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Engineering double-shifted hysteresis loops in Co/IrMn/Cu/Co films

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Co(3 nm)/IrMn(15 nm)/Cu(d_{Cu})/Co(7 nm) films were subjected to magnetic annealing where its temperature and duration as well as the direction and amplitude of the applied field were varied. We demonstrate that the exchange-bias field magnitude and sign of the subloop of the bottom-pinned Co layer can be tailored in a controlled manner allowing the whole hysteresis loop to be tuned from a double negatively/negatively shifted to a double negatively/positively shifted with the shifts of the subloops in antiphase. © *2009 American Institute of Physics*. [doi:10.1063/1.3227840]

The exchange-bias (EB) phenomenon[1–3] results from the magnetic coupling between a ferromagnet (FM) and partially uncompensated interfacial spins in an adjacent antiferromagnet (AF). During the past decades, EB has been extensively studied owing to the fascinating physics involved and also due to its applicability in magnetoelectronic devices. The most known EB manifestations are the hysteresis loop shift along the field axis, called EB field, H_{EB}, and an increase of the coercivity, H_C. The EB is defined (initialized) by a magnetic field applied during the fabrication of the sample, a thermal postannealing, or ion bombardment;[4,5] if the FM has very low anisotropy, even its remnant magnetization is able to set the EB.[6] Tailoring the magnetic properties of EB systems by using postdeposition treatment is very appealing from a technological point of view as well as for a better understanding of the microscopic mechanisms leading to the effect and their parameters.[7]

In the present letter, we demonstrate that in Co/IrMn/Cu/Co films the EB field magnitude and sign of the subloop of the bottom-pinned Co layer can be tuned in a controlled manner by means of appropriate magnetic annealing. This allows the hysteresis loop to be changed from a double negatively/negatively shifted to a negatively/positively shifted one, which can be useful for development of linear magnetic field sensors.[8,9]

Our Si/SiO_2(400 nm)/Ta(5 nm)/Ru(15 nm)/Co(3 nm)/IrMn(15 nm)/ Cu(d_{Cu})/ Co(7 nm)/Ru(3 nm) films were deposited by magnetron sputtering with base pressure of 5.0×10^{-8} mbar, Ar pressure of 1.0×10^{-2} mbar for the IrMn deposition, and 2.5×10^{-3} mbar for the others. Here, thicknesses of the Cu layer d_{Cu}=0.75, 1.0, and 1.2 nm were chosen in an attempt to produce samples with top Co layer weakly coupled to IrMn. This would allow easier modification of the respective EB direction, the value of H_{EB} and even its sign.[5,10]

The magnetic characterization was done at room temperature via alternating gradient-field magnetometer with the measurement magnetic field, H, applied in the plane of the films. After determining the common (to both FM and AF) easy magnetization axis (i.e., the EB direction) induced by the stray field from the magnetron during the deposition, hysteresis loops were traced for different H orientations. Subsequently, the films were subjected to different annealing procedures, varying the following parameters: each sample was heated to the annealing temperature, T_a and kept for a given time interval, t_a in certain magnetic field, H_a applied in different in-plane directions, θ_a (=0° for H_a along the EB direction of the as-deposited samples). None of our samples showed considerable training effects.

The easy-axis hysteresis loops of the as-deposited samples are shown in Fig. 1. The thinner, first-deposited (henceforth, hard magnetic phase, h) Co layers represent 30% of the films' magnetization, show relatively high H_C value of ≈60 Oe, and $H^h_{EB}\approx300$ Oe. A conventional for IrMn/Co annealing procedure, i.e., high H_a of 1.6 kOe, θ_a =0°, T_a=200 °C, and t_a=15 min, led to a 30% decrease of H_C and almost 50% lower H_{EB} for all films, as seen from the corresponding loops, also plotted in Fig. 1 (the loops for the sample with d_{Cu}=1.0 nm, not shown, are very similar to those for 0.75 nm). Whether the IrMn layer is deposited below (i.e., AF/FM, or bottom-pinned FM configuration) or on top of the FM (FM/AF, or top-pinned) layer is rather important in both in-plane and perpendicular EB systems.[11–14] Magnetic field annealing of bottom-pinned structures enhances H_{EB} and normally exceeds that of top-pinned films. However, large decrease of H_{EB} (also obtained here) has been found upon heating for the FM/AF configuration, fre-

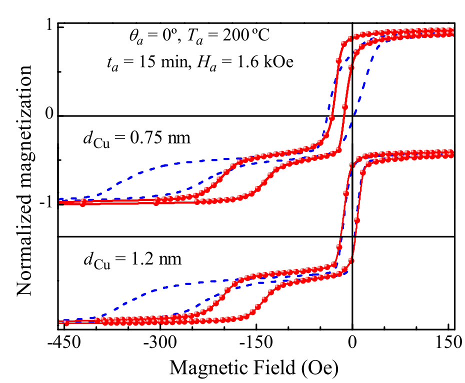

FIG. 1. (Color online) Hysteresis loops of the samples annealed for 15 min at 200 °C with high H_a along the original EB direction; the dashed lines correspond to the as-deposited samples. The lines are only guides to the eyes.

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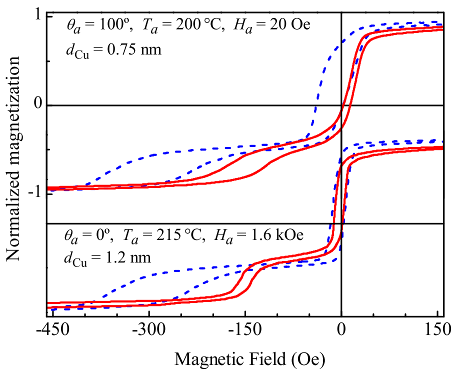

FIG. 2. (Color online) Solid lines: hysteresis loops for the samples with d_{Cu}=0.75 and 1.2 nm annealed for 60 min. The annealing parameters are given in the body of the figure; dashed lines correspond to the as-deposited samples.

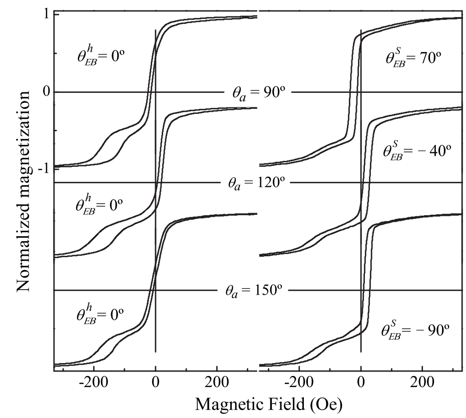

FIG. 3. Variation of the soft-phase shift with the direction of H_a for the sample with d_{Cu}=0.75 nm, annealed for 15 min at 200 °C in H_a=20 Oe. Left: loops traced for measuring field along the EB direction of the hard phase; right: the same for measuring field along the EB direction of the soft phase.

quently attributed to a realignment of the pinned interfacial AF spins.[11,14] Conflicting reports about the role of grain size and texture of IrMn films can be found (e.g., Refs. 12 and 15 and the references therein). Employing numerical simulations as in Ref. 10, we estimated the Co(3 nm)/IrMn exchange coupling constant of the as-deposited samples as $J_E = 0.15$ erg/cm^2, practically coinciding with that of a similar, however inverted, annealed IrMn/Co(5 nm) film.[10] The drop of H_{EB} and subsequently of J_E of the hard phase of top-pinned Co layers after heating could be due to Mn diffusion toward the Co layers; bottom-pinned CoFe/IrMn interfaces have been found to be more stable against such a Mn outdiffusion.[16]

The modifications of the bottom-pinned (soft magnetic phase, *s*) Co layers after this annealing are also in agreement with above-cited literature data. IrMn and Co are practically decoupled[10] for $d_{Cu} > 1.0$ nm, so no change is detected for this phase, as clearly shown in the bottom panel of Fig. 1. The corresponding *s*-part of the hysteresis loops of the annealed films are clearly displaced in the same direction as the *h* parts, as expected.

Annealing the as-deposited sample with d_{Cu}=1.2 nm at slightly higher T_a and the same H_a (Fig. 2) decreased significantly H_C of both phases; H^h_{EB} 15% lower than that of the sample annealed at 200 °C was measured as well. Thermally activated interdiffusion at the IrMn/Co interface during prolonged annealing could be responsible for the apparent difficulty to saturate the soft phase at low measuring fields. The same features regarding the hard phase are also observed when the sample with the thinnest spacer layer was subjected to prolonged annealing with H_a=20 Oe applied at $\theta_a = 100°$, i.e., a field with very small negative component along the original EB direction. The hysteresis loop of the soft phase, however, showed a significant positive field shift.

To study this effect, pieces of the as-deposited sample with d_{Cu}=0.75 nm were annealed for 15 min at the same temperature and field amplitude, where the in-plane direction of H_a was varied. The loops corresponding to the EB direction of the hard and soft phases for three particular orientations are plotted in Fig. 3. Neither treatment changed the EB direction of the hard-phase, i.e., $\theta^h_{EB}=0°$ for all θ_a; the only effect on the magnetization of this phase is a reduction of H_C when θ_a is increased. θ^s_{EB}, however, showed a very strong dependence on the H_a orientation. The soft-phase EB direction gradually rotates with θ_a, being $\theta^s_{EB}=-90°$ for $\theta_a=150°$.

The top panels of Fig. 4 show the hysteresis loops measured along the easy and hard directions for a sample treated in the same conditions as above but with H_a antiparallel to the original EB direction. The subloop of the soft phase is positively displaced and although the hard phase presents the same H^h_{EB} as that obtained after the conventional treatment (Fig. 1), now H_C is twice lower. This shows that low-field annealing with $\theta_a=180°$ is not the best treatment if one is looking for oppositely displaced and well-defined subloops. Note also that the magnetization of the 3 nm thick Co layer, when measured along the hard axis (Fig. 4, top right), is split into two identical and oppositely shifted subloops. This means that, as a result of the annealing, approximately half of the uncompensated AF spins at the interface (responsible for the EB of the as-deposited sample) are reversed during the annealing and the rest remained along the deposition-field direction.

Sizable reduction of H_{EB} and even change of its sign have been obtained after prolonged reverse-field annealing[17] in temperatures lower than 200 °C, attributed to reversal of AF spins due to thermally induced frustration of the coupling at the FM/AF interface, whatever the influence of the H_a on

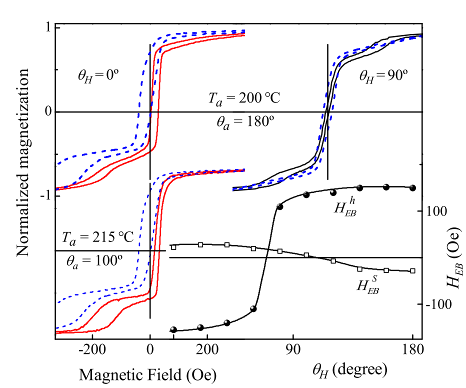

FIG. 4. (Color online) Magnetization curves for the sample with d_{Cu} =0.75 nm, annealed for 15 min in H_a=20 Oe; the other parameters are given in the figure. Left: the EB-direction loops. Top right: the corresponding hard-axis curve. The dashed lines correspond to the as-deposited samples. $H_{EB}(\theta_H)$ of both phases for the sample annealed at 215 °C and θ_a=100° are plotted in the bottom right corner.

the AF; theoretical interpretation of this effect can be found, e.g., in Refs. 18 and 19 and the references therein. The reorientation of AF interface spins only at the side of the incomplete Cu spacer layer during our short-duration thermal treatment strongly indicates that, in our case, if the effect is caused by exchange-coupling frustration, the latter comes from the incomplete separation of the IrMn and the 7 nm thick Co layer. At T_a, even a field of only 20 Oe is sufficient to completely orient the soft FM layer moments which, on their turn, align the interface AF spins, thus reorienting the EB direction. This field is not able to saturate the thinner Co layer, so no change is observed in its magnetic behavior. However, we verified that for $H_a \geq 50$ Oe, both *h* and *s* EB directions are reoriented along the h_a direction through annealing.

Thus, performing low-field off-aligned annealing at 200 °C we succeeded to reorient the EB direction of *only one* of the two Co layers. Still, as seen in Figs. 2 and 3, when the two subloops coexist, at least one of these is not very well defined. A better result was obtained for $T_a = 215$ °C and $\theta_a = 100°$. The hysteresis loop for measuring field along the original EB direction is shown in the bottom left corner of Fig. 4. Both hard and soft subloops have characteristics very similar to those of the conventionally annealed sample (Fig. 1), the only difference being the oppositely shifted soft-phase subloop. As seen from $H_{EB}(\theta_H)$ of both Co layers plotted in Fig. 4, these shifts are, practically, in antiphase.

In summary, we demonstrated that appropriate low-field off-aligned annealing of Co/IrMn/Cu/Co films can transform a double negatively/negatively shifted loop to a negatively/positively shifted one. Certainly, more systematic research needs to be conducted to further clarify the role of each of the annealing parameters on this rearrangement, which will be the aim of a forthcoming study.

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