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NiFe/Au/Co/Au layered films Magnetic and transport properties

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Magnetic and magnetoresistive properties of $[Ni_{80}Fe_{20}(t_{NiFe})/Au(t_{Au})/Co(t_{Co})/Au(t_{Au})]_N$ multilayers characterized by the in-plane magnetic anisotropy of NiFe layers and the perpendicular anisotropy of Co layers have been described for a range of sublayer thicknesses and various numbers of repetitions N. It has been shown that the magnetic stray fields originating from the domain structure of Co layers, and calculated from the theory of Draaisma can qualitatively explain the measured $R(H)$ dependence in the field range corresponding to the magnetization reversal of Co layers. The investigated MLs are resistant to low temperature annealings. It has also been shown that the magnetic properties of MLs can be additionally tailored by inserting thin Co layers on the NiFe/Au interfaces modifying the effective anisotropy of NiFe layers.

Key words: *giant magnetoresistance; magnetic domains; perpendicular anisotropy*

1. Introduction

The giant magnetoresistance (GMR) effect, i.e., the dependence of resistance of a magnetic system on the relative angle between magnetic moments of neighbouring layers or grains has been intensively investigated for almost two decades [1]. In our previous works [2–4] we have shown that GMR in NiFe/Au/Co/Au systems in which magnetic anisotropy alternates between in-plane and perpendicular orientation are potentially interesting from the application point of view as they exhibit a linear dependence of resistance (R) on the perpendicularly applied magnetic field (H). The perpendicular anisotropy of Co layers results from the influence of Au/Co interfaces. In this paper, we demonstrate that the most important features of the observed $R(H)$ dependences can be explained within the theory of Draaisma and de Jonge [5] although this theory describes multilayers (MLs) with purely perpendicular anisotropy.

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We also analyze how the modifications of the structure change the magnetoresistance characteristics of MLs.

2. Experimental

The $[Ni_{80}Fe_{20}(t_{NiFe})/Au(t_{Au})/Co(t_{Co})/Au(t_{Au})]_N$ MLs with t_{NiFe} = 2 nm, t_{Au} = 1.5–3 nm, t_{Co} = 0.6–1.2 nm and N = 1–15 were deposited in Ar atmosphere using UHV magnetron sputtering. The films were deposited directly onto Si(100) substrates with native oxide. The sputtering rates were 0.06, 0.05, and 0.045 nm·s^{-1}, for Au, NiFe and Co, respectively. The microstructure was investigated using X-ray diffraction with CuK_α radiation. The magnetization reversal processes were studied at ambient temperature with a vibrating sample magnetometer (VSM). Current in-plane magnetoresistance (MR) was measured in a four-point configuration. Magnetic fields up to 1600 kA·m^{-1} (2 T), applied in-plane and perpendicularly, were used. The $MR(H)$ dependence was calculated relative to the resistance at 1600 kA·m^{-1} and the maximum value determined from $MR(H)$ is called magnetoresistance (MR) amplitude. Some samples were cumulatively annealed in gas flow thermostat in dry N_2 at temperatures t_{anneal} up to 263 °C. Each annealing lasted 1 h.

3. Results and discussion

Exemplary resistance dependences of our MLs are shown in Fig. 1. The $R(H)$ characteristics depend markedly on the direction of the applied magnetic field relative to the surface of the sample. In the following, we focus on the case of the perpendicularly applied field (for the in-plane orientation, see Ref. [2] and references therein). The overall character of the $R(H)$ curve was explained previously [3]. The quasi linear part ranging from about 125 to 480 kA·m^{-1} corresponds to the reversal of NiFe layers being magnetized in hard direction with the saturation field resulting from the shape anisotropy ($H_S^{NiFe} = M_S^{NiFe} \approx 480$ kA·m^{-1}). The hysteretic part within ±125 kA·m^{-1} is the result of a simultaneous reversal of Co layers along their easy direction and NiFe layers along hard direction. The Co layers are characterized by perpendicular anisotropy as evidenced by the $M(H)$ dependence (Fig. 2) which is typical of systems with stripe domain structure [6]. It should be pointed that there is a very close correlation between $M(H)$ and $R(H)$ dependences as seen in Fig. 2. The characteristic fields corresponding to the nucleation of the stripe domain structure in Co layers (H_N^{Co}) and its annihilation (H_S^{Co}) as well as H_S^{NiFe} (see Fig. 1) are well visible in both curves (Fig. 2). We have shown [2] that the local minima of resistance observed in $R(H)$ dependence result from the dipolar coupling between Co and NiFe layers by means of magnetostatic fields originating from the stripe domain structure of Co layers.

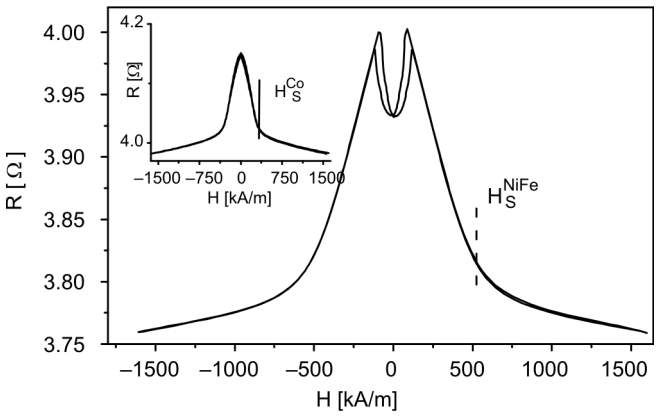

Fig. 1. The magnetic field dependence of the resistance of [NiFe(2 nm)/Au(1.5 nm)/Co(0.8 nm)/Au(1.5 nm)]$_{10}$ ML for the field applied perpendicularly to the sample plane (MR = 6.5%) and in-plane (4 %) (inset)

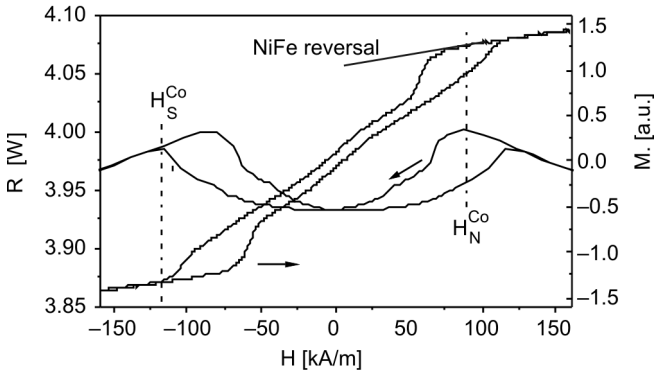

Fig. 2. The $R(H)$ dependence for [NiFe(2 nm)/Au(1.5 nm)/Co(0.8)/Au(1.5 nm)]$_{10}$ ML for the field applied perpendicularly to the sample plane and the corresponding $M(H)$ dependence. Shown are nucleation (H_N^{Co}) and annihilation (H_S^{Co}) fields of the Co domain structure

In the following, we demonstrate that the character of that dependence can be qualitatively described based on the theory of Draaisma and de Jonge [5] although, what should be strongly emphasized, it describes MLs with only perpendicular anisotropy of sublayers. To compute the average dipolar field acting on NiFe layers in function of the applied field we calculated the complement Co domain widths (Eqs. (5) and (6) from Ref. [5]). We assumed that the characteristic length τ, describing the domain wall energy, is 2 nm (as compared to 0.8 nm obtained by Draaisma for Co/Pd MLs) to achieve the widths corresponding to the experiment (approxiumately 150 nm in zero field [7]). The length τ corresponds to a domain wall energy of 2.5×10^{-3} J·m^{-2}. This value is ca. four times lower than the estimated one based on the theory of Schlömann [8] with the effective anisotropy constant $K = 1.53\times10^{6}$ J·m^{-3} corresponding to $t_{Co} =$ 0.8 nm [2]. Therefore, we treat τ as a fitting parameter which is only indirectly related

to the domain wall energy in our system. We ignore the component of magnetostatic field parallel to the sample surface as it is important only in the direct vicinity of domain walls and is negligible for $N = 10$. The average absolute value of field H_d calculated for the set of parameters of [NiFe(2 nm)/Au(1.5 nm)/Co(0.8 nm)/Au(1.5 nm)]$_{10}$ ML, assuming zero thickness of domain walls and the corresponding domain widths, are shown in Fig. 3. (To avoid the edge effects, the calculation was performed for 50 domains in a layer but the average was computed for 260 inner domains, i.e., 26 in each layer [9]).

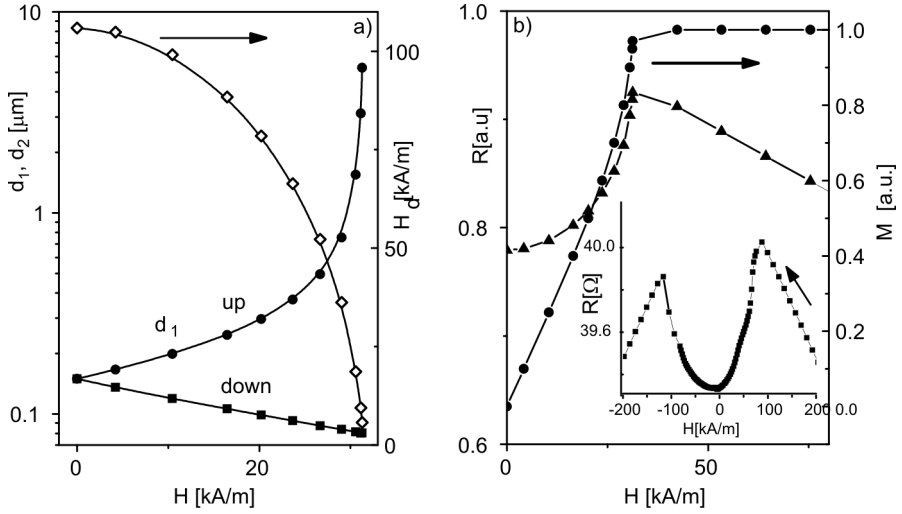

Fig. 3. The low field dependence (a) of the widths of up (•) and down (■) domains calculated from the theory of Draaisma and the corresponding average stray fields (at the midlines of NiFe layers) originating from magnetic domains (◇) and the calculated field dependence (b) of the magnetization of Co layers (•) and the appropriate $R(H)$ (▲) dependence calculated for [NiFe(2 nm)/Au(1.5 nm)/Co(0.8)/Au(1.5 nm)]$_{10}$ ML. The inset shows the $R(H)$ dependence measured in higher fields (■)

It is clearly visible that the magnetostatic fields acting on NiFe layers are strongly field dependent, approaching zero for $H \approx H_N^{Co}$ (the model of Draaisma does not describe hysteresis so we show data for positive fields only; here $H_N^{Co} = H_S^{Co}$). The field dependence of resistance can be approximated by the following equation [9]:

$$R(H) = a(r_0 - dr\cos(\phi_\uparrow)) + (1-a)(r_0 + dr\cos(\phi_\downarrow)) \tag{1}$$

where a denotes the fraction of Co layers magnetized in parallel to the initial direction of perpendicular magnetic field ($\hat{n}$) ($a(H) = (M^{Co}(H)/ M_S^{Co} + 1)/2$), ($r_0 - dr$) and ($r_0 + dr$) denote resistance in saturation and maximum attainable resistance, respectively; $\phi_\uparrow$ and $\phi_\downarrow$ designate the angle between the $\hat{n}$ direction and the magnetic moment direction of NiFe layer regions adjoining Co layers areas with up and down magnetization, correspondingly [9]. In view of the fact that, up to the saturation,

$\cos(\phi_{\uparrow})$ and $\cos(\phi_{\downarrow})$ depend linearly on the effective field ($\cos(\phi_{\downarrow}) = (H+H_d)/M_S^{NiFe}$), Eq. (1) can be rewritten as:

$$R(H) = r_0 - \frac{dr}{M_S^{NiFe}}[H\frac{M^{Co}(H)}{M_S^{Co}} + H_d(H)] \tag{2}$$

The calculated $R(H)$ dependence is shown in Fig. 3 (b) together with the corresponding measured one (inset). It is clearly visible that the overall character of the changes is relatively well described. The maximum resistance change attributed to the influence of domain fields ($H_d dr/M_S^{NiFe}$) is about $0.22dr$ in the model dependence and $0.44dr$ in the experimental one. The calculated and measured saturation fields of Co layers differ by some 60%. This is caused by the presence of stray fields originating from NiFe layers which were not present in the system analyzed by Draaisma and de Jonge. To adequately describe the $R(H)$ dependence one must utilize a dedicated theory which at present is not available in the literature. Nevertheless, the presented model confirms that the measured resistance characteristics of NiFe/Au/Co MLs have their origin in the dipolar fields of Co layers.

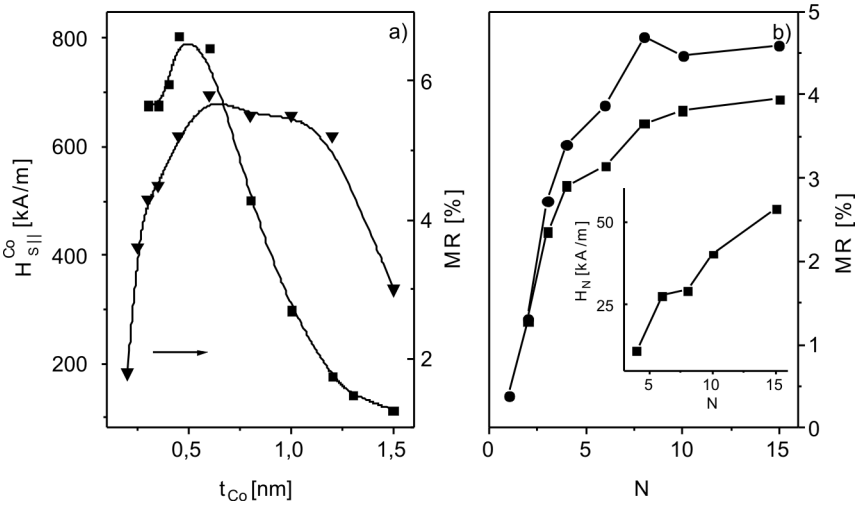

Fig. 4. MR amplitude (triangles) measured (a) with the magnetic field applied perpendicularly and the saturation field of Co layers measured with in-plane field determined for [NiFe(2 nm)/Au(3 nm)/Co(t_{Co})/Au(3 nm)]$_{15}$ MLs with various t_{Co} and MR amplitude (b) measured with magnetic field applied parallel (squares) and perpendicular (dots) for [NiFe(2 nm)/Au(3 nm)/Co(0.8 nm)/Au(3 nm)]$_N$ MLs with various numbers of repetitions N. The inset shows the corresponding nucleation fields

Depending on the structure and magnetic properties of cobalt layers, distinct ranges in the Co layer thickness dependence can be distinguished [10]. For $t_{Co} \leq 0.3$ nm, sublayers are superparamagnetic and the magnetoresistance dependence shows no saturation in available fields. The $MR(t_{Co})$ dependence exhibits a weak maximum at $t_{Co} = 0.6$ nm (Fig. 4a). Its exact position is the result of the interplay be-

tween the current shunting through the Co layers and the decreasing role of the more effective interface spin dependent scattering as compared to the scattering within the Co layers. Alternatively, the position of the maximum may correspond to the transition from a discontinuous to a continuous Co layer. The strong decrease of magnetoresistance amplitude for $t_{Co} \geq 1.2$ nm is caused by the transition to the in-plane anisotropy: both kinds of magnetic layers reverse in hard direction and the maximum attainable angle between them reduces. The in-plane saturation field H_{SII}^{Co} of the Co layers (Fig. 4a), which is a measure of the effective anisotropy, attains a maximum at $t_{Co} \approx 0.5$ nm and then decreases reflecting the decreasing role of Au/Co interfaces.

The MR amplitude increases with N (Fig. 4b) due to the diminishing role of scattering on outer interfaces and because for higher N the electron mean free path λ encompasses more magnetic/nonmagnetic interfaces ([2] and references therein). In an ideal system, when the total thickness of the multilayer exceeds λ, the MR amplitude asymptotically saturates. In our MLs, the saturation takes place at a higher total thickness because of a higher density of structural imperfections in regions close to the substrate. For the application of investigated multilayers as sensors characterized by a linear $R(H)$ dependence in a large field range, the low value of H_N is desired. As can be seen in Fig. 4b, this condition is fulfilled for multilayers with t_{Co} and t_{Au} for which high GMR values are recorded as well. H_N progressively increases with N (see the inset of Fig. 4b). However, structures with large N (i.e., corresponding to the saturation of the $MR(N)$ dependence) have quite low nucleation fields what is favourable for the applications.

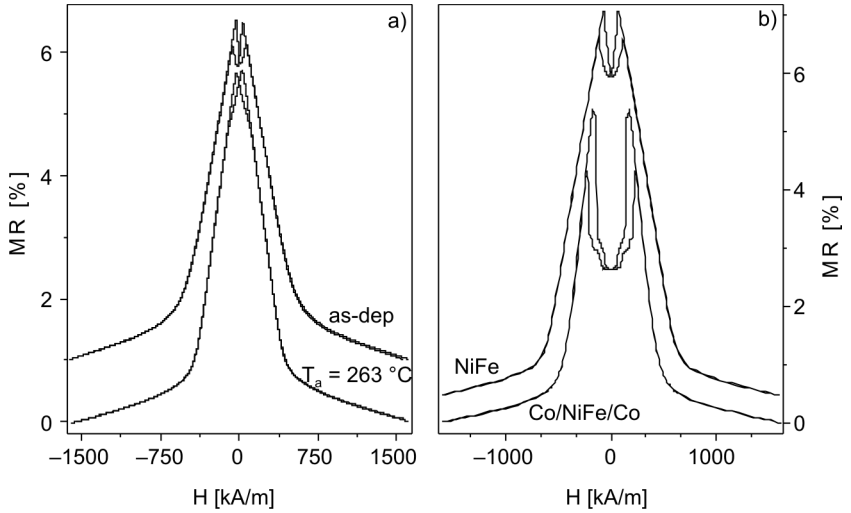

Fig. 5. The $R(H)$ dependences (a) for [NiFe(2 nm)/Au(2.2 nm)/Co(0.6 nm)/Au(2.2 nm)]$_{10}$ ML in the as-deposited state and after the series of cumulative annealings at temperatures up to 263 °C for the field applied perpendicular and the $R(H)$ dependences (b) for [Co(0.4 nm)/NiFe(2.4 nm)/Co(0.4 nm)/Au(2.2 nm) /Co(0.8 nm)/Au(2.2 nm)]$_{10}$ and [NiFe(3.2 nm)/Au(2.2 nm)/Co(0.8 nm)/Au(2.2 nm)]$_{10}$ MLs. The upper curves in both panels are vertically shifted for clarity

Thermal stability is on of the important factors from the application point of view. Figure 5a shows that the MLs are quite resistant to annealing at t_{anneal} up to 263 °C. The saturation field of $R(H)$ increases slightly but the magnetoresistance amplitude is preserved (detailed investigations of the thermal stability are shown in Ref. [11]).

The noticeable dependence of $R(H)$ curves on the properties of NiFe layers, namely their shape anisotropy, allows modifications of magnetoresistive characteristics by using composite layers instead of NiFe layers [3]. Figure 5b shows that replacing NiFe by Co/NiFe/Co leads to the diminishing of the saturation field of $R(H)$ dependence which is desirable for magnetic field sensing applications. One observes simultaneously that the change of resistance in a hysteretic range is much more pronounced. These changes in $R(H)$ can be explained remembering that thin Co layers adjoining to Au layers possess a perpendicular magnetic anisotropy [3]. Consequently, the composite layer, being strongly coupled by the direct exchange at NiFe/Co interfaces, behaves like a single magnetic layer with diminished in-plane anisotropy. Accordingly, the $M_{\text{S}}^{\text{NiFe}}$ in Eq. (2) should be replaced by $H_{\text{S}}^{\text{CoNiFeCo}}$ which is lower in value; this leads to larger changes of resistance under the influence of H_d.

4. Conclusions

The magnetostatic fields of Co layers computed within the theory of Draaisma and de Jonge qualitatively explain the behaviour of NiFe layers and the $R(H)$ characteristics. We showed that the use of the composite layers with in-plane anisotropy allow the modification of saturation fields of $R(H)$ dependence which is necessary from the application point of view.

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