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# Magnetization reversal in composition-controlled $Gd_{1-x}Co_x$ ferrimagnetic films close to compensation composition

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We report on a model system for micromagnetic studies, i.e., ferrimagnetic $Gd_{1-x}Co_x$ thin films with controlled composition gradient and, therefore, a controlled magnetization gradient along the film. By employing extraordinary Hall effect measurements and Kerr microscopy, we have studied magnetization reversal and shown that, around compensation, varying magnetization with temperature or composition is equivalent. In particular, the coercive field diverges close to the compensation temperature or close to the compensation interface. The position of the compensation interface is very sensitive to temperature and can be used as a probe of sample heating. © *2011 American Institute of Physics*. [doi:10.1063/1.3609860]

The composition of an intermetallic ferrimagnetic film can be chosen so that its net saturation magnetization $M_s$ vanishes at the so-called compensation temperature ($T_{comp}$). In the vicinity of $T_{comp}$, $M_s$ increases linearly as a function of temperature. Alternatively, a magnetization gradient at a fixed temperature can be obtained by fabricating films with controlled composition gradient around compensation.[1]

Both systems have the important property that magnetization can be changed continuously, without substantially varying the other magnetic properties such as anisotropy and sublattice magnetization. Intermetallic films at compensation have been recently exploited in data storage media[2] and magnetic tunnel junctions[3] and to demonstrate the feasibility of sub-picosecond magnetization reversal.[4] Compensated ferrimagnets have been also proposed recently as interesting candidates for spin torque induced domain wall (DW) motion applications,[5] as it is expected that spin torque efficiency should be enhanced in the vicinity of the compensation composition.

In this work, we report on the magnetization reversal properties of ferrimagnetic $Gd_{1-x}Co_x$ films with in-plane composition gradient, around the compensation composition $x_{comp}$. Our results show that the reversal mechanism close to $x_{comp}$ is controlled by the amplitude of the magnetization and not by thermal excitations, unlike usual ferromagnets where coercivity increases as magnetization increases at low temperature.

40 nm thick $Gd_{1-x}Co_x$ films, with 3 nm thick Ti buffer and capping layers, were deposited onto Si(001) substrates using dc magnetron sputtering in the facing target geometry. As widely reported in the literature (Refs. 6–8), perpendicular magnetic anisotropy is induced by the growth process.

A composition gradient in the deposited magnetic layer can be induced by placing the sample away from a high symmetry position and using non symmetrical targets. The films include a composition $x = 0.8$ for which $T_{comp}$ is close to room temperature.[9] UV lithography and lift-off technique were consequently used to pattern the film into 100 $\mu$m-wide wires parallel to the composition gradient direction.

The first aim of this work is to locate the plane of the film having the composition $x_{comp}$ (compensation interface) for which the total magnetization vanishes at RT. This is a very interesting and unusual micromagnetic object. Let us look at the evolution of the Co and Gd magnetization along the film. As sketched in Fig. 1(a), the magnetization increases linearly with $x$ on either side of $x_{comp}$. $M_s$ is in the direction of the Gd moments below $x_{comp}$ while it changes sign and becomes parallel to the Co moments when $x > x_{comp}$. Macroscopically, the compensation interface is a DW (with

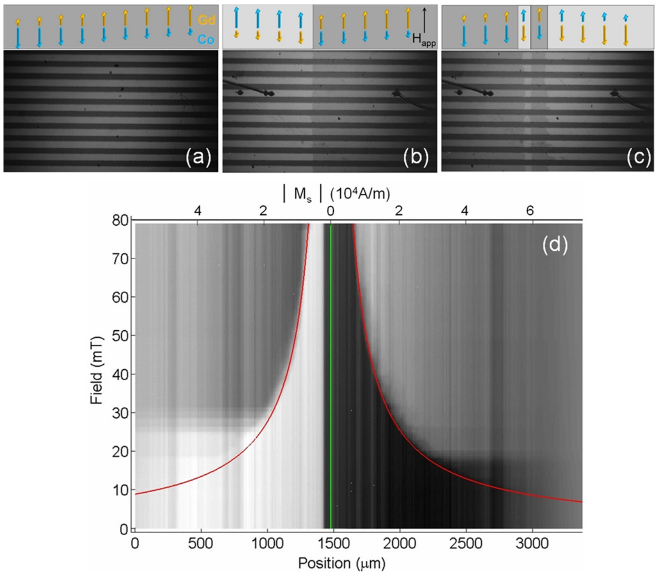

FIG. 1. (Color online) Kerr microscopy images of the GdCo wires and sketch of the corresponding Gd and Co magnetic moments. (a) image of the as-deposited state; (b) image taken after application of (500 mT) magnetic field oriented as indicated on the right side of the image: the compensation interface becomes visible; (c) starting from (b), a field of −80 mT is applied in the opposite direction to reverse the magnetization. The field of view is $3.3 \times 1.9$ mm$^2$; (d) Three dimensional map of magnetic reversal process. The vertical green line depicts the position of the compensation interface and the red curves correspond to $E_p/M_s$ fit with $E_p = 680$ J/m$^3$ (enhanced online) [URL: http://dx.doi.org/10.1063/1.3609860].

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vanishing magnetization and no exchange energy cost) as it separates two film regions with opposite magnetization. We can expect that such an interface could be visible and would give rise to opposite contrasts if a magnetic imaging technique sensitive to the total magnetization was used.

Microscopically, however, the magnetization of both Co and Gd sublattices changes continuously across $x_{\mathrm{comp}}$, and no discontinuity of the magnetic contrast is expected using a technique sensitive to one of the two magnetic sublattices. This explains the image of the as-deposited film measured at room temperature by magneto-optical Kerr microscopy, shown in Fig. 1(a).

In the visible range, Kerr microscopy is a mirror of the Co magnetization, since the Kerr rotation is larger for Co than for Gd magnetic sublattice.[10] Apart from some inhomogeneities due to an uneven illumination, the contrast in Fig. 1(a) is constant in the probed region. This proves that the sample is not demagnetized, as it is found very often in ferromagnetic films with perpendicular magnetization. This comes from the fact that the demagnetizing field is vanishing close to the $x_{\mathrm{comp}}$. As expected, no abrupt change in contrast is found along the wire as the composition changes since the direction of Co magnetization is the same all along the wire.

In order to locate the position of the compensation interface with Kerr microscopy, we have applied a large magnetic field in the easy axis direction, perpendicular to the film plane. In the region of the sample where $M_s$ is initially antiparallel to $H_{\mathrm{app}}$, the magnetization reverses and aligns with the field. The direction of the Co magnetization is then opposite on either side of the compensation composition and a contrast appears in the Kerr microscopy images, as shown in Fig. 1(b). Note that an infinite field would have to be applied to visualize the exact location of the compensation interface; the interface visualized by applying $\mu_0 H_{\mathrm{app}} = 500$ mT is 70 $\mu$m away from this interface.

In these conditions, the macroscopic magnetization does not change sign across the film, but microscopically an ideal, chargeless, Bloch DW is present in the Co and Gd sublattices at the compensation composition location.

In order to obtain quantitative information on the composition gradient along the wires, hysteresis loops were measured by extraordinary Hall effect (EHE) as a function of temperature on a Hall cross patterned 1.9 mm away from the compensation interface. Measurements were carried out in an area of $100 \times 100$ $\mu$m$^2$, between 50 K and 300 K in fields up to 6 T, using dc currents. The Hall resistance $R_{\mathrm{H}}$ was determined using $V = R_H I/t$, where $t$ is the film thickness. The EHE loops as a function of applied field are square, allowing easily to determine $H_c$. The measurements presented in Fig. 2 show that the coercive field increases and diverges when approaching a temperature $T = 218$ K.

Divergence of the coercive fields when approaching compensation temperature is well established and is due to the constant Zeeman energy necessary to overcome switching energy barriers as $M_s$ tends to zero. We then deduce that $T = 218$ K is the local compensation temperature.

We can now use the quantitative information obtained so far, to extract the composition gradient in our films. From Kerr microscopy and EHE, we have determined that the $T_{\mathrm{comp}}$ changes by (300-218) K = 72 K over a 1.9 mm distance along the film. The compensation temperature gradient is 45 K/mm. From the mean, field model follows that, around the compensation composition, a 1% change of composition induces a 44 K shift of $T_{\mathrm{comp}}$.[9] The composition gradient is then of the order of 1%/mm and the magnetization gradient of 4.10$^4$ A/m/mm (3.2 mT/mm).

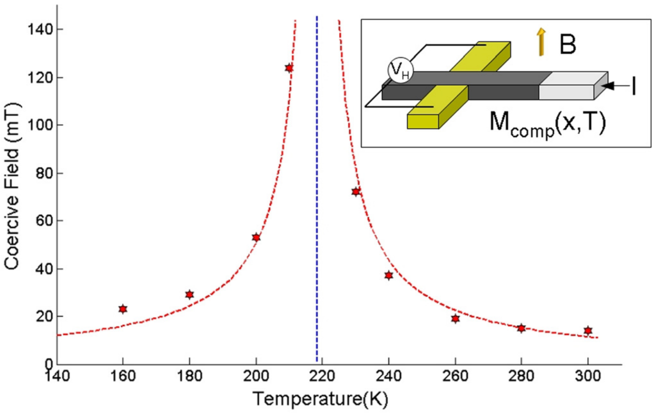

FIG. 2. (Color online) Coercive field as a function of temperature measured on a Hall cross located 1.9 mm away from the compensation interface $x_{\mathrm{comp}}$ at 300 K. The red dashed curves correspond to the $1/M_s$ fit.

In the same way, as the coercive field diverges in a homogeneous system at the compensation temperature, we expect that at 300 K, the coercive field will diverge along the film, when approaching the compensation composition $x_{\mathrm{comp}}$.

This has been proved using Kerr microscopy measurements at room temperature. Starting from the magnetic configuration depicted in Fig. 1(b) (where a DW has been created in the sublattices very close to the $x_{\mathrm{comp}}$ by applying a 500 mT field), an opposite field of varying strength is applied to reverse the magnetization. An example of Kerr image, obtained with a field of $-60$ mT, is shown in Fig. 1(c).

The magnetization reversal is governed by nucleation of reversed domains and their propagation along the wires. In these films with composition (i.e., magnetization) gradient, for a fixed applied field, the DWs stop when the Zeeman energy associated to the field is no longer sufficient to overcome the local propagation barrier. The magnetization reversal is initiated at the far edges of the wires, outside the field of view presented in Fig. 1. Note that in all the wires, the nucleation field is systematically smaller on the Co-rich part of the wires ($\mu_0 H_N \sim 20$ mT) than on the Gd-rich part ($\mu_0 H_N \sim 27$ mT). This asymmetry can be explained by the fact that the magnetization is lower at the sample edge on the Gd rich side as the edge is closer to the compensation interface.

The two DWs propagate along the wires and stop on either side of $x_{\mathrm{comp}}$, at symmetric positions that depend on the applied field amplitude. The value of the applied field is a measure of the local coercive field at the position (i.e., for the composition) where the DWs stop.

Magnetic fields ranging from zero up to $-80$ mT were applied and the sequence of corresponding images was recorded with 0.5 mT steps. The images were analyzed using the methods described in Ref. 9 and the outcoming results are summarized in Fig. 1(d).

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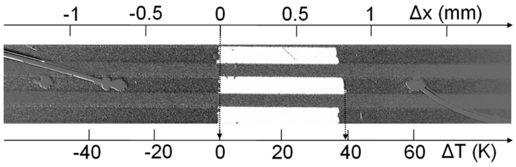

FIG. 3. Differential Kerr microscopy image showing the displacement of the compensation interface position due to the Joule heating created by a dc current of 34 mA flowing in the central GdCo wire. Using our previous results, the temperature increase with respect to RT can be estimated as $\Delta T = 39$ K.

The magnetization is unreversed over a zone that gets narrower as the applied field increases. This is the consequence of the expected divergence of the local coercive field as the composition approaches the compensation composition. In the most general model,[11] the coercive field is related to $M_s$ by the expression $\mu_0 H_c = E_p/M_s$, where $E_p$ is the propagation energy barrier and $M_s$, the local magnetization.

Similarly to the $H_c(T)$ curve obtained with EHE effect, the $H_c(x)$ curve can be indeed fitted using the same expression for the coercive field (and $E_p = 680$ mJ/m$^3$ at RT). This demonstrates that the reversal mechanism close to $x_{comp}$ is controlled by $M_s$ (Zeeman energy) and not by $T$ (thermal excitations), unlike usual ferromagnets where coercivity increases as magnetization increases at low temperature.

Finally, we would like to show that the position of the compensation surface can be moved along the film by changing its temperature. This was obtained by connecting the central wire (visible in Fig. 1) to a current source delivering a dc current of 34 mA.

The difference between the initial magnetic configuration at RT and the magnetic configuration obtained in the presence of the dc current (and of a field of 500 mT) is shown in Fig. 3. The compensation surface was displaced by 0.87 mm, which according to our results corresponds to temperature change of 39 K. Due to the important heat dissipation into the Si substrate, the wires close to the central wire considerably heat up. Note that the position of the compensation interface can be used as a sensitive thermometer: a 500 nm displacement corresponding to the spatial resolution of Kerr microscopy corresponds to a change in temperature of 20 mK. In our case however, the limiting factor defining DW position is the pinning centers distribution, which gives rise to DW roughening with average period of 5 $\mu$m corresponding to 200 mK.

In conclusion, we have shown that $Gd_{1-x}Co_x$ with composition gradient around compensation is a very unusual and interesting micromagnetic system. We have discussed in particular, the presence of an ideal uncharged Bloch DW in the Co and Gd sublattices in a system with continuous macroscopic magnetization, whose position can be visualized with Kerr microscopy after application of a strong magnetic field. We have also shown that, as expected, the coercive field diverges as $1/M_s$ as the compensation interface is approached, in the same way as the $H_c$ diverges close to $T_{comp}$. This proves that the propagation barriers are homogeneous all over the sample. We have shown that the compensation interface can be continuously displaced by heating the sample by Joule effect.

The model system described in this work can be interesting for spin torque induced DW motion studies. This has been recently proposed by Komine *et al.*,[5] who suggest that spin torque efficiency should be enhanced in the vicinity of the compensation composition. Up to now however, no convincing experimental proof of such an efficiency has been reported in the literature. Our results suggest that this may be partly due to sample heating during the application of current pulses, which may give rise to an important change of local magnetization when the proper current densities are used. If the sample composition was optimized so that compensation is obtained at room temperature, the departure from vanishing magnetization conditions during the application of current pulses may explain the failure to evidence spin torque effect in these systems. Our work suggests that composition should be optimized taking into account thermally induced magnetization variations.

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