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Many spintronic devices such as thermally-assisted magnetic random access memories take advantage of the ferromagnetic(F)/antiferromagnetic(AF) interaction to pin the magnetization of a reference layer. However, they suffer from detrimental blocking temperatures distributions from memory cell to memory cell. A low-temperature contribution to the distribution was ascribed to spin-glass like regions which are randomly spread over the film. We report on an attempt to reduce the amount of these spin-glass like regions due to interdiffusion of species by adding a diffusion barrier at the F/AF interface. © *2011 American Institute of Physics*. [doi:10.1063/1.3558983]

INTRODUCTION

Setting a reference direction for the spin of conduction electrons is a prerequisite for most spintronic devices.[1] Exchange bias (EB), which refers to the exchange coupling between a ferromagnet (F) and an antiferromagnet (AF) is used for that purpose of defining a preferential magnetic orientation. It allows pinning the magnetization of an F layer by shifting its hysteresis loop along the magnetic field axis.[2] Some technological applications such as magnetic random access memories[3] require the patterning of films comprising EB layers into arrays of cells. These technologies suffer from detrimental distributions of the magnetic properties from cell to cell. The distribution of blocking temperature (T_B) (Refs. 4–7), (the temperature (T) over which AF grains are no longer thermally stable when cycling the F magnetization) lies at the heart of the problem. We have recently evidenced that T_B distributions consist of two parts:[8] (i) a commonly observed high T peak associated to T-activated reversal of the AF grains spin-lattice[4–10] and (ii) a more unconventional low T peak,[8,11] ascribed to F/AF interfacial spin-glass like regions characterized by low freezing T.[12,13] These regions result from frustrations due to interfacial roughness, grain boundaries, dislocations, interdiffusion of species… They are randomly spread over the film. Once patterned into arrays of cells they might contribute to the distributions of properties from cell to cell: a cell with many spin-glass like regions will show a weaker EB field (H_E) at room T since the F/AF interfacial coupling is disrupted on a large part of the cell area.[8] In addition the risk of loosing the reference spin direction for this cell increases. Even if the devices are operated at elevated T (typically around room T), the low-T part of the T_B distribution can lead to functional failure. If the continuous film already contains a significant low-T distribution of T_B, some cells will exhibit a too weak pinning of their reference layer and therefore be unusable at the working T. The eradication of the low-T contribution to T_B is thus mandatory for the implementation of EB in devices of steadily decreasing lateral dimensions. In our previous work on Mn-based AF combined with NiFe and Co based F, it was identified that the diffusion of Mn in the F layer plays an important role in creating interfacial frustrations.[8] Eradicating this source of frustration (and thus getting rid of the low-T contribution to the T_B distribution) may significantly contribute to reducing the dispersion of EB properties from cell to cell. Pt proved to be an efficient barrier to the diffusion of Mn.[14] In this study, we attempted to reduce the amount of spin-glass entities with low freezing T by adding Pt spacer at the F-AF interface of Co/IrMn bilayers. We report on the effect of these additions on the low-T distribution of T_B. A strong reduction of the amplitude of the low-T contribution to the distribution was initially expected from the insertion of the Pt barrier to Mn diffusion.

The samples were DC sputtered onto thermally oxidized sheet substrates of silicon: Si(300μm)/SiO_2(500nm).[8] The stacks were (from bottom to top) Ta (3 nm)/Cu (3 nm)/Co (3 nm)/Pt (t_{Pt})/IrMn (3.8 nm)/Pt (2 nm) with Pt thickness equal to 0, 0.4, and 1 nm. IrMn was deposited from an $Ir_{20}Mn_{80}$ target. Ta (3 nm)/Cu (3 nm) serves as buffer layers, Pt (2 nm) is a capping layer, Co (3 nm)/Pt (t_{Pt})/IrMn (3.8 nm) is the F/AF bilayer of interest, with the addition of a spacer between F and AF layers.

The initial EB was set in the insert of a superconducting quantum interference device (SQUID) magnetometer by post-annealing and field-cooling (FC) from 450 K, i.e., from above their maximum blocking temperature ($T_{B,max}$) down to 4 K in an in-plane positive magnetic field of 10 kOe, i.e., large enough to saturate the magnetization of the F layer. The T_B distributions in the range of 4 to 450 K were derived from hysteresis loops measured by SQUID at the same reference T of 4 K, following a specific FC procedure described in Ref. 15. Typical raw hysteresis loops measured during the procedure are shown in Fig. 1. The small linear contribution to the loops with negative slope is ascribed to the diamagnetic contribution of the sample holder and substrate. After the initial FC, all the AF entities contributing to EB are oriented toward the positive direction, which results in a loop shift with the maximum amplitude (for $T_a = 4$ K). The procedure then consists in gradually orienting AF entities toward the opposite direction. For that purpose, T is raised up

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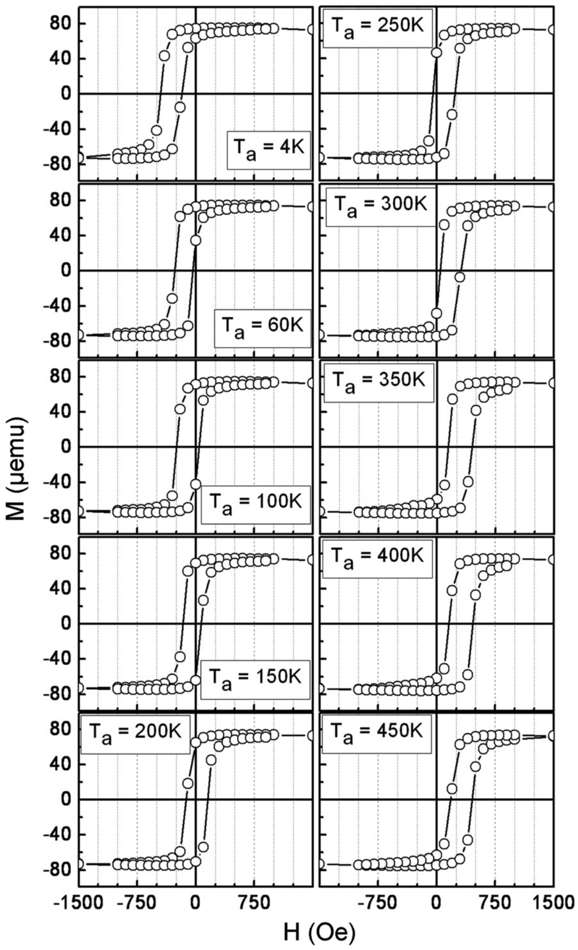

FIG. 1. Representative hysteresis loops measured by SQUID at 4 K along the field cooling (FC) direction, for different annealing temperatures (T_a), for a film of Ta (3 nm)/Cu (3 nm)/Co (3 nm)/Pt (1 nm)/IrMn (3.8 nm)/Pt (2 nm). The loops are subsequent to a specific cooling procedure: the samples were initially FC under a positive saturating field of 10 kOe from 450 K down to 4 K. T was then raised up to T_a and the samples were FC under −10 kOe from T_a down to 4 K.

to an intermediate annealing temperature (T_a). The layers are subsequently FC down to 4 K under a negative magnetic field and a hysteresis loop is measured. The above step is repeated for increasing values of T_a. At each increment, all AF entities with T_B between T_a and 4K are reoriented toward the negative direction.[8,15] Increasing T_a thus results in a gradual change in the amplitude and sign of H_E, which can indeed be observed in Fig. 1.

The values of H_E were determined from the series of hysteresis loops after correction from the linear diamagnetic contribution. The dependences of H_E on T_a are plotted in Fig. 2 and represent the integrand of the T_B distribution. Indeed, H_E at each increment of T_a is proportional to the difference between the amount of entities oriented positively and negatively.[8,15] H_C is related to the AF grains whose spin-lattice is dragged during the magnetization reversal of the F layer.[2] This effect is independent on the cooling history which explains the independence of H_c on T_a in Fig. 2 since all the loops are measured at 4 K. Within error bars, H_C can be estimated slightly above 300 Oe for $t_{Pt} = 0$, slightly below 300 Oe for $t_{Pt} = 0.4$ nm, and around 125 Oe for $t_{Pt} = 1$ nm. It thus appears that H_C is reduced when t_{Pt} is inserted: the short range coupling between the F and AF weakens and the ability for the F to drag the AF entities is lowered.[2] Also, due to the fact that EB involves a short range interfacial F/AF coupling,[2] one observes in Fig. 2 that the maximum value of H_E decreases when the Pt insertion is added. As expected from the above explanation of the experimental procedure, we observe in Fig. 2 that for all samples H_E increases with increasing T_a. It changes sign and it levels off when $T_{B,max}$ has been reached. It however seems that the value $T_{B,s}$ at which the distribution starts to level off, as indicated in Fig. 2, increases when the thickness of the Pt layer is increased. Striking differences in the shape of the dependences of H_E on T_a before leveling off are noticeable. For $t_{Pt} = 1$ nm an additional inflection point clearly appears at low T. For better comparison the dependences are shown together in Fig. 3(a) after normalization by the maximum value (at $T_a = 4$ K). The features in the H_E versus T_a plots yield significant differences in the derivatives of H_E versus T_a ($\delta H_E/\delta T_a$), which actually represent the T_B distributions,[8,15] An inflection point in Fig. 3(a) converts into a peak in the T_B distributions. The normalized distributions are plotted together in Fig. 3(b). The normalization results from the fact that the integral of the distributions is equal to 1: when the whole distribution has been covered 100% of the AF entities have responded.[8,15] We note that for $t_{Pt} = 0$ the shape of the T_B distribution is in agreement with previous works, for which a single broad T_B distribution is observed for an IrMn thickness below 4 nm.[8] It was observed that this broad peak is the result of the overlap of the high- and low-T contributions. For IrMn thicknesses of 4 nm and thicker, the two peaks in the distribution become distinct. The high-T peak shifts toward lower T when the IrMn thickness is reduced. For low enough thicknesses, the two peaks start overlapping.[8] The shift toward lower T of the high-T peak is yet not linear and the slope of the evolution of the average T_B with the AF layer thickness becomes more important for thin layers.[6] This might explain why our present sample with 3.8 nm of IrMn shows overlapping peaks in contrast to a sample with 4 nm of IrMn like in Ref. 8. We note that results for a 3.5 nm sample (not shown here) also shows a single peaked T_B distribution.

From Fig. 3(b) one observes that the addition of a Pt spacer has two consequences on both the low-T and high-T peaks. First, it results in a shift of the high-T peak toward higher T. Such a shift means that, on average, the AF grains are more stable against thermal activation.[7–10] This can be ascribed to an increase of AF anisotropy or to an increase in grain volume. Indeed, the structural growth of IrMn on top of a bare Co likely benefits from the presence of a Pt (111) spacer.[16] This shift of the T_B distribution toward higher T can also be explained by the reduction of the F/AF interfacial coupling due to the Pt insertion. It reduces the torque exerted by the F magnetization on the AF spin lattice of each individual AF grains when a hysteresis loop is performed.[8] Consequently, this allows the AF grains to resist to switching up to higher T. As a second noticeable consequence, which is the main point of this paper, the addition of a Pt spacer of 1 nm (i.e., thick enough to ensure the formation of a continuous layer of Pt) leads to the formation of a low-T peak with larger amplitude. Since the addition of a Pt spacer has been proven to limit the diffusion of Mn toward the Co layer,[14]

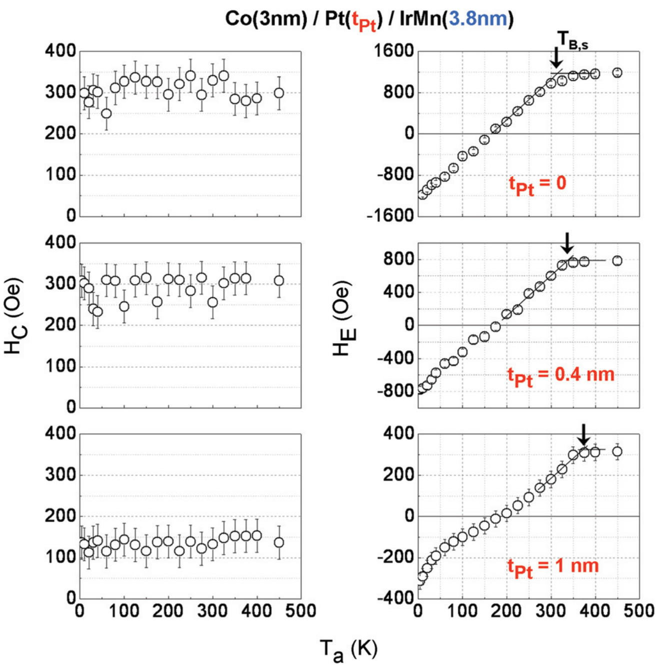

FIG. 2. (Color online) Dependence on T_a of the coercive field (H_C) and loop shift (H_E) as deduced from hysteresis loops measured at 4 K (cf. Fig. 1).

we were expecting a reduction of frustrations at the Co/IrMn interface and thus a reduction of the amplitude of the low-T peak in the T_B distribution.[8,14] Strikingly, it appears that despite a reduced Mn diffusion, the addition of a Pt spacer adds disordered interfacial spins which exhibit spin-glass like behavior with low freezing T. This may be ascribed to the formation of $CoPt_x$ alloys,[17] with reduced ordering T around the Co-Pt interface. This may also be linked to the weak magnetic polarization that could be induced in Pt rich clusters, which would enhance the amount of interfacial regions becoming stable only at low-T.[18] The derivative in Fig. 3(b) for the case of the Pt 0.4 nm gives an intermediate situation between no Pt and 1 nm of Pt. It would mean that the thicker the Pt spacer, the larger the amplitude of the low-T peak, as possibly consistent with the formation of less Pt content in interfacial clusters or with the formation of incomplete interfacial layer of Pt. The relevance of this intermediate case is

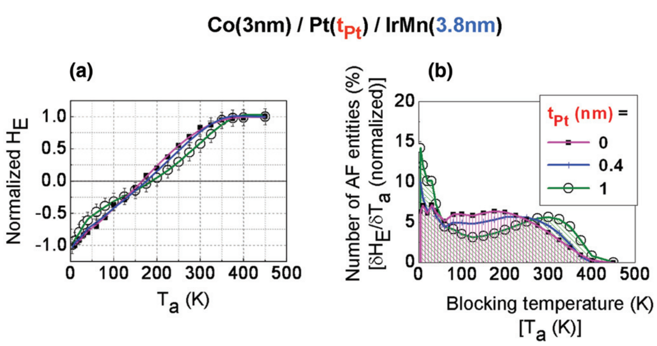

FIG. 3. (Color online) (a) Comparison of the normalized H_E vs T_a dependences as reproduced from the raw data of Fig. 2. For $t_{Pt} = 0$ and 0.4 nm the error bars along the y axis are of the same size as that of the symbol. The full lines result from smoothing of the raw data. (b) Dependence on T_a of the normalized derivative $\delta H_E/\delta T_a$ as deduced from the full lines in (a). $\delta H_E/\delta T_a$ vs T_a represents the blocking temperature distributions. The integral of any of the distributions is equal to 1.

yet hard to discuss given that most of the experimental values in Fig. 3(a) for 0 and 0.4 nm of Pt overlap when taking into account error bars.

In conclusion, for Co/IrMn bilayers, we evidenced two consequences on the T_B distributions, of the addition of a complete Pt layer as a diffusion barrier for Mn between Co and IrMn. On the one hand it improves the IrMn growth and reduces the interfacial F/AF coupling, which shifts the high-T part of the T_B distribution toward higher T. This is beneficial in terms of thermal stability of the EB characteristics for applications. Yet, on the other hand, it adds interfacial frustrations likely via the formation of CoPt alloys, which increases the amount of spin-glass entities with low freezing T rather than reducing it as expected from its role in limiting the Mn diffusion. This is on the contrary detrimental for applications. Simply adding a full Pt insertion to limit the diffusion of Mn at the Co-IrMn interface and to eradicate spin-glass like regions in Mn based F/AF bilayers does not seem to be a viable solution. For Co/IrMn based bilayers, one should turn toward a material exhibiting good wetting properties on Co, being immiscible with Co, Ir, and Mn and preventing interdiffusion of species. This will be investigated in further studies.

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