Tailoring Size Effects on the Exchange Bias in Ferromagnetic-Antiferromagnetic <100 nm Nanostructures

V. Baltz, J. Sort, S. Landis, B. Rodmacq, and B. Dieny

1SPINTEC (URA 2512), CEA/CNRS, 17 Av. Martyrs, 38054 Grenoble Cedex 9, France
2LETI/D2NT, CEA, 17 Av. Martyrs, 38054 Grenoble Cedex 9, France

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The hysteresis loop shift in sub-100 nm ferromagnetic-antiferromagnetic (AFM) nanostructures can be either enhanced or reduced with respect to continuous films with the same composition, with varying the AFM layer thickness. An enhancement of the coercivity and a reduction of the blocking temperature are also observed. These effects are mainly ascribed to the physical limitations that the dot sizes impose on the AFM domain size and the concomitant weakening of the pinning strength exerted by the AFM during magnetization reversal of the FM.

Exchange bias (EB) refers to the shift of the hysteresis loop along the magnetic field axis observed in exchange interacting ferromagnetic-(FM-)antiferromagnetic (AFM) materials. The loop shift is usually accompanied with an enhancement of coercivity [1]. The majority of theoretical models dealing with EB attribute such effects to the formation of domains and pinning of domain walls (DWs) either in the FM [2,3] or in the AFM layer [4,5].

During recent decades, EB properties have been extensively investigated, mainly in thin films, due to their technological applications in magnetic random access memories and magnetoresistive read heads based on spin valves or tunnel junctions [6]. Recently, the drastic increase in the areal density of magnetic recording media has motivated the study of EB properties in systems of reduced lateral dimensions [7–19]. The reduction of the lateral dimensions of an EB system down to length scales comparable to FM or AFM magnetic domain sizes (typically hundreds of nanometers) is also interesting from a fundamental point of view since this results in a confinement and subsequent alteration of the FM and AFM domain structures [10–13], hence allowing us to probe the role of domains on EB.

Although a considerable number of studies have been reported in the literature on micron or submicron spin valve systems [20], the effects of reduced lateral dimensions on EB have been far less investigated. Yet when the dimensions of the FM-AFM nanostructures are reduced down to the magnetic domain sizes, an enhancement of the coercive field, \( H_C \) [8,9,11–14], or changes in the asymmetry of the hysteresis loops have been observed [9,17]. More controversial is the effect of reduced lateral dimensions on the magnitude of the EB field, \( H_E \). Indeed, some authors reported that \( H_E \) is enhanced in nanostructures [7–10], whereas others observed the opposite trend [11–18]. Up to now, this discrepancy has been attributed to the different materials or the nanostructuring techniques employed.

In this Letter, we report on the dependence of EB on the AFM and FM layers thicknesses in sub-100 nm FM-AFM bilayers sputtered on prepatterned Si square dots. We demonstrate that, at room temperature, although all samples were fabricated using the same lithography technique and all of them consist of the same FM and AFM materials, it is possible to either enhance or reduce the magnitude of \( H_E \) in the nanostructures, with respect to continuous films with the same composition, by varying the thickness of the AFM layer, while keeping the dot size constant. Such a behavior is not observed when varying the FM thickness.

To fabricate the nanostructures, Si wafers were first patterned by electron beam lithography and reactive ion etching to form arrays of Si square dots with lateral sizes of 90 nm, height of 300 nm, and periodicity of 200 nm [see inset of Fig. 1(a)]. Two series of multilayers with compositions Ta(5 nm)/Py(12 nm)/IrMn(12 nm)/Pt(2 nm) and Ta(5 nm)/Py(tPy)/IrMn(5 nm)/Pt(2 nm) (where Py is permalloy: Ni81Fe19 and IrMn stands for Ir20Mn80) were deposited by sputtering on the patterned Si wafers. The thicknesses of the ferromagnetic layers were measured using a spectroscopic ellipsometer.

The hysteresis loops of the continuous films (- - -) and the nanostructures (■■■), with compositions Ta(5 nm)/Py(12 nm)/IrMn(12 nm)/Pt(2 nm) (for \( t_{IrMn} = 5, 13, 16, \) and 19 nm), measured at room temperature by longitudinal Kerr effect along the field cooling direction, after cooling from \( T = 550 \) K in the presence of a 2.4 kOe field. The inset is a scanning electron microscopy (SEM) image of the nanostructures.
samples under a negative field from temperatures ranging the continuous film were evaluated by field cooling the effect. The blocking temperatures in both the dots and netic signal from the trenches when measured by Kerr effect setup. The geometry was selected to avoid the mag-

along the field cooling direction using a longitudinal Kerr and the nanostructures were measured at room temperature of all systems) under a 2.4 kOe in plane temperature. The loop shearing is ascribed to the switching field distri-

bution for the different dots [22]. The loop is observed for large $t_{\text{IrMn}}$ values. The decrease of $H_E$ with increasing the AFM layer thickness ($t_{\text{AFM}}$) in the continuous films has already been observed and tentatively explained [5,23–29]. Following the arguments initially proposed by Imry and Ma [24], and using the random field model for FM-AFM exchange biased systems, Malozemoff predicted an inversely proportional relation-

ship between $H_E$ and the AFM layer thickness [5]. His argument is based on the assumption that due to the random coupling through the FM-AFM interface, the AFM spin lattice breaks up into domains, the size of which is determined by a balance between a gain in FM-AFM interfacial energy provided by aligning the local net AFM moment with the FM magnetization and an energy cost due to domain walls (DWs) formation in the AFM. The interfacial coupling energy per unit area is $\sigma_{\text{FM-AFM}} = -J_{\text{FM-AFM}}/(aD_{\text{AFM}})$, where $J_{\text{FM-AFM}}$ is the interatomic exchange constant of the FM-AFM coupling, $a$ the distance between AFM spins, and $D_{\text{AFM}}$ the AFM domain size. On the other hand, the DW energy per unit area of DW in the AFM can be written as $\sigma_{\text{DW,AFM}} = \pi^2 J_{\text{AFM}}/(4aD_{\text{AFM}})$, where $J_{\text{AFM}}$ is the exchange constant of the AFM spins. The equilibrium domain size can be obtained by minimizing the total energy per unit FM-AFM interfacial area $\sigma_{\text{FM-AFM}} + \pi(t_{\text{AFM}}/D_{\text{AFM}})$, which yields the following relationship between the AFM domain size and $t_{\text{AFM}}$: $D_{\text{AFM}} = \pi^3 J_{\text{AFM}} t_{\text{AFM}}/(2J_{\text{FM-AFM}})$. Hence, substituting this equation in the above expression of $\sigma_{\text{FM-AFM}}$, the inverse proportionality relation between $\sigma_{\text{FM-AFM}}$ and $t_{\text{AFM}}$ is readily obtained, which explains the variation of $H_E$ on $t_{\text{IrMn}}$ for continuous FM-AFM bilayers shown in Fig. 2. It is noteworthy that the domain state model for EB can also account for the AFM thickness dependence of $H_E$ [23].

In contrast to continuous films, $H_E$ seems to be roughly independent of $t_{\text{AFM}}$ for the nanostructures. This suggests that the mechanism responsible for the $H_E$ reduction in the continuous film is probably not operative in FM-AFM nanostructures. Actually, taking into account reasonable values of $J_{\text{FM-AFM}}$ and $J_{\text{AFM}}$ from the literature, i.e., $J_{\text{FM-AFM}} = 7.6 \times 10^{-15}$ erg and $J_{\text{AFM}} = 16.1 \times 10^{-15}$ erg [26], one can estimate that $D_{\text{AFM}}$ is about 160 nm for $t_{\text{IrMn}} = 5$ nm and it increases progressively up to 620 nm for $t_{\text{IrMn}} = 19$ nm, which is consistent with direct x-ray photoemission electron microscopy observations [30]. Additionally, taking into account the values of the magnetic stiffness for IrMn ($A_{\text{IrMn}} \approx 10^{-6}$ erg cm$^{-1}$)
and its magnetic anisotropy \((K_{IrMn} = 1.8 \times 10^4 \text{ erg cm}^{-3})\), the Bloch domain wall width in \(IrMn\), \(\delta_{IrMn}\), can be estimated to be \(\delta_{IrMn} = \pi(A_{IrMn}/K_{IrMn})^{1/2} = 25 \text{ nm}\) [31,32]. Since the dot size is smaller than the AFM domain size, it is likely that only a single AFM domain (with still local variations in the AFM order due to frustrations such as roughness, defects...) can form inside the nanostructures. Hence, for sub-100 nm dots, the lateral dimensions of the dots physically limit the AFM domain size, thus keeping \(H_E\) constant for all \(IrMn\) thicknesses.

Malozemoff’s model provides an estimation of the AFM domain size. However, it is a static model which does not take into account any reorganization of the AFM spin lattice during the magnetization reversal of the FM. The fact that the AFM domain size imposed by the lateral dimensions of the dots is always smaller than \(D_{AFM}\) in continuous films but, conversely, that \(H_E\) is not always larger in the nanostructures, seems to indicate that the AFM spin structure is less effectively pinned in the nanostructures than in the continuous films. Evidence for this hypothesis is obtained from the behavior of the blocking temperature, \(T_B\), and the coercivity. The dependence of \(T_B\) on \(t_{IrMn}\) for the continuous bilayers and the arrays of dots is shown in Fig. 3. The figure reveals that, in both systems, \(T_B\) increases with \(t_{IrMn}\), as typically observed in EB bilayers [1]. However, \(T_B\) remains lower for the nanostructures in the overall range of \(t_{IrMn}\). Hence, thermal activation effects in the AFM are more pronounced in the nanostructures than in the continuous films. Moreover, a \(H_E\) enhancement is observed in the nanostructures with respect to the continuous films (see inset of Fig. 2). This can be ascribed to an AFM spin dragging occurring during the magnetization reversal of the FM, which could be more pronounced in the dots than in continuous films. Interestingly, some published temperature-dependent results on EB in nanostructures [13] also show that, at room temperature, the pinning strength exerted by the AFM can be weaker in the nanostructures. Moreover, at low temperatures, where thermal activation effects are minimized, SQUID measurements also indicate that exchange bias in the nanostructures can be larger than for the continuous films, in agreement with Malozemoff’s model prediction (since the dot size, i.e., the AFM domain size in the nanostructures, is smaller than the estimated AFM domain size for the continuous films).

The weakening of the pinning strength in the nanostructures could stem from the reduced coordination number of AFM spins located at the edges of the dots. These spins can be more easily rotated during magnetization reversal of the FM. In addition, defects located in the bulk of the AFM layer would, in principle, favor the formation of domain walls in the AFM [33]. However, since the formation of domain walls is energetically unfavorable in the nanostructures, these defects are likely to distort the AFM order only locally, causing local twists of the AFM spins, which can easily unwind during the magnetization reversal of the FM.

It is noteworthy that the difference of blocking temperatures, \(\Delta T_B = T_{B,\text{continuous}} - T_{B,\text{dots}}\), increases for low \(t_{IrMn}\) values (see inset of Fig. 3), where the \(H_E\) enhancement is also more pronounced (see inset of Fig. 2). Additionally, although \(H_E\) remains constant at room temperature for all AFM thicknesses, \(H_E\) for thin AFM layers becomes progressively further reduced than for thicker AFM layers when the samples are heated again to a certain temperature \(T < T_B\) and cooled back to room temperature using negative fields, i.e., when trying to reset exchange bias in the

![FIG. 3. Dependence of the blocking temperature, \(T_B\), on the \(IrMn\) thickness, \(t_{IrMn}\), for the continuous films (○○○) and the nanostructures (■■■). The inset shows the dependence of the relative difference of blocking temperatures between the continuous bilayers and the nanostructures, \(\Delta T_B\), on \(t_{IrMn}\). The lines are guides to the eye.](image)

![FIG. 4. Dependence of the exchange bias field, \(H_E\), on the Py thickness, \(t_{Py}\), for both the continuous films and the nanostructures with compositions \(Ta(5 \text{ nm})/Py(t_{Py})/IrMn(5 \text{ nm})/Pt(2 \text{ nm})\). The full and dashed lines are \(1/t_{Py}\) fits of the \(H_E\) evolutions in continuous films and in nanostructures, respectively.](image)
opposite direction. Thus, the relative unpinning of AFM domains in the dots, compared to continuous films, seems slightly more manifest for thin AFM layers. It has to be noted that the technological inconvenience of a $T_B$ reduction in nanostructures can be minimized by using thick AFM layers, where one tends to recover the same $T_B$ as in continuous films.

Figure 4 shows the evolutions of $H_E$ as a function of $1/t_{py}$ for continuous films and nanostructures with compositions Ta (5 nm)/Py ($t_{py}$)/IrMn (5 nm)/Pt (2 nm). In both cases, $H_E$ follows the well known $1/t_{py}$ law, as typically observed in EB continuous bilayers [1,28,29,34], but the slopes are different, a consequence of the different FM-AFM coupling strength already discussed for low $t_{AFM}$ values. Up to now, the scaling of this law to small lateral dimensions ended in the micron sized elements range [10,14]. Such $1/t_{FM}$ evolution of $H_E$ highlights the main interfacial character of the FM-AFM coupling in both continuous films and nanostructures.

In conclusion, in sub-100 nm Py/IrMn nanostructures, the hysteresis loop shift can be either larger or smaller than that of continuous films with the same composition, depending on the AFM layer thickness. An enhancement of the coercivity and a reduction of the blocking temperature were also observed. These effects are ascribed to the three-dimensional confinement of AFM domains in the nanostructures. Our work sheds light towards the understanding of the controversial results published in the literature about dimensional confinement of AFM domains in the nanostructures on EB. This work was supported by the European Community through the NEXBIAS Grant No. HPRN-CT-2002-00296.

*Electronic address: baltz@drfmc.ceng.cea.fr

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