_c^+)/2$, (○). Inset: $\mu_0H_c(T)$ (●) and the exchange offset $(\mu_0H_c^- - \mu_0H_c^+)/2$ (□) at higher temperatures. Dashed lines are guides for the eye.

fcc Co layer was a reasonable approximation. In addition, the lower branches of the MHLs in Fig. 2 exhibit a smooth S-shape step close to the field of magnetization reversal in the fcc Co layer with a similar S shape being only partially visible on the upper MHL branches. This feature can be explained by a finite ferromagnetic Néel coupling between the two Co layers.

The temperature dependence of the coercive field μ_0H_c in the hcp Co layer is displayed in Fig. 3. $\mu_0H_c(T)$ is a just few mT at the higher temperatures and increases sharply below 220 K due to the interaction with the adjacent CoO layer. In addition to the increasing coercivity, the MHLs begin to be asymmetrical below 150 K resulting in different coercive fields for each MHL branch, which are shown as $\mu_0H_c^-$ and $\mu_0H_c^+$ along with the average coercive field, $(\mu_0H_c^- + \mu_0H_c^+)/2$ in Fig. 3. This difference, or correspondingly, the MHL offset shown as $(\mu_0H_c^- - \mu_0H_c^+)/2$ in the inset, arises from the exchange-coupled antiferromagnetic CoO layer being polarized by cooling in the presence of a magnetic field. A similar temperature dependence has been previously observed in a Co–CoO system.[7]

This difference in magnetic properties clearly indicates the potential of our two Co-layered system for spin-dependent tunneling devices. Magnetotransport studies on similar samples made with the two ferromagnetic layers separated only by a thinner BN insulating layer to assure tunneling between the magnetic layers is currently in progress.

IV. CONCLUSIONS

One type of magnetic structure intended for spin tunneling devices consists of two magnetic layers with different coercivities separated by an insulating layer. In this article we show results in samples prepared with two magnetic layers from the same metal (Co), but with different properties due to their different crystallographic structure. One layer was grown as single-crystalline fcc Co while the other magnetic layer was grown as polycrystalline hcp Co. One advantage of this type of magnetic layered structure is the presence of easy and hard axes arising from the crystallographic symmetry given by the in-plane fourfold anisotropy of the fcc Co layer.[12]

The magnetic hysteresis loops exhibited two distinct, characteristic behaviors in the magnetization reversal, which permitted the identification of the contributions to the MHLs from each individual Co layer. The single-crystalline fcc Co layer exhibited a nearly ideal switching of the magnetization when the field was applied along the easy axis. The coercive field changed only slightly between 14 and 16 mT in the temperature range between 5 and 350 K. In contrast, the magnetization of the hcp Co layer changed smoothly within a wide field range due to its polycrystalline structure. Moreover, the magnetic behavior of the hcp Co layer was further modified by the exchange coupling with an adjacent antiferromagnetic CoO layer, which significantly enhanced the coercivity at low temperatures. The coercive field of the hcp Co layer in the high-temperature range was a few mT and increased sharply below 220 K reaching 150 mT at 5 K. The MHLs of the hcp Co layer start to be asymmetrical below 150 K and the offset becomes greater than 100 mT at 5 K.

ACKNOWLEDGMENTS

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