In-plane anomalies of the exchange bias field in $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ bilayers on Cu(110)

S. Riedling,[a] M. Bauer, C. Mathieu, and B. Hillebrands[b]
Fachbereich Physik and Schwerpunkt Materialwissenschaften, Universität Kaiserslautern, Erwin-Schrödinger-Str. 56, 67663 Kaiserslautern, Germany

R. Jungblut, J. Kohlhepp,[c] and A. Reinders
Philips Research Laboratories, Prof. Holstlaan 4, 5656 AA Eindhoven, The Netherlands

(Received 2 June 1998; accepted for publication 24 December 1998)

We report on the exchange bias effect as a function of the in-plane direction of the applied field in twofold symmetric, epitaxial $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ bilayers grown on Cu(110) single-crystal substrates. An enhancement of the exchange bias field, H_{eb}, up to a factor of 2 is observed if the external field is nearly, but not fully aligned perpendicular to the symmetry direction of the exchange bias field. From the measurement of the exchange bias field as a function of the in-plane angle of the applied field, the unidirectional, uniaxial and fourfold anisotropy contributions are determined with high precision. The symmetry direction of the unidirectional anisotropy switches with increasing NiFe thickness from [110] to [001]. [S0021-8979(99)01907-6]

I. INTRODUCTION

Metallic bilayer systems, consisting of a ferromagnetic (F) and an antiferromagnetic (AF) layer in contact, may show the so-called exchange bias effect, if they are deposited or cooled down from above the Néel temperature in the presence of a magnetic field. The main features are a shift of the hysteresis curve (B vs H loop) along the field axis as well as a sinusoidal torque curve in an otherwise isotropic material.[1] The phenomenon of exchange biasing, first observed in 1957 by Meiklejohn and Bean in the Co/CoO system,[2] has been under investigation since then, with only partial success in uncovering the physical origin.[3–7] It is now thought, that the appearance of exchange biasing is due to the exchange interaction between the F and the AF layer at the interface involving domains in the antiferromagnet[3,4,7] and/or statistical arguments in the case of exchange biasing between compensated layers.[5,6] However, polarized neutron reflectometry on exchange biased $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ bilayers has found no evidence for planar domain walls in the AF layer.[8]

For the $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ bilayer system, it has been reported that the exchange bias field H_{eb}, as well as the coercivity field H_c, depend on the crystal orientation, and therefore on the interface structure, but no indication for a preference of an uncompensated (110) or a compensated (111) spin orientation was observed.[9,10] The (110)-oriented $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ bilayer system consists of an uncompensated AF interface with a AF layering sequence of the moments in the atomic planes as shown in Fig. 1.[9,10] It should be noted here for the later discussion, that AF-spin components exist in all in-plane directions.

It was previously reported that the F-AF exchange coupling mechanism does not only generate the exchange bias field H_{eb}, causing a unidirectional anisotropy described by the anisotropy constant $K_p^{(1)}$, but also influences strongly all other contributing in-plane anisotropies, which are the twofold anisotropy ($K_p^{(2)}$) and the four-fold anisotropy ($K_p^{(4)}$).[11,12] This unexpected large uniaxial anisotropy contribution causes the easy axis of magnetization of the F layer to switch with increasing F-layer thickness from $[1\bar{1}0]$ to [001], i.e., by 90° near 40 Å.

We have studied the dependence of the exchange bias field on the in-plane direction of the external field in detail. We find that the exchange bias field depends in a very sensitive manner on all contributing in-plane anisotropies as well as on the direction of the external field, which, in turn, allows for a very precise determination of the anisotropy constants. Our measurements show a clear correlation of the symmetry axis of the exchange bias, i.e., the unidirectional anisotropy contribution, with the F-thickness dependent rotation of the symmetry axis of the twofold anisotropy contribution.

In order to understand the behavior of the exchange bias field as a function of the in-plane angle ϕ between the direction of magnetization and the [001] direction, we simulate the remagnetization process assuming a pure rotation of the magnetization using the free energy expression:

$$F_{ani} = +K_p^{(1)} \cos(\phi - \phi_{uni}) + K_p^{(2)} \cos^2(\phi) + K_p^{(4)} \cos^2(\phi) \sin^2(\phi). \tag{1}$$

The angle ϕ_{uni} describes the reference direction of the unidirectional anisotropy with respect to the $[1\bar{1}0]$ direction, in which the growth field is applied.

[a]Electronic mail: riedling@physik.uni-kl.de
[b]Corresponding author.
[c]Present address: Department of Physics, Eindhoven University of Technology, P.O. Box 513, 5600 MB Eindhoven, The Netherlands.

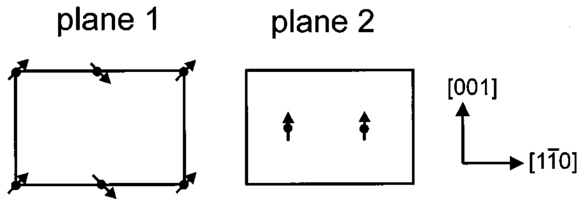

FIG. 1. Spin structure of the γ-FeMn(110) surface, according to the $\langle 111 \rangle$ model. For the (110) orientation, one has to distinguish between two different planes. The moments of plane 1 are oriented out of the (110) plane by an angle of $\pm 54.7°$ as indicated, whereas in plane 2 the moments lie in the (110) plane. Therefore the (110) plane is uncompensated.

II. EXPERIMENT

The sample was grown by molecular beam epitaxy onto a Cu(110) single-crystal substrate and consists of four staircase shaped permalloy ($Ni_{80}Fe_{20}$) layers of 18, 24, 37, and 90 Å. The preparation procedure is described elsewhere.[11] Half of the film surface is covered by a 80 Å thick antiferromagnetic $Fe_{50}Mn_{50}$ film, sufficiently thick to saturate the exchange bias effect.[9] To protect the sample against oxidation a 30 Å thick Au cap layer was deposited. During the growth of the sample a field of 250 Oe along the $[1\bar{1}0]$ direction was applied in the film plane. The structural and chemical quality of the samples was monitored using scanning tunneling microscopy, Auger spectroscopy, and low energy electron diffraction (LEED). The LEED patterns clearly indicate the (110) orientation of all films.

All hysteresis loops were measured *ex situ* at room temperature using the longitudinal magneto-optical Kerr effect (MOKE). The incident laser light (670 nm) was perpendicularly polarized to the plane of incidence by a linear polarizer and then focused onto the sample. The angle between the incident light and the plane normal of the sample was chosen to 55° in order to maximize the Kerr rotation signal. For the detection of the Kerr rotation, a differential intensity method was used.[13]

III. RESULTS

We start the discussion of our experimental investigations with the results for the uncovered NiFe layers for reference. We observe no exchange bias field and a strong uniaxial anisotropy contribution for all investigated F-layer thicknesses with the easy axis of magnetization uniformly pointing along the [001] direction. From the saturation fields of the prevailing hard directions, we determined a thickness independent twofold anisotropy constant $K_p^{(2)}$ of $(-3.6 \pm 0.5) \times 10^5\ \mathrm{erg/cm^3}$. According to scanning tunneling microscopy images the morphology of the $Ni_{80}Fe_{20}$ layer shows long, cigar shaped islands with a length-to-width ratio of about 10, lying along the $[1\bar{1}0]$ direction, which has been

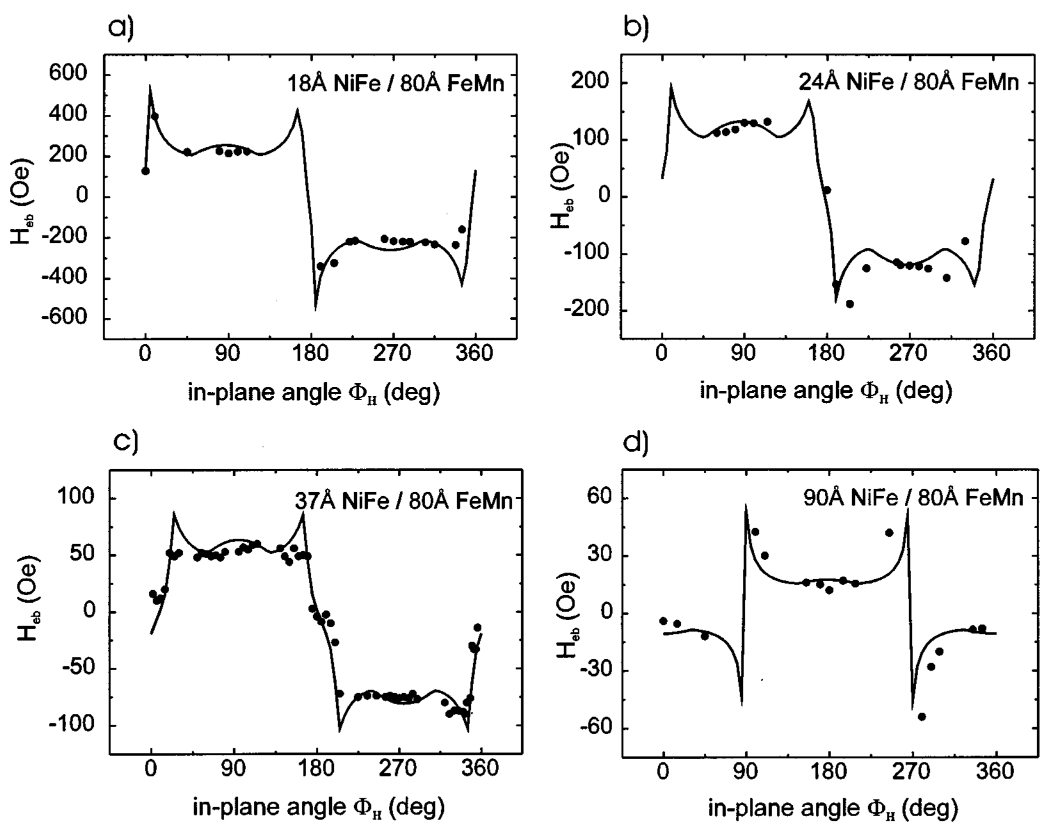

FIG. 2. Exchange bias field H_{eb} as a function of the angle of the in-plane applied field, ϕ, for the Cu(110)/$Ni_{80}Fe_{20}Fe_{50}Mn_{50}$ staircase type sample with $Ni_{80}Fe_{20}$ layer thicknesses of (a) 18, (b) 24, (c) 37, and (d) 90 Å. The full lines show fit to the data based on Eq. (1) with the anisotropy constants as the fit parameters.

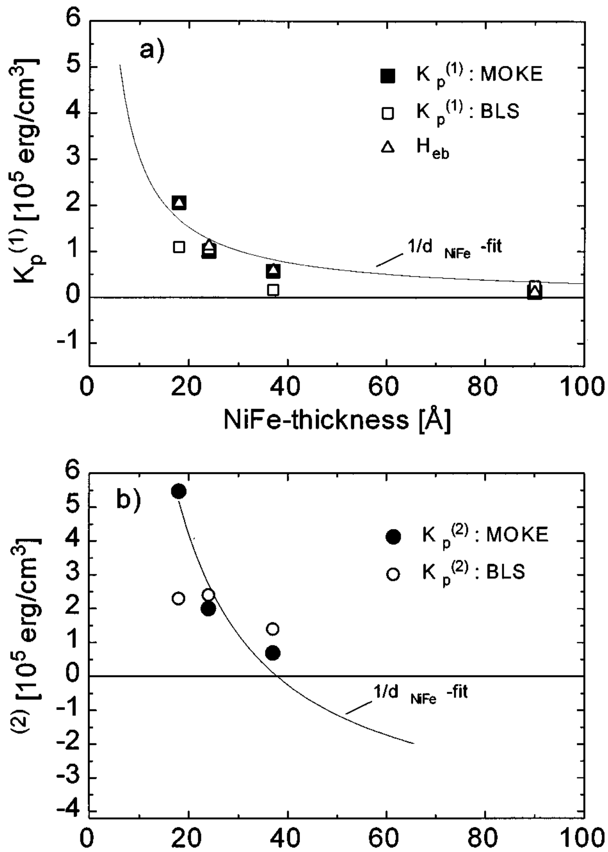

FIG. 3. (a) Obtained unidirectional and (b) uniaxial anisotropy constants of the staircase type wedge sample as a function of the $Ni_{80}Fe_{20}$ layer thickness compared to anisotropy constants obtained by Brillouin light scattering measurements. The full lines represent $1/d_{\mathrm{NiFe}}$ fits. Note the switching of the uniaxial easy axis equivalent to the change of sign of $K_p^{(2)}$ at the F-layer thickness of 40 Å.

identified as the magnetically hard direction.[14] This growth mode indicates that the observed strong uniaxial [001] behavior is likely to be of magnetoelastic origin. The value obtained for $K_p^{(2)}$ of about -3.6×10^5 erg/cm^3 corresponds to a saturation field H_s of 1 kOe along the magnetically hard direction. Thus the applied growth field of 250 Oe is insufficiently strong to saturate the magnetization of the F layer for $d_{\mathrm{NiFe}}=90$Å along the $[1\bar{1}0]$ direction, and the symmetry direction of the exchange bias field is not collinear with the direction of the growth field. This is an important fact to understand the thickness dependence of the unidirectional anisotropy in the AF-covered layers which will be discussed in the following part.

In Figs. 2(a)–2(d) the measured exchange bias field H_{eb} is plotted as a function of the in-plane angle of the applied field for all four F-layer thicknesses. It is evident, that the angular dependence of H_{eb} is very distinct from a $\sin(\phi)$ behavior of an otherwise isotropic film. A behavior similar to the latter case has been reported by Ambrose *et al.* in the NiFe/CoO exchange biased system.[15] Near the hard axis of the resulting twofold anisotropy, where H_{eb} switches sign, an enhancement of H_{eb} is observed. By fitting Eq. (1) to the data, very precise values of all in-plane anisotropy constants can be obtained. Figure 2 shows the result of the fit by full lines.

Inspecting Fig. 2 clear evidence is found that the symmetry direction of the exchange bias field switches from $[1\bar{1}0]$ for the samples with the F-layer thickness between 18 and 37 Å to [001] for the 90 Å thick F layer. For the latter film thickness, the exchange bias field points into the [001] direction, which is perpendicular to the direction of the applied field during growth. Whether this change of direction is a slow rotation or a switching could not indisputably be concluded from the available experimental data.

In Fig. 3 the obtained unidirectional and uniaxial anisotropy constants $K_p^{(1)}$ and $K_p^{(2)}$ as well as the exchange bias field, measured along the prevailing uniaxial easy axis, are plotted as a function of the F-layer thickness in comparison with data determined by Brillouin light scattering (BLS). For both anisotropy constants, an inverse thickness dependence can be verified, which is not affected by the rotation of the symmetry direction of the unidirectional anisotropy within the error margins.

For the uniaxial anisotropy [Fig. 3(b)], a thickness dependent contribution favouring the $[1\bar{1}0]$ direction, introduced by the AF layer, was observed. This contribution, in competition with the thickness independent twofold anisotropy of the uncovered permalloy layers causes the observed switching of the uniaxial easy axis equivalent to the change of sign in $K_p^{(2)}$. From a $1/d_{\mathrm{NiFe}}$ fit, the critical thickness for the switching of the uniaxial anisotropy from $[1\bar{1}0]$ to [001] is estimated to 40 Å. We have observed that the rotation of the twofold axis in the F-AF system does not only take place in samples grown with an applied field along the $[1\bar{1}0]$ direction but also in samples grown in a field along the [001] direction.[10] For an explanation of this behavior we first note, that for $d_{\mathrm{NiFe}}>40$ Å the uniaxial easy axis is perpendicular to the direction of the field applied during growth. Assuming that the interfacial spins in the AF layer are frozen in directions parallel and antiparallel to the internal field during growth, a frustration effect between the moments in the F layer, locally exchange coupled to the AF spins, occurs, resulting in a 90° orientation of the magnetization. Note that this switching of the magnetization of the F layer to a direction perpendicular to the direction of the growth field is the opposite of the perpendicular coupling as discussed by Koon,[7] where the AF moments switch to the applied field direction during sample preparation, as was recently found in the Fe_3O_4/CoO system by neutron diffraction.[16,17] A more explicit description of the mechanism described here, based on Slonczewski's fluctuation mechanism[18] for biquadratic exchange coupling, is given by Dekker and Ramstöck.[14]

IV. DISCUSSION

For the interpretation of all experimental data, we will sketch a scenario which will provide an understanding of the observed salient features, based on the growth properties of the AF layer. During growth of the AF layer two critical thicknesses can be considered. The first is the minimum thickness d_B to establish local exchange coupling between the F and AF layer (corresponding to the blocking temperature on the temperature scale). If the AF-layer thickness is larger than d_B, local F-AF exchange coupling together with

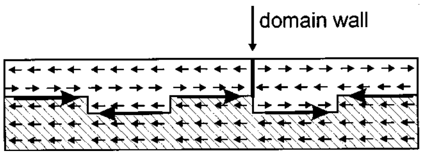

FIG. 4. Microscopic model of a F-AF uncompensated interface. A spatially varying interface exchange interaction leads to a frustration of the ferromagnetic magnetization and therefore induces a unidirectional anisotropy.

the frustration mechanism described by Dekker *et al.*[14] will provide for a mechanism to generate the interface contribution to $K_p^{(2)}$, as described above. The easy axis of this contribution is perpendicular to the easy axis of the original anisotropy of the F layer resulting in the observed reorientation of the direction of magnetization near $d_{\mathrm{NiFe}}=40$ Å. The second critical AF-layer thickness d_N, which is larger than d_B, is defined by the onset of macroscopic antiferromagnetic order in the AF layer, evidenced by the appearance of the exchange bias effect.[10] The symmetry direction of the corresponding exchange anisotropy is determined by the direction of the internal field. As we mentioned before, in the case of the 90 Å thick $Ni_{80}Fe_{20}$ layer, the applied growth field was insufficient to turn the magnetization into the $[1\bar{1}0]$ direction, which is the hard direction for this sample. Consequently, the symmetry axis of the exchange bias effect shows an in-plane rotation near the same F-layer thickness, where the direction of magnetization, i.e., of the internal field undergoes the in-plane rotation.

Although the crystallographic symmetry is two-fold, and although all AF spins experience a strong local twofold anisotropy, it is interesting to note, that both the in-plane [001] and the $[1\bar{1}0]$ axes may provide the easy directions of the unidirectional anisotropy, depending on the F-layer thickness. We assume that during the AF layer growth, domain walls are generated in the AF layer upon AF ordering of the layer, which are frozen when the layer thickness exceeds d_N. Atomic steps at the F-AF interface due to existing interface roughness provide for efficient pinning centers for the domain walls. This is schematically shown in Fig. 4. The domain walls have a magnetic dipolar moment at the atomic steps, which interact with the external field during the AF-layer growth. The Zeeman energy causes a dominant generation of domain walls in the lower energy state until the AF-layer thickness exceeds d_N and the domain walls are frozen. The magnetic dipole moments of the domain walls generate the exchange bias mechanism.

V. CONCLUSION

We have shown the angular dependence of the exchange bias field in the $Ni_{80}Fe_{20}/Fe_{50}Mn_{50}$ system. A switching of the unidirectional anisotropy to a direction perpendicular to the direction of the growth field has been observed at a F-layer thickness of about 40 Å, which could be attributed to the growth field strength of 250 Oe. Further work is needed to develop a full model of the exchange bias effect, in particular to clarify the real spin structure at the interface, including possible canting effects.

ACKNOWLEDGMENT

The authors would like to thank the Deutsche Forschungsgemeinschaft for financial support.

[1] O. Massenet, R. Montmory, and L. Néel, IEEE Trans. Magn. **1**, 63 (1965); W. H. Meiklejohn, J. Appl. Phys. **33**, 1328 (1962).
[2] W. H. Meiklejohn and C. P. Bean, Phys. Rev. **102**, 1413 (1956); **105**, 904 (1957).
[3] D. Mauri, H. C. Siegmann, P. S. Bagus, and E. Kay, J. Appl. Phys. **62**, 3047 (1987).
[4] J. S. Kouvel and L. Néel, J. Phys. Chem. Solids **21**, 57 (1961); **24**, 795 (1963).
[5] A. P. Malozemoff, Phys. Rev. B **35**, 3679 (1987); **37**, 7673 (1988).
[6] A. P. Malozemoff, J. Appl. Phys. **63**, 3874 (1988).
[7] N. C. Koon, Phys. Rev. Lett. **78**, 4865 (1997).
[8] A. R. Ball, H. Fredrikze, P. J. van der Zaag, R. Jungblut, A. Reinders, A. van der Graaf, and M. Th. Rekveldt, J. Magn. Magn. Mater. **148**, 46 (1995).
[9] R. Jungblut, R. Coehoorn, M. T. Johnson, J. aan de Stegge, and A. Reinders, J. Appl. Phys. **75**, 6659 (1994).
[10] R. Jungblut, R. Coehoorn, M. T. Johnson, Ch. Sauer, P. J. van der Zaag, A. R. Ball, Th. G. S. M. Rijks, J. aan de Stegge, and A. Reinders, J. Magn. Magn. Mater. **148**, 300 (1995).
[11] C. Mathieu, M. Bauer, B. Hillebrands, J. Fassbender, G. Güntherodt, R. Jungblut, J. Kohlhepp, and A. Reinders, J. Appl. Phys. **83**, 2863 (1998).
[12] R. Jungblut *et al.* (unpublished).
[13] Q. Fan, X. Chen, H. Chen, and Y. Zhang, J. Magn. Magn. Mater. **145**, 398 (1995).
[14] M. J. Dekker and K. Ramstöck, J. Appl. Phys. **82**, 5633 (1997).
[15] T. Ambrose, R. L. Sommer, and C. L. Chien, Phys. Rev. B **56**, 83 (1997).
[16] P. J. van der Zaag, A. R. Ball, L. F. Feiner, R. M. Wolf, and P. A. A. van der Heijden, J. Appl. Phys. **79**, 5103 (1996).
[17] Y. Ijiri, J. A. Borchers, R. W. Ewin, S. H. Lee, P. J. van der Zaag, and R. M. Wolf, Phys. Rev. **80**, 608 (1998).
[18] J. C. Slonczewski, Phys. Rev. Lett. **67**, 3172 (1991); B. Dieny and A. Vedyayev, Europhys. Lett. **25**, 723 (1994).