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# Antiferromagnetic thickness and temperature dependence of the exchange bias properties of Co/IrMn nanodots and continuous films: A Monte Carlo study

H. Kanso, R. Patte and D. Ledue

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## Abstract

Motivated by the challenge of understanding the complex influence of the antiferromagnetic (AF) thickness and the temperature on exchange bias (EB) properties, and by the necessity of miniaturization of devices, we investigate EB properties of Co/IrMn nanodots and of continuous films by using kinetic Monte Carlo simulations. To that purpose, we use a granular model, which takes into account disordered interfacial phases in the AF layer and, in the case of nanodots, disordered phases at the edges in the AF layer. Our results show that the AF thickness dependence of the exchange field  $H_E$  (measured at room temperature) in both nanodots and continuous films exhibits a maximum in agreement with experimental results. We explain these results in terms of superparamagnetic and blocked grains in the AF layer at room temperature and also not polarized AF grains during the initial field-cooling. The simulated values of  $H_E$  in nanodots are smaller than that in continuous films for small AF thicknesses and larger for larger ones due to the contribution of the disordered phases at the edges in the AF layer. Also, we investigate the temperature and AF thickness effects on  $H_E$  and on the coercive field  $H_C$ . We found that  $H_E$  slightly decreases at low temperatures due to the disordered interfacial phases. Importantly, at the maximum blocking temperature of the AF grains,  $H_E$  vanishes and  $H_C$  exhibits a maximum. Our numerical results are successfully compared to experimental data on Co/IrMn bilayers for various IrMn thicknesses and all temperatures. In addition, our results indicate that  $H_E$  is smaller in nanodots at low measurement temperature due to the presence of disordered phases at the edges. Concerning  $H_C$ , our data show that it can be either larger or smaller in nanodots depending on the measurement temperature.

## I. INTRODUCTION

The phenomenon of EB [1-4] has been extensively investigated in the last few decades, mainly in continuous films, from both experimental and fundamental points of view. EB occurs due to the coupling at the interface between ferromagnetic (F) and AF materials and depends on various parameters such as bulk and interfacial exchange couplings, bulk anisotropies, layer thicknesses, measurement and cooling temperatures ... Among these parameters, the measurement temperature and the AF layer thickness ( $t_{AF}$ ) dependence of EB properties are the most frequently investigated from an experimental point of view. It should be noted that most of the theoretical models for EB properties do not account for thermal effects [5-9]. Recently, many experimental studies were conducted to well understand the thermal effects on the EB properties and they have shown some common behaviors of EB properties and some contradictory behaviors [10-17]. The main common behavior is that the exchange field  $H_E$  and the coercive field  $H_C$  decrease as the temperature increases at low temperatures. As the temperature increases,  $H_C$  increases and reaches a peak, while  $H_E$  still decreases in some studies [10-14] or increases and reaches a peak in other studies [15-17]. At higher temperatures, both  $H_E$  and  $H_C$  decrease again with increasing temperature. Now, considering the effect of  $t_{AF}$  the results indicate either a decrease [18,19] or a maximum [20, 21] of  $H_E$  as  $t_{AF}$  increases. The decrease of  $H_E$  is usually ascribed, in a random field model, to the presence of AF domains, whose structure strongly depends on  $t_{AF}$  [7,22, 23]. The presence of a maximum in  $H_E$  is attributed to a thermal effect on the stability of the grains.

More recently, since EB is used in spintronic devices such as spin valves and magnetic tunnel junctions [24, 25], and because of the necessity of increasing the magnetic storage density and reducing devices sizes [26], investigations have been carried out on nanodots with lateral sizes of few hundred nanometers [27]. Then, in addition to the various parameters which affect EB properties in continuous films, size and boundary effects play an important role in nanodots which complicates the understanding of EB properties. Indeed, contradictory results on  $H_E$  in nanodots compared to that of continuous films have been reported. For example, in NiFe/IrMn bilayers, it was observed that  $H_E$  at room temperature is larger in nanodots for thicknesses  $t_{IrMn} > 11$  nm and smaller below this thickness compared to continuous films [18,28]. On the other hand, in Ref. [29] opposite results have been found. In another recent study on Co/IrMn nanodots at room temperature [20], it was shown that the dot lateral size has no significant effect on  $H_E$  ( $3 \text{ nm} \leq t_{IrMn} \leq 15 \text{ nm}$ ). For the same system Co/IrMn ( $t_{IrMn} = 7 \text{ nm}$ ), it was observed that  $H_E$  at 4 K is smaller in nanodots than in

continuous films [30]. An important difference between nanodots and continuous films is the dot edges which consist of additional locations for the formation of disordered magnetic phases [30]. Nevertheless, few numerical studies on EB properties in F/AF nanodots have been carried out. A granular approach has shown a significant difference in  $H_E$  between nanodots and continuous films [31]. However, we wish to emphasize that the model used is rather simple since it does not include disordered phases neither at the F/AF interface, nor at the edges in the AF layer although these disordered phases have significant effect on EB properties [7,32,33]. Finally, let us mention an atomistic model which has shown an increase of  $H_E$  for systems patterned with small FM dots compared to continuous films [34].

In this study, we investigate EB properties of square F/AF nanodots and continuous films based on a granular model which accounts for the disordered interfacial phases by considering less stable magnetic grains at the interface in the AF layer [35, 36]. In addition, we model the effect of the nanofabrication process by the presence of less stable grains at the edges of the nanodots in the AF layer as experimentally demonstrated [30]. We first study the effect of the AF thickness on the EB properties of nanodots in comparison with continuous films at room temperature. Then we investigate the effect of temperature on the EB properties of nanodots and continuous films. In these two cases, our results are compared to experimental data on Co/IrMn bilayers. Our work is carried out by kinetic Monte Carlo simulations.

The remaining of the paper is organized as follows: The model and simulation technique are described in Sec. II. Numerical results of the effect of the AF thickness and temperature in Co/IrMn nanodots and continuous films in comparison with experimental data are given in Sec. III, and a conclusion is given in Sec. IV.

## II. MODEL AND SIMULATION

The model is essentially that described in details in Refs [35, 36]. The F and AF layers are modeled by sets of grains coupled by exchange interactions. Since we assume columnar growth, they have the same granular microstructure which is created using a Voronoi tessellation [37] in two-dimensions. In nanodots, we model the effect of the nanofabrication process by the presence of grains with reduced areas at the edges due to grain cutting in the two layers. In Fig. 1 we present the top view of the granular microstructure for the continuous film and a nanodot, and the corresponding grain volume distribution for each case.

Importantly, the presence of smaller grains at the edges is noticeable in the nanodot compared to the continuous film.

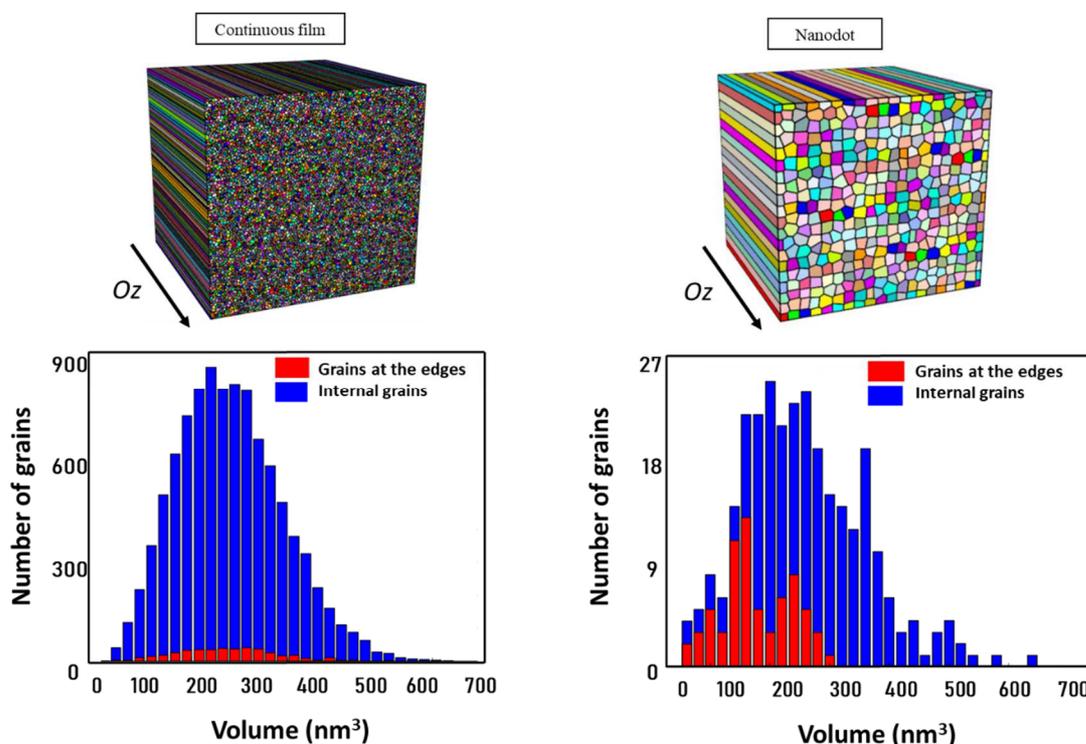


Fig. 1. (Color online) Top view of the granular microstructure created by Voronoi tessellation for the continuous film ( $10^4$  grains) and a nanodot (289 grains), and the corresponding grains volume distribution.

In addition to F and AF grains, small grains (SG) of thickness  $t_{SG} < t_{AF}$  are spread randomly over the F/AF interface within the AF layer in both nanodots and continuous films (Fig. 2) [35, 36]. These SG take into account disordered interfacial phases [7-32, 33] due to magnetic frustration produced by defects at the F/AF interface (e. g. interlayer diffusion and stacking faults). So, we consider that these SG exhibit altered magnetic properties (reduced anisotropy and coupling compared to the bulk AF grains). Assuming that the disordered interfacial magnetic phases extended over 3-4 atomic planes [38, 39], we set  $t_{SG} = 2$  nm. Since, the fraction of SG at the F/AF interface  $x_{SG}$  may be varied between about 20% and 80%, as it depends on the fabrication process [40], we set  $x_{SG}$  to an average value of 50% in both nanodots and continuous films. In addition to that, we assume that the grains which are at the edges in the AF layer in nanodots (which will be referred as  $SG_E$ ) exhibit the same magnetic

properties as the SG (Fig. 2) [30]. Accordingly, the fraction of grains with altered magnetic properties (SG and SG<sub>E</sub>) is larger in nanodots since there is no SG<sub>E</sub> in continuous films. Note that since  $t_{SGE} = t_{AF}$  is larger than  $t_{SG}$ , the blocking temperature ( $T_B$ ) distribution of a nanodot differs from that a continuous film.

The exchange energy between two grains is

$$E_{ex} = -J_{ij} \boldsymbol{\sigma}_i \cdot \boldsymbol{\sigma}_j$$

where  $J_{ij}$  is the exchange constant ( $J_{F-AF}$ ,  $J_{F-SG}$  or  $J_{F-F}$ ) and  $\boldsymbol{\sigma}_i$ ,  $\boldsymbol{\sigma}_j$  are unit vectors representing the magnetization orientation of a F grain, a SG, a SG<sub>E</sub> or the interfacial uncompensated magnetization orientation of an AF grain. Note that the AF grains are not linked to each other [41,43], and nor are the SG and the SG<sub>E</sub> (Fig. 2). **Indeed the SG (and SG<sub>E</sub>) and the AF grains are not linked to each other because such couplings do not affect directly the exchange bias properties of the bilayer. The couplings which contribute directly to  $H_E$  are the interfacial couplings: between the F grains and the AF grains ( $J_{F-AF}$ ) and between the F grains and the smaller grains ( $J_{F-SG}$ ) (the smaller grains are the SG in continuous films and the SG and SG<sub>E</sub> in case of nanodots). Actually, a coupling between SG or SG<sub>E</sub> or between AF grains would slightly affect their blocking temperature and thus could affect very slightly the exchange bias properties, so we have neglected them. In addition, concerning SG (or SG<sub>E</sub>), couplings between them are expected to be small because of the non-collinearity of the spins in the disordered interfacial and edges phases, we expect that they do not affect the exchange bias properties. Assuming an uniaxial anisotropy energy along a common easy axis (y-axis) in the plane of the layer, the anisotropy energy is**

$$E_a = -K_i V_i (\boldsymbol{\sigma}_i \cdot \mathbf{e}_y)^2$$

where  $K_i$  is the anisotropy constant per unit volume ( $K_F$ ,  $K_{AF}$  or  $K_{SG}$ ) and  $V_i$  is the volume of the grain. **We make this assumption for simplicity and also because we think that the results would not change qualitatively. Indeed, in case of random easy axes, the energy barriers to overcome during the reversal of the grains would be slightly modified which result in a small change in the blocking temperature of the grains. So, only a small change in our results could be observed. Finally, let us mention that such a magnetic texture could be obtained by applying an external field along the y-axis during the fabrication of the samples. A linear temperature dependence of the anisotropy constants per unit volume is implemented [44, 14] with  $T_N = 690$  K [3]. Also, a Zeeman term**

$$E_Z = -\mu_0 \mathbf{H} \cdot \mathbf{m}$$

is taken into account where  $\mathbf{m}_i$  is the magnetic moment of a F grain or a SG or a SG<sub>E</sub> (in the nanodots) and  $\mathbf{H}$  is the field applied along the  $y$ -axis. The magnetic moments  $m_i$  of F grains, SG and SG<sub>E</sub> are calculated according to  $m_i = M V_i$  where  $M$  is the magnetization of the grain and  $V_i$  is its volume. For the F grains, we have taken the Co magnetization  $M = 1,44 \times 10^6$  A m<sup>-1</sup> [45].

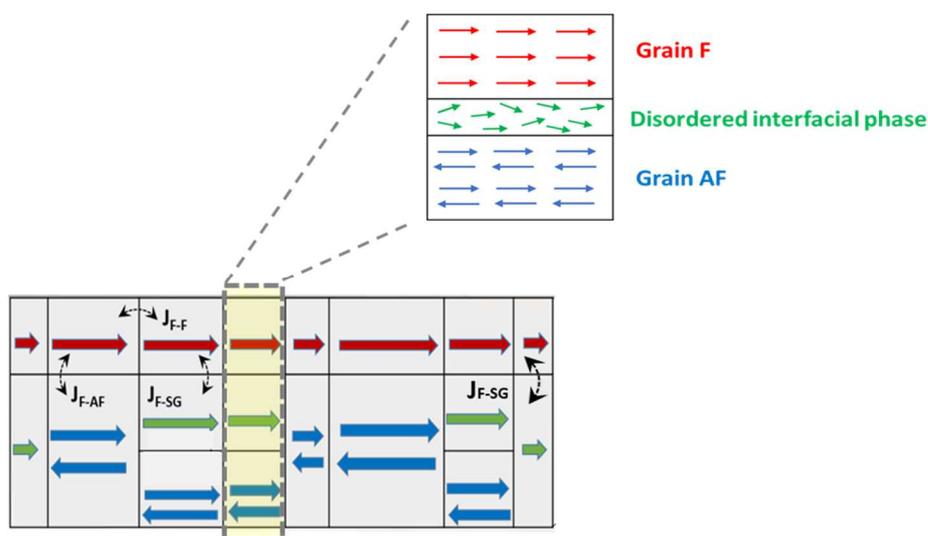


Fig. 2. (Color online) Sketch of the F/AF nanodot with SG randomly spread over the F/AF interface (in green) and SG<sub>E</sub> at the edges in the AF layer (in green too). The small sketch above represents the collinear atomic moments of one F grain and one AF grain, and the non-collinear atomic moments of the disordered interfacial phase which are on-average along the  $y$ -axis as shown in the large sketch.

Some parameters of our model can be estimated from experimental results. For example, from experimental values of  $H_E$  [19], we found that  $j_{F-AF} = 4.5 \times 10^{-4}$  J m<sup>-2</sup> is a reasonable value in Co/IrMn bilayers. Likewise, the 0 K anisotropy constant for the AF grains was estimated to be  $K_{AF}^0 = 4 \times 10^5$  J m<sup>-3</sup> [46]. By contrast, the coupling per unit area,  $j_{F-SG}$ , and the effective anisotropy of SG and SG<sub>E</sub> are unknown. However,  $j_{F-SG}$  is assumed to be smaller than  $j_{F-AF}$ , so we set  $j_{F-SG} = 3 \times 10^{-4}$  J m<sup>-2</sup>. The effective anisotropy of the SG was also assumed to be weaker than that of the AF grains [30], we arbitrarily set  $K_{SG}^0$  to  $0.75 K_{AF}^0$ . In order to be in the weak coupling regime where the exchange field is independent on the F coupling ( $j_{F-F}$ ) and there is

no variability between one nanodot to another, we choose  $j_{F-F} = 10^{-4} \text{ J m}^{-2}$  [47]. The fixed parameters of all our simulations are summarized in Table 1.

$j_{F-F} = 10^{-4} \text{ J m}^{-2}$	$j_{F-AF} = 4.5 \times 10^{-4} \text{ J/m}^2$	$j_{F-SG} = 3 \times 10^{-4} \text{ J/m}^2$
$T_N = 690 \text{ K}$	$K_{AF}^0 = 4 \times 10^5 \text{ J/m}^3$	$K_{SG}^0 = 3 \times 10^5 \text{ J/m}^3$
$x_{SG} = 50\%$	$t_{SG} = 2 \text{ nm}$	$t_{SGE} = t_{AF}$

Table 1. Fixed parameters of all our simulations.

Our Monte Carlo simulations [14,35,48,49] are capable of simulating cooling or heating under an external magnetic field. Thus, we can calculate the temperature dependence of the magnetic properties such as hysteresis loops from which the exchange and the coercive fields can be extracted.

### III. RESULTS AND DISCUSSION

For all results presented below, we have simulated an initial field cooling (FC) from  $T_0$  down to  $T_f$ . The hysteresis loops are simulated at a fixed  $T_M = 298 \text{ K}$  in Sec. III.A or at increasing  $T_M$  ( $T_f \leq T_M < T_0$ ) in Sec. III.B. To ease the understanding, we want to emphasize that the grains (in the AF layer) in contact with the F layer can be separated into three groups depending on their  $T_B$  and only one group contributes to  $H_E$ : (i) the grains with  $T_B > T_0$ , are not polarized by the external field and thus have a zero net magnetization and do not contribute to  $H_E$ ; (ii) the grains with  $T_M < T_B < T_0$ , are polarized by the external field and are blocked at the measuring temperature  $T_M$ , so they contribute to  $H_E$ ; (iii) the grains with  $T_B < T_M$ , are polarized by the external field but they are superparamagnetic at  $T_M$  so they do not contribute to  $H_E$ . Since, the experimental data that we compare with them correspond to an average over numerous nanodots, we therefore average our simulation results over several nanodots. We found that 60 nanodots of averaging is enough for convergence.

#### A. Effect of the AF thickness ( $T_M = 298 \text{ K}$ )

In this part, we study the influence of the AF thickness on the EB properties, and then we compare our results with experimental data presented in Ref. [20]. To do so, we model Co/IrMn bilayers with  $t_{Co} = 5 \text{ nm}$  and various  $t_{IrMn}$  ( $3 \text{ nm} < t_{IrMn} < 14 \text{ nm}$ ). For each value of  $t_{IrMn}$ , the hysteresis loops are simulated at  $T_M = 298 \text{ K}$  after FC under  $H_{FC} = 2 \text{ kOe}$  from

$T_0 = 473$  K to  $T_f = 298$  K. Figure 3 presents the hysteresis loops for  $t_{\text{IrMn}} = 4, 8$  and  $14$  nm for the continuous film and nanodots. The  $t_{\text{IrMn}}$ -dependence of  $H_E$  in comparison with experimental data is shown in Fig. 4 [20]. Similar to the experimental data, our results show that the  $H_E$  curve exhibits a maximum in both cases of nanodots and continuous film. Note that these maxima are found at a slightly larger thickness than that reported experimentally (6.5 nm).

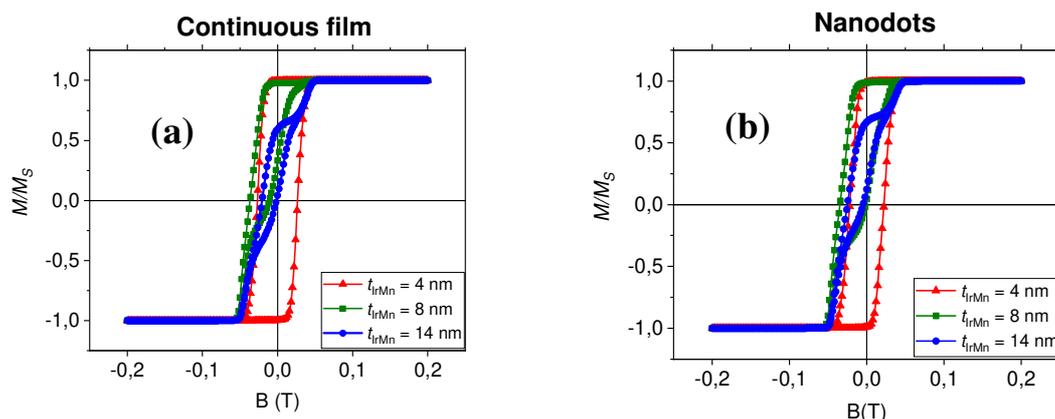


Fig. 3. (Color online) Simulated hysteresis loops at 298 K for various  $t_{\text{IrMn}}$ , for (a) the continuous film and (b) nanodots (averaged over 60 nanodots).

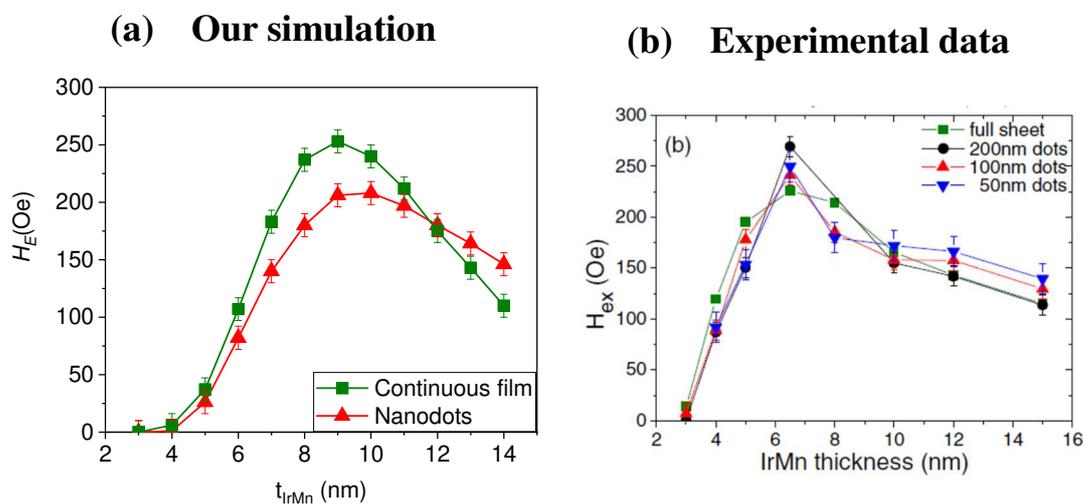


Fig. 4. (Color online) (a) Simulated  $t_{\text{IrMn}}$ -dependence of  $H_E$  measured at  $T_M = 298$  K for Co/IrMn bilayers in comparison with (b) experimental data [20].

The behavior of  $H_E$  results from the  $t_{\text{IrMn}}$ -dependence of the fraction of grains in contact with the F layer which contribute to  $H_E$ , namely those which have blocking temperatures between  $T_M = 298$  K and  $T_0 = 473$  K. More precisely, in the continuous film, SG do not contribute to

$H_E$  since their  $T_B$  are smaller than 298 K (Fig. 5), while only a fraction of AF grains does contribute where this fraction depends on  $t_{\text{IrMn}}$ . For small thicknesses (Fig. 5.a), the fraction of AF grains which contributes to  $H_E$  is small and thus the value of  $H_E$  is small. When  $t_{\text{IrMn}}$  increases, this fraction increases and reaches a maximum at  $t_{\text{IrMn}} \approx 9$  nm (Fig. 5.b) leading to a maximum in  $H_E$ . For  $t_{\text{IrMn}} > 9$  nm, the fraction of the contributing AF grains decreases and thus  $H_E$  (Fig. 5.c). The small shift of our curves toward higher thicknesses compared to the experimental data is due to the small shift of the blocking temperature distribution of the AF grains in our model compared to the experimental one. Note that no difference between nanodots and the continuous films can be detected in experimental data due to error bars whereas our results indicate a smaller  $H_E$  in nanodots for  $t_{\text{IrMn}} < 12$  nm, and a larger one if  $t_{\text{IrMn}} > 12$  nm. This can be explained by the contribution of  $\text{SG}_E$  in nanodots which becomes more pronounced as  $t_{\text{IrMn}}$  increases. For  $t_{\text{IrMn}} < 12$  nm, all  $\text{SG}_E$  are superparamagnetic and do not contribute to  $H_E$ , therefore the simulated values of  $H_E$  are slightly smaller than that in continuous films. However, for  $t_{\text{IrMn}} > 12$  nm, most of  $\text{SG}_E$  contribute to  $H_E$ , and thus the simulated values of  $H_E$  are slightly larger than that in continuous films.

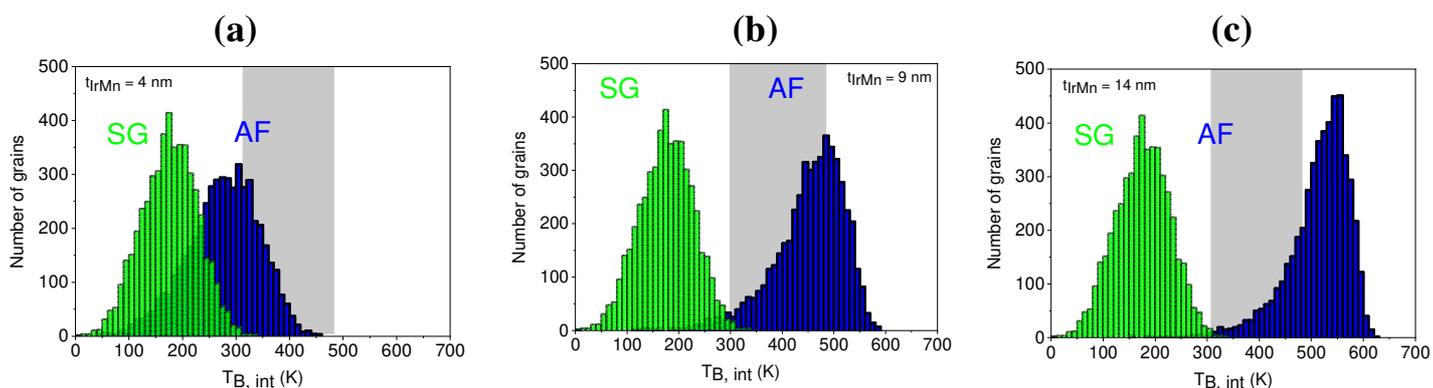


Fig. 5. (Color online) Intrinsic blocking temperature distributions for the grains in contact with the F layer at the interface in a Co/IrMn continuous film for (a)  $t_{\text{IrMn}} = 4$  nm, (b)  $t_{\text{IrMn}} = 9$  nm and (c)  $t_{\text{IrMn}} = 14$  nm (the grey rectangles indicate the grains which contributes to  $H_E$ )

The  $t_{\text{IrMn}}$ -dependence of  $H_C$  is plotted in Fig. 6 in comparison with experimental data [20]. Our results reproduce qualitatively the experimental observations in a continuous film with a maximum of  $H_C$  at  $t_{\text{IrMn}} = 3-4$  nm. In regards to  $H_C$ , the grains which contribute are those which are in contact with the F layer and having  $T_B \approx T_M$ . Since all SG have  $T_B < T_M$  for all IrMn thicknesses (see Fig. 5), only a fraction of the AF grains in contact with the F layer contribute to  $H_C$ . Then according to Fig. 5, it can be seen that the fraction of these grains is

maximum at  $t_{\text{IrMn}} = 4$  nm and decreases for  $t_{\text{IrMn}} > 4$  nm which explains the behavior of  $H_C$ .

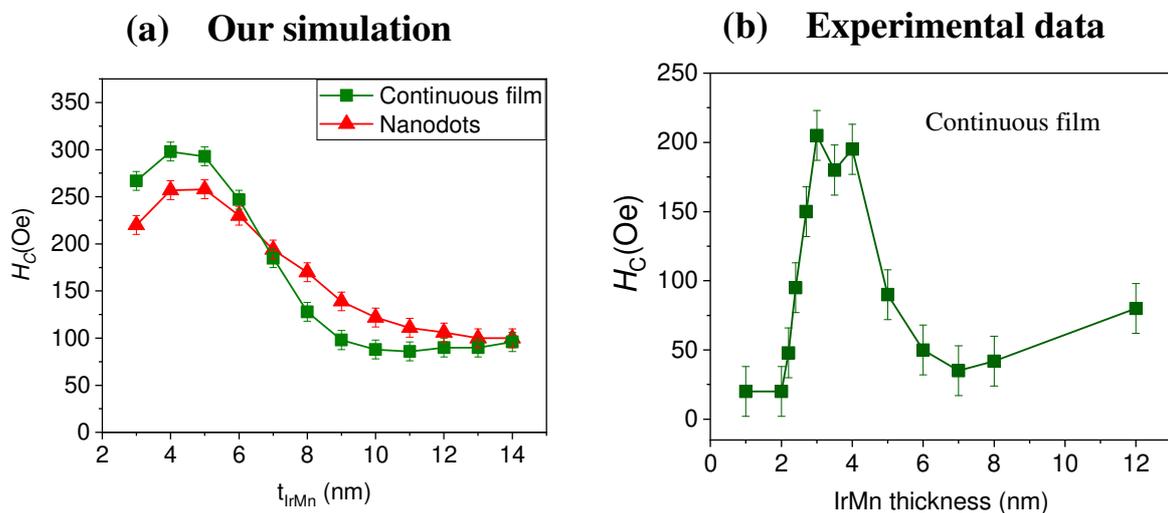


Fig. 6. (Color online) (a) Simulated  $t_{\text{IrMn}}$ -dependence of  $H_C$  measured at  $T_M = 298$  K for Co/IrMn bilayers in comparison with (b) experimental data [20].

Comparing the case of the continuous film with nanodots, our results indicate that  $H_C$  is smaller in nanodots for  $t_{\text{IrMn}} < 7$  nm while it is larger when  $t_{\text{IrMn}} > 7$  nm. This is because the fraction of grains which contribute to  $H_C$  is smaller in nanodots for  $t_{\text{IrMn}} < 7$  nm due to the presence of the  $\text{SG}_E$  which are superparamagnetic and thus do not contribute to  $H_C$ . However, for  $7 \text{ nm} < t_{\text{IrMn}} < 12$  nm,  $H_C$  is larger in nanodots because the blocking temperatures of  $\text{SG}_E$  becomes closer to  $T_M$ , so they contribute to  $H_C$ .

## B. Effect of the temperature and AF thickness

In this section, we study the effects of the AF thickness on the temperature dependence of the exchange and coercive fields in comparison with experimental data [19]. For that we consider a Co/IrMn bilayer with  $t_{\text{Co}} = 4$  nm and various IrMn layer thicknesses ( $t_{\text{IrMn}} = 3$  nm, 9 nm and 15 nm). The simulated procedure consists of a FC under  $H_{FC} = 4$  kOe from  $T_0 = 550$  K down to  $T_f = 2$  K. Then successive hysteresis loops are simulated at increasing temperatures  $T_M \geq 2$  K. The simulated temperature dependence of  $H_E$  and  $H_C$  for various  $t_{\text{IrMn}}$  in the continuous film is shown in Fig. 7 in comparison with those of Ref. [19].

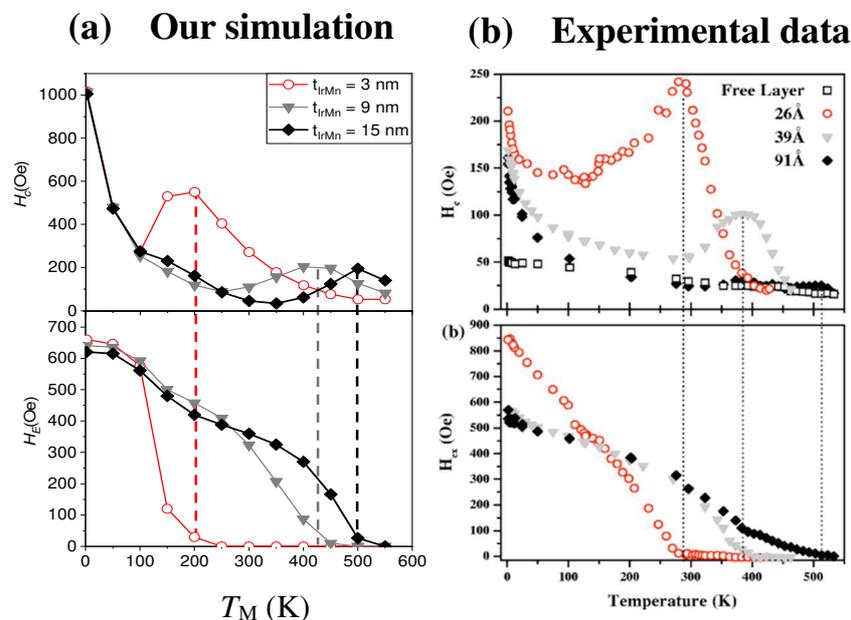


Fig. 7. (Color online) (a) Simulated temperature dependence of  $H_E$  and  $H_c$  in Co/IrMn bilayers (for various  $t_{IrMn} = 3$  nm, 9 nm and 15 nm) in the continuous film in comparison with (b) experimental data [19].

Concerning  $H_E$ , a good qualitative agreement with experimental data is obtained above 100 K:  $H_E$  decreases as  $T_M$  increases and vanishes at a temperature which increases with  $t_{IrMn}$ . The decrease of  $H_E$  is due to the increase of the fraction of superparamagnetic grains in contact with the F layer (AF grains and SG) when  $T_M$  increases. Note that the decrease below 200 K is attributed to the SG, *i.e.* to the disordered interfacial phases since a model without SG cannot reproduce this decrease [14]. Then  $H_E$  vanishes at a temperature corresponding to the

maximum blocking temperature of the AF grains which increases with  $V_{AF}^{\max} \propto t_{IrMn}$ . We can compare the vanishing temperature of  $H_E$  with the maximum of the intrinsic blocking temperature  $T_B^{\max} = \frac{K_{AF}^0 V_{AF}^{\max} T_N}{k_B T_N \ln(2n) + K_{AF}^0 V_{AF}^{\max}}$  where  $n$  is the number of Monte Carlo steps [36]

(Fig. 8). The vanishing temperature of  $H_E$  exhibits, as expected, the same  $t_{IrMn}$ -dependence as  $T_B^{\max}$ , however it is roughly 150-200 K lower than  $T_B^{\max}$  clearly showing that the coupling with

the neighboring F grains results in a decrease of the (effective) blocking temperature of the AF grains. Finally, note that, at  $T_M = 2$  K, the simulated value of  $H_E$  slightly decreases as  $t_{\text{IrMn}}$  increases. This is because the fraction of grains which contribute to  $H_E$  (those with  $T_B < 550$  K) decreases as  $t_{\text{IrMn}}$  increases as shown in Fig. 9.

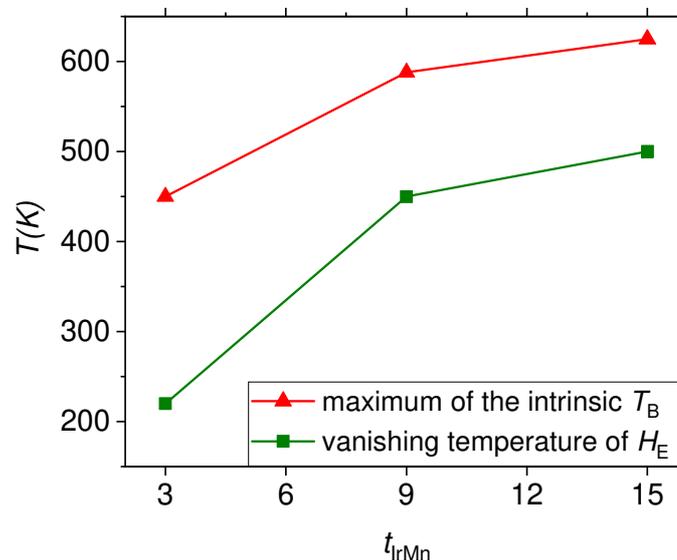


Fig. 8. Comparison between the vanishing temperature of the simulated  $H_E$  and the maximum of the intrinsic blocking temperature versus  $t_{\text{IrMn}}$ .

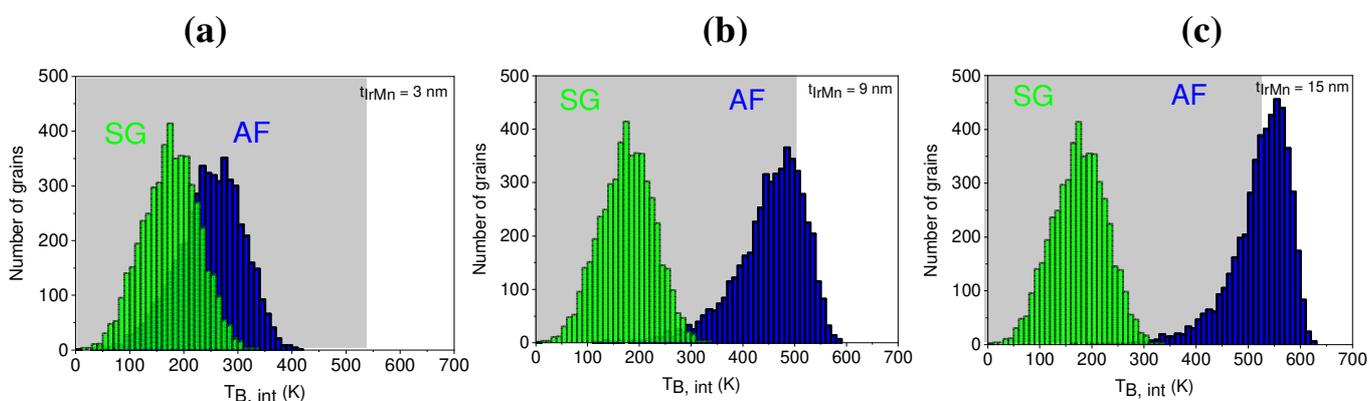


Fig. 9. (Color online) Intrinsic blocking temperature distributions for the grains in contact with the F layer at the interface in a Co/IrMn continuous film for (a)  $t_{\text{IrMn}} = 3$  nm, (b)  $t_{\text{IrMn}} = 9$  nm and (c)  $t_{\text{IrMn}} = 15$  nm (the grey rectangles indicate the grains which contribute to  $H_E$  at  $T_M = 2$  K).

Regarding  $H_C$ , a good qualitative agreement with experimental data is also obtained. In particular, a maximum in  $H_C$  is observed simultaneously at the vanishing temperature of  $H_E$ . This maximum shifts toward higher temperatures and its amplitude decreases as  $t_{\text{IrMn}}$  increases. Actually, this maximum results from both the shift of the blocking temperature

distribution of AF grains as  $t_{\text{IrMn}}$  increases and the fact that only the AF grains having  $T_B$  close to  $T_M$  contributes to  $H_C$  (see Fig. 9).

In order to see the effects of reducing the lateral size, we perform the same simulations for nanodots with  $t_{\text{IrMn}} = 9$  nm. The temperature dependence of  $H_E$  and  $H_C$  is plotted in Fig. 10 in comparison with that of a continuous film. The same behavior is observed in nanodots as in the continuous film. Note that at  $T_M = 2$  K, all grains in contact with the F layer contribute to  $H_E$ , where  $H_E$  is smaller in nanodots due to the presence of  $\text{SG}_E$  at the edges (which are less coupled with the F layer than the AF grains). Note also that the location of the minimum and the maximum of  $H_C$  is shifted towards lower temperatures in nanodots. This is because the fraction of grains contributing to  $H_C$ , at low temperatures, increases in the nanodots due to presence of  $\text{SG}_E$  (Fig. 11).

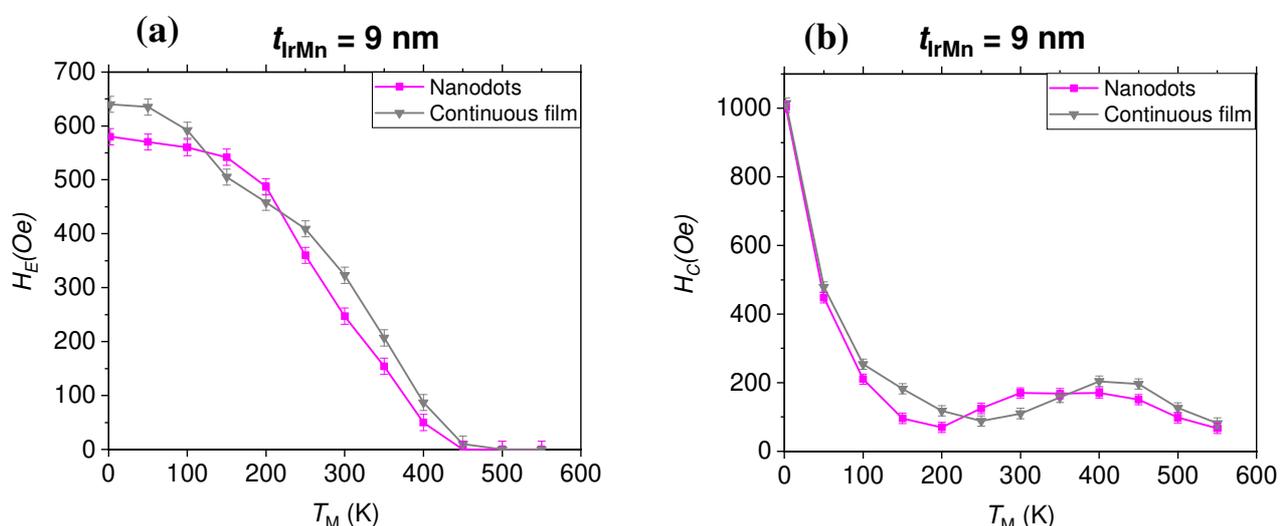


Fig. 10. (Color online) Simulated temperature dependence of (a)  $H_E$  and (b)  $H_C$  in Co/IrMn bilayers ( $t_{\text{IrMn}} = 9$  nm) in both nanodots and continuous film.

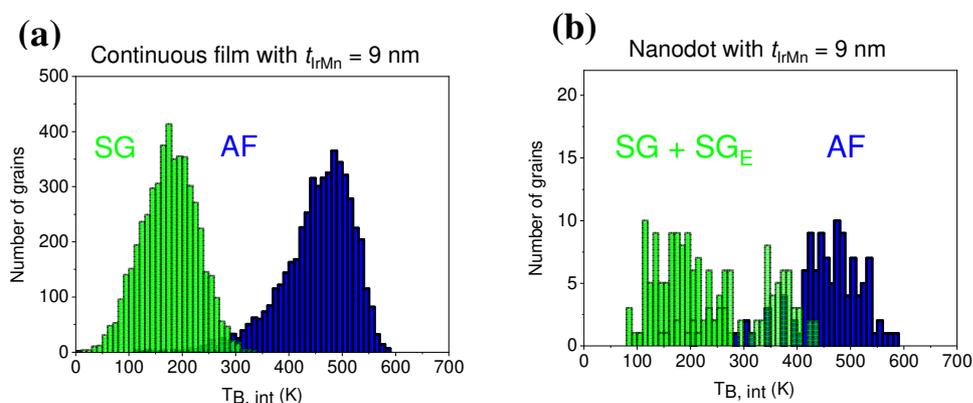


Fig. 11. (Color online) Intrinsic blocking temperature distributions of the grains in contact

with the F layer at the interface in a Co (4 nm)/IrMn (9 nm) for (a) the continuous film and (b) a nanodot.

#### IV. CONCLUSION

By using a granular model which takes into account disordered magnetic phases at the F/AF interface and at the edges in the AF layer in nanodots, we have proposed a comprehensive description of the AF thickness and temperature effects on the EB properties in Co/IrMn nanodots and in continuous films. Our model is based on the assumption of single domain grains and the disordered interfacial phases are modeled by less stable magnetic grains. Our numerical results are in good qualitative agreement with experiments for various AF thicknesses and measurement temperatures. More precisely, the AF thickness dependence of  $H_E$  exhibits a maximum at room temperature. Also, the temperature dependence of  $H_E$  shows a decrease at low temperatures due to the disordered phases and vanishes at the maximum blocking temperature of the AF grains for which  $H_C$  is maximum. Our results show that the simulated values of  $H_E$  in nanodots can be smaller or larger than that in continuous films depending on the AF thickness and the measurement temperature due to contribution of the disordered phases at the edges in the AF layer. Note that these results provide an explanation of the various experimental results found in literature about the comparison between EB properties of nanodots and continuous films. To achieve a deeper understanding of the disordered interfacial phases, it would be of great interest to use an atomic approach which takes into account grain boundaries and atomic interdiffusion in the AF layer. Note that to avoid huge computational time, such a study would be restricted to nanodots.

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