Exchange biasing by Ir$_{19}$Mn$_{81}$

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Exchange biasing by Ir$_{19}$Mn$_{81}$: Dependence on temperature, microstructure and antiferromagnetic layer thickness

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We have investigated the thermal stability of the exchange biasing interaction in antiferromagnetic/ferromagnetic bilayers of Ir$_{19}$Mn$_{81}$ and Ni$_{80}$Fe$_{20}$ or Co$_{90}$Fe$_{10}$. The exchange-biasing field and the coercive field were found to depend strongly on the thickness of the Ir$_{19}$Mn$_{81}$ layer and on the crystallographic texture of the bilayer. The exchange-biasing field at room temperature has a maximum at 4 nm Ir$_{19}$Mn$_{81}$ layer thickness and then decreases for increasing Ir$_{19}$Mn$_{81}$ layer thickness. This coincides quite well with the thickness dependence of the (111) texture measured in Ir$_{19}$Mn$_{81}$ layers. Removing the Ta seed layer resulted in the disappearance of the (111) texture and at the same time the exchange-biasing field decreased for all Ir$_{19}$Mn$_{81}$ layer thicknesses. For films with a Ta seed layer and 8 nm Ir$_{19}$Mn$_{81}$ or more, we have found a blocking temperature of 560 K, decreasing with decreasing Ir$_{19}$Mn$_{81}$ layer thickness to around room temperature for 2 nm Ir$_{19}$Mn$_{81}$. Removing the Ta seed layer resulted in a decrease of the blocking temperature for all layer thicknesses investigated. We have concluded that the (111) texture is an important factor governing the size of the exchange-biasing field and the thermal stability. © 2000 American Institute of Physics. [S0021-8979(00)01212-3]

I. INTRODUCTION

For high-temperature applications of giant-magnetoresistance exchange-biased spin valves, the thermal stability of the exchange biasing interaction is one of the limiting factors. The exchange-biasing interaction pinning the magnetization direction of one of the ferromagnetic ($F$) layers of the spin valve decreases with increasing temperature until it becomes zero at the blocking temperature $T_B$. For the “traditional” exchange-biasing materials Fe$_{50}$Mn$_{50}$ and NiO $T_B$ is too low (420 and 470 K, respectively) for high-temperature applications (typically up to 470 K). The relatively new exchange-biasing material Ir–Mn, with approximately 20 at. % Ir, was found to combine a blocking temperature above 470 K with a relatively strong exchange-biasing interaction. $^{4,5}$ In a recent article, we have reported an interfacial exchange energy up to $J_{eb} = 0.13 \text{mJ/m}^2$, combined with a blocking temperature of approximately 560 K, after deposition at room temperature and without any post-deposition annealing. $^{6}$ Our results showed that during heating there are irreversible changes in the magnetic structure of the AF/$F$ bilayer, influencing the size of the exchange-biasing interaction. There was a difference in behavior between samples with an Ir–Mn (18±1 at. % Ir) layer of 10 and 30 nm. Whether the Ir–Mn layer was deposited below or on top of the $F$ layer was also found to be important. We have also investigated previously the microstructure of the bilayers in relation to the use of Ir$_{19}$Mn$_{81}$ biasing layers in giant magnetoresistance spin valves for magnetic sensors. $^{7}$

In this article, we will report on further investigations of the influence of the Ir–Mn layer thickness on the strength of the exchange-biasing interaction and its thermal stability in Ir$_{19}$Mn$_{81}$/F bilayers with Ni$_{80}$Fe$_{20}$ or Co$_{90}$Fe$_{10}$ as the $F$ layer material. The microstructure of the bilayers is varied and is related to their exchange-biasing behavior.

It will be shown that the exchange-biasing field in the temperature range of interest depends strongly on the AF layer thickness. The same dependence is found for the degree of (111) texture in the Ir$_{19}$Mn$_{81}$ layer, which indicates that the degree of (111) texture in the Ir$_{19}$Mn$_{81}$ layer has a strong influence on the exchange-biasing interaction. Heating and cooling of different types of exchange-biased films at temperatures above room temperature shows irreversible changes in the exchange-biasing field. Since no evidence is found for atomic diffusion between the $F$ and the AF layer, we ascribe the change in $H_{eb}(T)$ to an irreversible change of the magnetic structure of the AF layer.

II. EXPERIMENTAL SETUP

Bilayers of Ni$_{80}$Fe$_{20}$ or Co$_{90}$Fe$_{10}$ and Ir$_{19}$Mn$_{81}$ were deposited onto Si(100) substrates by means of dc magnetron sputtering (base pressure $\sim 10^{-5}$ Pa). Ir$_{19}$Mn$_{81}$ was sputtered from a Mn target with Ir chips attached to it. The Ir–Mn composition has been chosen on the basis of analyzing the measured exchange-biasing field of films with Ir–Mn layers containing between 17 and 30 at. % Ir. In this composition range, a monotonic decrease of the exchange-biasing inter-
action was found with increasing Ir content. This is in agreement with what has been reported by Hoshino et al. and Fuke et al., who also find an increase of the exchange biasing field with decreasing Ir content, in the composition ranges from 20 to 40 at. % Ir and 15 to 50 at. % Ir, respectively.

The Ar pressure was typically 0.67 Pa (~5 mTorr) during deposition. All films were deposited at room temperature and were situated in a magnetic field of 20 kA/m during deposition to align the F layer, thereby inducing exchange anisotropy. Two different configurations were deposited: top: Si(100)/3.5 nm Ta/t_F nm F/t_AF nm Ir_{19}Mn_{81}/5 nm Ta, and bottom: Si(100)/3.5 nm Ta/2 nm Ni_{80}Fe_{20}/t_AF nm Ir_{19}Mn_{81}/t_F nm F/5 nm Ta.

The thicknesses of the F and AF layers (t_F and t_AF) both ranged between 2 and 30 nm. Ta was used as a seed layer to induce (111) texture in the F layer (top configuration) or the 2 nm Ni_{80}Fe_{20} buffer layer (bottom configuration). The Ni_{80}Fe_{20} buffer layer was used to promote (111) to another film deposited on top of it. Although in an earlier study no distinct relationship between the degree of (111) texture and the exchange-biasing field has been observed, it is still relevant to the application in spin valves, in which growth of a free magnetic layer with (111) texture leads to good soft magnetic properties. Also, a series of samples was deposited without the Ta seed layer to investigate the role of this layer for the exchange-biasing interaction. A Ta layer was used as a capping layer to protect the underlying layers against oxidation. In the remainder of this article, only the F layer and the Ir_{19}Mn_{81} layer will be mentioned.

The composition and the layer thicknesses of the films were determined by means of Rutherford backscattering spectroscopy. The Ir content was 19±1 at. %. Hysteresis loops were measured in a superconducting quantum interference device magnetometer, in a vibrating sample magnetometer or by means of the magneto-optical Kerr effect.

III. EXPERIMENTAL RESULTS

First, H_{eb} was determined for F=Ni_{80}Fe_{20} and F=Co_{80}Fe_{10} for both configurations. Figure 1 gives the magnetization loop of a 4 nm Ir_{19}Mn_{81}/20 nm Co_{80}Fe_{10} bottom configuration film as measured by means of the magneto-optical Kerr effect. The relevant quantities are defined in the figure. The exchange-biasing field is defined as the midpoint between the two points where the magnetization is zero at increasing and decreasing field (H_1 and H_2): H_{eb}=(H_1+H_2)/2. The coercive field is defined as H_c=(H_1-H_2)/2. From H_{eb} of the individual films the interfacial exchange energy J_{eb} was determined with the phenomenological relation

$$ J_{eb} = \mu_0 H_{eb} M_s t_F, $$

where M_s is the saturation magnetization of the F layer.

Figure 2 shows H_{eb}(T) for t_AF nm Ir_{19}Mn_{81}/20 nm Co_{80}Fe_{10} bottom configuration films with t_AF ranging between 2 and 30 nm. The as-deposited films have been slowly heated and cooled (~2 K/min) in an external field parallel to the exchange-biasing direction. The maximum temperature is approximately 460 K, except for films with 2 and 30 nm Ir_{19}Mn_{81}, which were heated to 400 and 620 K, respectively. The whole procedure took place in vacuum (p<5×10^{-3} Pa). It is shown in Fig. 2 that a 2 nm Ir_{19}Mn_{81} layer...
does not produce an exchange-biasing interaction at room temperature in the as-deposited state. However, upon cooling from 400 K, a finite $H_\text{eb}$ appears around 300 K at decreasing temperature. A film with 4 nm Ir$_{19}$Mn$_{81}$ already shows a large $H_\text{eb}$ at room temperature in the as-deposited state, even larger than what is found for samples with $t_{AF}$=6 nm. However, $H_\text{eb}$ decreases strongly with increasing temperature until it is zero at a blocking temperature of approximately 450 K. At decreasing temperature a strong increase of $H_\text{eb}$ as compared to the as-deposited sample is found. Reheating the sample with $t_{AF}$=4 nm to 460 K does not lead to any further change of $H_\text{eb}$.

$H_\text{eb}$ measured at 460 K increases with increasing Ir$_{19}$Mn$_{81}$ layer thickness until $t_{AF}$=8 nm, above which the value remains the same. This indicates an increase of the blocking temperature for increasing Ir$_{19}$Mn$_{81}$ layer thickness. Extrapolation of $H_\text{eb}(T)$ at increasing temperature for $t_{AF}$=6 and 8 nm leads to $T_B=490$ K and 550 K, respectively. Samples with 10 and 30 nm Ir$_{19}$Mn$_{81}$ have been heated to 620 K as well, and for both samples a blocking temperature of 560±20 K was found. Also, samples with 30 nm Ni$_{80}$Fe$_{20}$, instead of 20 nm Co$_{60}$Fe$_{40}$, as the F layer and $t_{AF}$=10 or 30 nm have been heated to 620 K. These samples show similar behavior upon heating and cooling as the ones with 20 nm Co$_{60}$Fe$_{40}$. The blocking temperature is equal ($T_B=560±20$ K) and the exchange-biasing field as a function of temperature normalized to the value at 300 K, $H_\text{eb}(T)/H_\text{eb}(300)$, shows the same behavior. Only $H_c$ is smaller when Ni$_{80}$Fe$_{20}$ is used as the F layer.

The measured blocking temperatures are higher than those reported previously by Devasahayam et al.,$^5$ who report $T_B=520$ K for $t_{AF}$=40 nm and a strong decrease of $T_B$ for thinner Ir–Mn layers with approximately 20 at.% Ir.

Using plan-view transmission electron microscopy (TEM), images of the grain structure could be obtained and selected area electron diffraction (SAED) was performed on films with varying Ir$_{19}$Mn$_{81}$ layer thicknesses deposited on Si$_3$N$_4$ membranes (TEM windows). The films had the layer stack: Si$_3$N$_4$/5 nm Ta/2 nm Ni$_{80}$Fe$_{20}$/t$_{AF}$ nm Ir$_{19}$Mn$_{81}$/5 nm Ta. The intensity of the (220) Ir$_{19}$Mn$_{81}$ ring with respect to the (220) Ni$_{80}$Fe$_{20}$ ring gives a quantitative measure of degree of (111) texture. This relative intensity obtained from the SAED patterns is normalized by the values for the layer thicknesses. The results are shown in Fig. 3. A more detailed description of the analysis can be found in Ref. 7. In Fig. 3 can be seen that a 5 nm Ir$_{19}$Mn$_{81}$ layer with Ta and Ni$_{80}$Fe$_{20}$ underlayers has the strongest (111) texture. The (111) texture becomes weaker for thicker Ir$_{19}$Mn$_{81}$ films. For the film with 30 nm Ir$_{19}$Mn$_{81}$, the (220) ring of Ir$_{19}$Mn$_{81}$ obscures the (220) Ni$_{80}$Fe$_{20}$ ring which obstructs the accurate determination of its relative intensity. However, our analysis of other visible diffraction rings shows that the degree of (111) texture for 30 nm Ir$_{19}$Mn$_{81}$ layer has certainly decreased further as compared to a 10 nm Ir$_{19}$Mn$_{81}$ layer. The thickness of the Ni$_{80}$Fe$_{20}$ underlayer was found to have no further influence on the texture of the Ir$_{19}$Mn$_{81}$ layer, as long as it is equal to or larger than 2 nm. During heating no change in texture was found. No evidence was found for the presence of an ordered crystallographic phase (IrMn$_3$) in the layer.$^{11}$

Using plan-view TEM we have also estimated the average grain sizes in the films discussed above. The average grain size cannot be determined with high accuracy, because of the presence of different layers which obscure the picture. However, it is clear that the average grain size increases with increasing Ir$_{19}$Mn$_{81}$ layer thickness. The largest increase is observed when going from 2 to 4 nm layer thickness. For $t_{AF}$=6 nm, no further increase has been observed and the estimated average grain size is approximately 10 nm.

The samples for which $H_\text{eb}(T)$ was measured during heating to 460 K, as shown in Fig. 2, were subsequently heated to 580 K (in air), also in a field parallel to the exchange-biasing direction. The highest temperature was maintained for not more than a few seconds before cooling down again. Figure 4 shows $H_\text{eb}$ at room temperature for the samples in the as-deposited state, and after heating to 460 and 580 K, respectively. Also included are the results for the films with 10 and 30 nm Ir$_{19}$Mn$_{81}$ that were heated to 620 K.

![Figure 3](image3.png)  
**FIG. 3.** Relative intensity of the Ir$_{19}$Mn$_{81}$ diffraction ring as a function of the Ir$_{19}$Mn$_{81}$ layer thickness for Si$_3$N$_4$/5 nm Ta/2 nm Ni$_{80}$Fe$_{20}$/t$_{AF}$ nm Ir$_{19}$Mn$_{81}$/5 nm Ta films.

![Figure 4](image4.png)  
**FIG. 4.** Exchange-biasing field at room temperature as function of Ir$_{19}$Mn$_{81}$ layer thickness for as-deposited films and after cooling down from 460, 580 or 620 K in a field parallel to the exchange-biasing direction. Cooling down from 580 K was performed in a relatively high magnetic field of 2 MA/m. The lines are only guides to the eye.
After heating, there is an increase of $H_{eb}$ for 2 nm ≤ $t_{AF}$ ≤ 8 nm and a decrease of $H_{eb}$ for $t_{AF}$ = 30 nm. For $t_{AF}$ = 10 nm, $H_{eb}$ is almost unchanged after heating to 460 K, but an increase is found after heating to 580 K. Heating to 460 and 620 K and subsequent cooling down took place in a field high enough to saturate the magnetization of the $F$ layer, approximately 15 kA/m, whereas heating to 580 K was performed in a field of 2 MA/m. The large applied field of 2 MA/m apparently has a beneficial influence on the exchange-biasing interaction. For example, for a 4 nm Ir$_{19}$Mn$_{81}$/20 nm Co$_{50}$Fe$_{50}$ film after cooling from 580 K, the calculated exchange energy, $J_{ex}$ = 0.23 mJ/m$^2$, is the highest value reported so far for Ir–Mn. Cooling from 460 K for the same film leads to $J_{ex}$ = 0.20 mJ/m$^2$. Comparing the results for a film with 10 nm Ir$_{19}$Mn$_{81}$, heating to 580 K (and cooling down) in a “high” field results in a larger exchange-biasing field than heating to 620 K in a “low” field. The positive effect of the large applied field on the exchange-biasing interaction is possibly due to a direct influence on the magnetic moments in the AF layer itself.

Figure 5 shows the coercive field $H_c$ as a function of temperature for $t_{AF}$ = 2 nm ≤ 8 nm and a decrease of $H_c$ for $t_{AF}$ = 30 nm. For $t_{AF}$ = 10 nm, $H_c$ is almost unchanged after heating to 460 K, but an increase is found after heating to 580 K. Heating to 460 and 620 K and subsequent cooling down took place in a field high enough to saturate the magnetization of the $F$ layer, approximately 15 kA/m, whereas heating to 580 K was performed in a field of 2 MA/m. The large applied field of 2 MA/m apparently has a beneficial influence on the exchange-biasing interaction. For example, for a 4 nm Ir$_{19}$Mn$_{81}$/20 nm Co$_{50}$Fe$_{50}$ film after cooling from 580 K, the calculated exchange energy, $J_{ex}$ = 0.23 mJ/m$^2$, is the highest value reported so far for Ir–Mn. Cooling from 460 K for the same film leads to $J_{ex}$ = 0.20 mJ/m$^2$. Comparing the results for a film with 10 nm Ir$_{19}$Mn$_{81}$, heating to 580 K (and cooling down) in a “high” field results in a larger exchange-biasing field than heating to 620 K in a “low” field. The positive effect of the large applied field on the exchange-biasing interaction is possibly due to a direct influence on the magnetic moments in the AF layer itself.

FIG. 5. Coercive field $H_c$ as a function of temperature for $t_{AF}$ = 2 nm Ir$_{19}$Mn$_{81}$/20 nm Co$_{50}$Fe$_{50}$ films having the bottom configuration with: (a) $t_{AF}$ = 30, 10 and 8 nm and with (b) $t_{AF}$ = 6, 4 and 2 nm. Values for increasing and decreasing temperature are shown as open and solid symbols, respectively.

$H_c(T)$ curve has a very distinct shape for samples with $t_{AF}$ ≈ 6 nm, with a maximum of $H_c$, which seems to coincide with the blocking temperature. These findings are in agreement with the results of Jungblut et al.$^{12}$ for MBE-grown Ni$_{50}$Fe$_{20}$/Fe$_{50}$Mn$_{50}$ bilayers. They find a maximum of the coercive field as a function of the Fe$_{50}$Mn$_{50}$ layer thickness around the onset of the exchange-biasing interaction. The increase of the coercivity is probably due to the occurrence of regions in the AF layer, that do have an interaction with the $F$ layer, but in which the AF spin configuration does not remain stable when the $F$ layer is reversed. This gives rise to a uniaxial anisotropy instead of a unidirectional anisotropy.

Possibly, the change of $H_{eb}$ after annealing is caused by atomic diffusion. However, from sputter Auger analysis, we have found no evidence for diffusion across interfaces.$^6$ Another indication that the change of $H_{eb}$ is of a different origin than atomic diffusion is that $H_{eb}$ was already changed after cooling from above 340 K, a temperature which is considered too low for atomic diffusion to occur in these materials.$^5$

To investigate the influence of the degree of (111) texture on the exchange-biasing field and the thermal stability, films without the Ta seed layer were deposited. The 2 nm Ni$_{50}$Fe$_{20}$ buffer layer was still present. TEM analysis of 2 nm Ni$_{50}$Fe$_{20}$/111 nm Ir$_{19}$Mn$_{81}$/5 nm Ta, deposited on Si$_3$N$_4$ membranes (TEM windows), showed that a film with 2 nm Ir$_{19}$Mn$_{81}$ does have a weak (111) texture. Films with thicker Ir$_{19}$Mn$_{81}$ layers (4 ≤ $t_{AF}$ ≤ 10 nm) are all randomly oriented. In films with a Ta seed layer present, a clear (111) texture was observed. The average grain size in the films is affected by the removal of the Ta seed layer as well. TEM images show that for films without the Ta seed layer, the average grain size increases monotonically with an increasing Ir$_{19}$Mn$_{81}$ layer thickness, up to an average grain size of approximately 5 nm for a film with 10 nm Ir$_{19}$Mn$_{81}$. For films with a Ta seed layer the average grain size shows no significant increase for $t_{AF}$ > 6 nm, the average grain size for a film with 10 nm Ir$_{19}$Mn$_{81}$ is approximately 10 nm.

Figures 6(a) and 6(b) show $H_{eb}(T)$ and $H_c(T)$ for Si(100)/2 nm Ni$_{50}$Fe$_{20}$/111 nm Ir$_{19}$Mn$_{81}$/20 nm Co$_{50}$Fe$_{50}$/5 nm Ta films without a Ta seed layer. For clarity, we have only included the curves for $t_{AF}$ = 4, 6 and 8 nm. For $t_{AF}$ = 2 nm, no exchange-biasing field is found in this temperature range and $H_c(T)$ ranges between 1.6 and 1 kA/m. Comparison with Figs. 2 and 5 reveals that the values of $H_{eb}(T)$ are much lower for films without the Ta seed layer. On the other hand $H_c$ has increased for all films, except for the film with 2 nm Ir$_{19}$Mn$_{81}$. Also the distinct shape of the $H_c(T)$ curve with a maximum around the blocking temperature is no longer found. The blocking temperatures are strongly decreased for the films without Ta seed layer. Figure 6(c) shows the exchange-biasing field at room temperature for films in the as-deposited state and after cooling down from 460 K. $H_{eb}$ after cooling from above $T_B$, is increasing monotonically with increasing Ir$_{19}$Mn$_{81}$ layer thickness up to 8 nm [see Fig. 6(c)]. A further increase to 10 nm Ir$_{19}$Mn$_{81}$ leads to a small decrease of $H_{eb}$. We note that this datapoint has a rather large experimental error (± 0.2 kA/m).
IV. DISCUSSION

A. Top/bottom configuration

The first unexpected result found for Ir$_{19}$Mn$_{81}$ is the existence of a considerable exchange-biasing field when the AF layer is deposited below the F layer (bottom configuration). Usually it is observed that the exchange-biasing interaction can only be induced directly, i.e., already during deposition, by depositing an AF layer on top of a magnetically saturated F layer. The magnetic moments of the AF atoms at the interface are then assumed to arrange such that the magnetization of the F layer is maintained by the interaction between the interfacial atoms, even after the saturating magnetic field has been removed. Obviously, when the AF layer is deposited first, the arrangement of the magnetic moments of the AF atoms at the interface will not match the proper arrangement to induce a unidirectional biasing in the F layer deposited on top. In films in which the Fe$_{50}$Mn$_{50}$ biasing layer was deposited below the ferromagnetic layer, a large increase of the exchange biasing field is obtained after field cooling from above the blocking temperature, indicating an imperfect magnetic configuration for biasing in the as-deposited films.$^{10,13}$

In Fig. 7, a comparison is made between $H_{eb}(T)$ for the top and bottom configuration of films with 10 nm Ir$_{19}$Mn$_{81}$ as the biasing layer and 30 nm Ni$_{80}$Fe$_{20}$ as the biased layer. It is shown that the top configuration has the highest $H_{eb}$ in the as-deposited state. After cooling the films from 620 K, $H_{eb}$ for the top configuration has decreased dramatically, whereas for the bottom configuration $H_{eb}$ has increased slightly, however still being lower than the as-deposited value for the top configuration. It is uncertain whether the different behavior of these two configurations is caused by magnetic or microstructural differences. It is possible that the degree of (111) texture varies over the Ir$_{19}$Mn$_{81}$ layer, which would result in a difference in (111) texture at the two interfaces of the AF layer. This is in agreement with the results shown in Fig. 3, where we found a decrease of the degree of (111) texture for $t_{AF} > 5$ nm, which leads us to conclude that the part of the Ir$_{19}$Mn$_{81}$ layer which is deposited after the first 5 nm, has a (111) texture which is less strong than in the first 5 nm and which decreases monotonically with increasing thickness. Similar behavior of $H_{eb}(T)$ has been found if Ni$_{80}$Fe$_{20}$ is replaced by Co$_{90}$Fe$_{10}$.

Apparently, the Ir$_{19}$Mn$_{81}$ layer forms such a magnetic domain structure before or during deposition of the F layer, that the exchange-biasing interaction is present directly in the as-deposited bilayer. At higher temperatures the magnetic structure in the AF layer is able to relax to a state of lower energy, which is maintained when cooling again, resulting in the different values for $H_{eb}$.

B. Microstructural influence on the exchange-biasing interaction

Regarding the influence of the (111) texture and grain size on the exchange-biasing field, conflicting reports have been given in the literature. Hoshino et al.$^8$ and Nakatani et al.$^{14}$ both ascribe an increase of the exchange-biasing field
to an increase of the (111) texture and to an increase in grain size. Devashayam et al.\textsuperscript{5} reported a decrease of the exchange-biasing field together with an increase of the (111) texture. They influenced the (111) texture by varying the sputter parameters of their rf sputter-deposition procedure. However, in our dc-magnetron sputtered films it is observed that a variation of the sputter parameters may have an influence on the Ir content of the films, which will also have an influence on the exchange-biasing interaction. Ro et al.\textsuperscript{15} conclude from experiments with different buffer layers that the grain size and not the (111) texture in the Ir–Mn layer determines the exchange-biasing interaction.

With TEM analysis the (111) texture and the average grain sizes were determined for films with and without a Ta seed layer (see Sec. III). The strength of the (111) texture as a function of the Ir\textsubscript{19}Mn\textsubscript{81} layer thickness is presented in Fig. 3. Figure 8 gives a schematic representation of the grain size in the Ir\textsubscript{19}Mn\textsubscript{81} layer compared to the exchange-biasing field at room temperature after cooling from 460 K, as a function of the Ir\textsubscript{19}Mn\textsubscript{81} layer thickness. It is shown that for films without a Ta seed layer both the exchange-biasing field and the grain size increase with increasing Ir\textsubscript{19}Mn\textsubscript{81} layer thickness. Adding the Ta seed layer results in an exchange-biasing field that is peaked around 4 nm Ir\textsubscript{19}Mn\textsubscript{81}, whereas the grain size still increases monotonically. On the other hand, a strong resemblance is found between the behavior of the (111) texture and $H_{\text{eb}}$ as a function of Ir\textsubscript{19}Mn\textsubscript{81} layer thickness. This suggests that the (111) texture is a more important factor than the grain size for the exchange-biasing interaction in the Ir\textsubscript{19}Mn\textsubscript{81} layer.

C. Exchange biasing as a function of AF layer thickness

In Fig. 9, we have shown the dependence of the exchange-biasing field on the thickness of the AF layer predicted by the model of Malozemoff.\textsuperscript{16} Within this model, the interface energy is random on the atomic scale. Exchange biasing arises when this interfacial energy is large enough, which will make it favorable for the magnetic structure of the AF layer to consist of domains, with domain walls that are perpendicular to the interface. Due to the fact that the domains have a finite size, the random field per domain area is not zero, not even for rough or compensated interfaces. At a certain thickness $t_{\text{AF,1}}$ the domain wall energy and the net interface exchange energy balance and below this thickness a domain pattern in the AF layer and a finite $H_{\text{eb}}$ will appear. For $t_{\text{AF,2}}<t_{\text{AF}}<t_{\text{AF,1}}$, the domain area at the interface will remain constant as well as the exchange-biasing field.\textsuperscript{16} For $t_{\text{AF}}<t_{\text{AF,2}}$, the domain size depends linearly on the AF layer thickness and the exchange-biasing field will increase with decreasing $t_{\text{AF}}$. There is also a lower limit to the AF layer thickness. For $t_{\text{AF}}<t_{\text{AF,3}}$, the domain structure is no longer stable and domain walls will move through the material when the $F$ layer is reversed, resulting in $H_{\text{eb}}=0$. We note that the effect of thermal fluctuations on exchange biasing is not included in this model, so it is essentially only valid in the limit to 0 K.

We can compare the predictions for $H_{\text{eb}}(t_{\text{AF}})$ with the results of our measurements as shown in Figs. 4 and 6(c). We note that these measurements were performed at room temperature instead of 0 K, which will influence the values of the exchange-biasing field for different Ir\textsubscript{19}Mn\textsubscript{81} layer thicknesses, due to the finite-size effect. However, assuming that the curves for $H_{\text{eb}}$ at decreasing temperature can be smoothly extrapolated to 0 K, one expects that the basic features of Figs. 4 and 6(c) will not change dramatically. Figure 4 shows a dependence of the exchange-biasing field at room temperature on the thickness of the AF layer which is of a similar shape as predicted by the model of Malozemoff (Fig. 9). Since we are dealing with films with the bottom configuration, it is best to consider the exchange-biasing field measured after cooling down to room temperature from above the blocking temperature. In spite of the fact that a peak is found for $H_{\text{eb}}$ as a function of AF layer thickness, as expected in the Malozemoff model, we do not regard this as firm support for this model. The peak position could, alternatively, also be explained from the fact that it coincides with the maximum degree of (111) texture. The latter point of view is supported by the experimental results for films without the Ta seed layer [see Fig. 6(c)]. The lower $H_{\text{eb}}$ in
films without a Ta seed layer is accompanied by the disappearance of the (111) texture, i.e., random orientation, in the Ir$_{19}$Mn$_{81}$ layer. This suggests that the thickness of the AF layer is not the main factor in the value of the exchange-biasing field.

**D. Blocking temperature as a function of Ir$_{19}$Mn$_{81}$ layer thickness**

Figure 10 shows the blocking temperatures as a function of the AF layer thickness of films with and without the Ta seed layer. Solid symbols give measured values, whereas open symbols indicate extrapolated values. The lines give the fits with the finite-size scaling model, assuming that $T_B(\infty) = 560$ K. The correlation length $\xi_0$ is found to be 1.2 nm for films with a Ta seed layer and 1.7 nm for films without. The shift exponent $\delta$ is 1.5 and 1.2 for films with and without a Ta seed layer, respectively.

![FIG. 10. Blocking temperature as a function of the Ir$_{19}$Mn$_{81}$ layer thickness for films with (squares) and without (circles) the Ta seed layer. Solid symbols give measured values, whereas open symbols indicate extrapolated values. The lines give the fits with the finite-size scaling model, assuming that $T_B(\infty) = 560$ K. The correlation length $\xi_0$ is found to be 1.2 nm for films with a Ta seed layer and 1.7 nm for films without. The shift exponent $\delta$ is 1.5 and 1.2 for films with and without a Ta seed layer, respectively.](image)

In the model of finite-size scaling, it is assumed that the decrease of $T_B$ is caused by the decrease of the Néel temperature $T_N$ for thinner AF layers. It is not clear whether this is really the case. Van der Zaal et al. have in fact found an increase of $T_N$ for Fe$_2$O$_3$/CoO layers, when decreasing the layer thickness.

Analyzing Fig. 2, we observe a difference in the shape of the $H_{eb}(T)$ curve at increasing and decreasing temperature. At increasing temperature the decrease of $H_{eb}$ is almost linear, whereas at decreasing temperature the $H_{eb}(T)$ curve is more convex. According to the model of Fulcomer and Charap the monotonous decrease of the exchange biasing with temperature occurs due to the fact that the AF exchange-biasing layer consists of different regions which have different blocking temperatures, which depend on the size of the regions. At increasing temperature more and more of these regions will lose their exchange-biasing interaction, which results in an overall decrease of the exchange-biasing field. The width of the distribution of region sizes will determine the shape of the $H_{eb}(T)$ curve. This would indicate that after heating the magnetic structure in the AF layers in our films is more homogeneous, and that the distribution of blocking temperatures is less wide as compared to the as-deposited situation.

**V. CONCLUSIONS**

We have fabricated exchange-biased layers with Ir$_{19}$Mn$_{81}$ as the biasing layer and Ni$_{80}$Fe$_{20}$ or Co$_{80}$Fe$_{10}$ as the ferromagnetic biased layer both in the top and bottom configuration and with or without a Ta seed layer. The highest interfacial exchange energy, $J_{eb} = 0.23$ mJ/m$^2$, has been found for a 3.5 nm Ta/2 nm Ni$_{80}$Fe$_{20}$/4 nm Ir$_{19}$Mn$_{81}$/20 nm Co$_{80}$Fe$_{10}$ film in the bottom configuration after cooling from 580 K in a high field (2 MA/m) parallel to the exchange-biasing direction. In general, taking Co$_{80}$Fe$_{10}$ as the $F$ layer gives a higher interfacial exchange-biasing energy than with Ni$_{80}$Fe$_{20}$. No difference is observed between the two types of $F$ layers concerning the thermal stability of the exchange-biasing interaction.

Ir$_{19}$Mn$_{81}$ has been found to be a remarkable exchange-biasing material in the sense that as-deposited films with the bottom configuration show already a considerable exchange-biasing field. Heating and subsequent cooling changes the exchange-biasing field both for top and bottom configuration. For films with the bottom configuration the exchange-biasing field has been found to increase for small Ir$_{19}$Mn$_{81}$ layer thicknesses (<10 nm) and to decrease for thicknesses larger than 10 nm. Subsequent annealing did not result in any further change of $H_{eb}$ as a function of temperature. Since no clear evidence has been found for the occurrence of atomic diffusion or any other change in the microstructure during...
annealing, especially at moderate temperatures (<550 K), it is concluded that the change of $H_{eb}$ is caused by a change of the magnetic (domain) structure in the Ir$_{19}$Mn$_{81}$ layer. It will also be very interesting to investigate this for films with the top configuration. Furthermore, preliminary results indicate that the exchange-biasing field might depend on the value of the external field during cooling. More experiments will therefore be very interesting. It would be very useful to deposit the films on single crystalline substrates.

Films with the bottom configuration were deposited with and without the Ta seed layer. Removal of the Ta seed layer results in a strong decrease of the (111) texture and the average grain size in the Ir$_{19}$Mn$_{81}$ layer. Analysis of the dependence of the exchange-biasing field, the average grain size and the (111) texture on the Ir$_{19}$Mn$_{81}$ layer thickness leads to the conclusion that the degree of (111) texture is the most important factor determining the exchange-biasing field. However, conflicting reports are found in the literature about the influence of grain size and (111) texture. Further investigation will therefore be very interesting. It would be very useful to deposit the films on single crystalline substrates, thereby creating a distinct crystallographic orientation at the AF/F interface.

A blocking temperature of 560±20 K has been found for films with 10 and 30 nm Ir$_{19}$Mn$_{81}$, both in the top and bottom configuration. For smaller Ir$_{19}$Mn$_{81}$ layer thicknesses $T_B$ decreases. Removing the Ta seed layer was also found to decrease the blocking temperature. In view of the high blocking temperature and the high interfacial exchange energy, it can be concluded that Ir$_{19}$Mn$_{81}$ is a good candidate for the biasing layer in spin valves used for high-temperature applications.

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