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Deposited 07/16/2019

Citation of published version:

Wu, S., Mankey, G., Huang, F., Willis, R. (1994): Spin Reorientation Transition in Ni Films on Cu(100). *Journal of Applied Physics*, 76(10).

DOI: https://doi.org/10.1063/1.358240

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Journal of
Applied Physics

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Cite as: Journal of Applied Physics **76**, 6434 (1994); https://doi.org/10.1063/1.358240
Published Online: 04 June 1998

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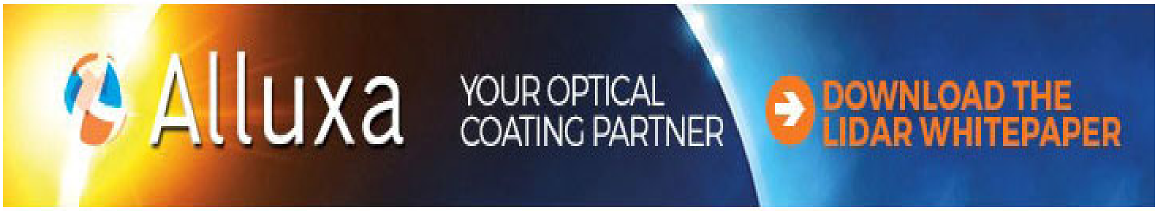

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Spin reorientation transition in Ni films on Cu(100)

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The magnetic anisotropy of Ni films grown on single-crystal Cu(100) was studied *in situ* using the surface magneto-optic Kerr effect. The easy axis of magnetization lies in the plane of the film for ultrathin films and it is perpendicular to the film above a switching thickness. This behavior is attributed to a specific contribution to the magnetocrystalline anisotropy energy induced by a change in the film microstructure above a critical thickness. In the Ni/Cu(100) system, the magnetoelastic interface anisotropy favors perpendicular magnetization which becomes comparable to the shape anisotropy at the switching thickness. We compare the switching thickness and magnetization of films grown using different processing conditions.

For ultrathin ferromagnetic films, there is a magnetization reorientation phase transition where the easy axis of magnetization is perpendicular to the film surface and switches to in plane as film thickness is increased.[1,2] Within the framework of the magnetic Hamiltonian, this transition can be interpreted as the increasing dominance of the thickness-dependent shape anisotropy which favors in-plane magnetization overcoming the spin-orbit magnetocrystalline anisotropy term which favors perpendicular magnetization. In addition to a thickness-dependent transition, a reorientation transition can also occur as a result of varying temperature of a film with fixed thickness. The temperature dependence can be modeled with an empirical temperature-dependent coefficient for the spin-orbit term in the Hamiltonian.[3] This transition has been observed for Fe/Ag(100),[4] Fe/Cu(100),[5] and Co/Au(111).[6] For the Ni films studied in this work, there is a different behavior.

Bulk Ni is the canonical Heisenberg ferromagnet. It has a spin moment of 0.6 bohr magnetons,[7] a Curie temperature T_C of 627 K,[8] a critical exponent $\beta\sim0.4$, and a weak magnetocrystalline anisotropy with the easy axis of magnetization in the [111] direction and the hard axis in the [100] direction.[9] Ni monolayers have been studied theoretically with the self-consistent local orbital method[10,11] and a perturbative tight-binding approach.[12] These calculations predict that a (100) oriented Ni monolayer has a preferred direction of magnetization in the plane of the film. More recent state-tracking first principals calculations demonstrate the magnetocrystalline anisotropy energy for monolayers is dependent on band filling and strain.[13,14] To what extent the results for truly two-dimensional monolayers can be used for describing the magnetic properties of films of finite thickness has yet to be determined. The critical behavior of Ni films on Cu(100) has been studied[15] and it was shown that the magnetic phase transition power law exponent crosses over from a two-dimensional *XY* model behavior to a three-dimensional Heisenberg model behavior at a film thickness of 7 monolayers (ML). These measurements show films thicker than a single monolayer exhibit two-dimensional magnetic properties.

Both Ni and Cu are fcc crystals with lattice constants of 3.52 and 3.61 Å, respectively, so Ni has a lattice constant 2.5% smaller than Cu. Ni films maintain an in-plane lattice constant identical to Cu up to a critical thickness of 10 ML.[16,17] The surface free energies for Ni and Cu are 2.45 and 1.85 J/m^2, respectively.[18,19] Since Cu has a surface free energy 0.6 J/m^2 lower than Ni, it tends to segregate to the surface at elevated temperatures.[20] In addition, Ni and Cu form a continuous series of solid solutions. In order to produce atomically smooth pseudomorphic layers of Ni on Cu, the films must be grown at a temperature high enough for the incoming Ni to form smooth layers yet low enough to suppress bulk diffusion which favors both alloy formation and surface segregation. The formation of a surface alloy can be detected by a reduced magnetization since $\mathrm{Ni}_x\mathrm{Cu}_{1-x}$ alloys are nonmagnetic when x is below 0.4.[21]

Our sample preparation procedures and film thickness calibration techniques have been reported before.[22] The magnetic behavior of films produced with three different processing conditions were studied in this work: Ni films grown at 300 K on a substrate roughened by 500 eV Ar$^+$ bombardment, Ni films grown at 300 K on a smooth substrate, and Ni films grown at 400 K on a smooth substrate. The composition of the samples was determined by Auger electron spectroscopy, with all of the films showing a small (<10%) but measurable amount of carbon and oxygen contamination. The magnetic properties were measured *in situ* by the surface magneto-optic Kerr effect (SMOKE) with the substrate cooled to 200 K.

To perform the SMOKE measurements, a linearly polarized He-Ne laser was incident on the sample surface at 70° from the surface normal with the polarization vector in the incident plane. The reflected light is analyzed by a Wollaston prism in combination with two photodiodes which allows simultaneous detection of the two orthogonal light components. The laser was rotated to ~0.08° from extinction in the null channel. By measuring ΔI, the difference between the Kerr intensity at remanence, the Kerr ellipticity can be derived: $\phi''=\delta/4\Delta I/I$, where I is the Kerr intensity at zero net magnetization and δ is the angle the laser is rotated from extinction. A four-pole electromagnet applies the external magnetic field either parallel or perpendicular to the film plane with a maximum magnetic field of 150 Oe.[23] This arrangement allows the measurement of both the longitudinal and polar Kerr Effects.[24] A typical hysteresis loop is shown in Fig. 1, the height of the loop at zero external field is proportional to the remanent magnetization M_r and the external magnetic field at zero net magnetization is called the coercive field or coercivity H_c.

To study the magnetization of the films, we measured the

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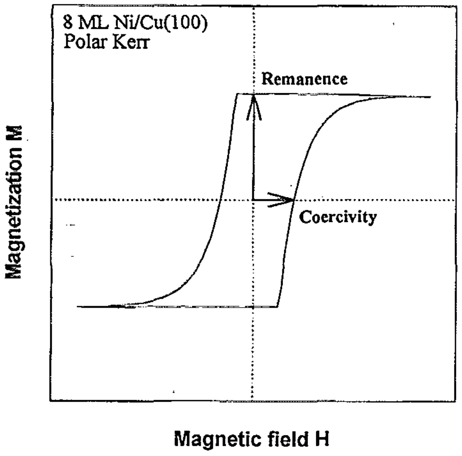

FIG. 1. A hysteresis loop for the out-of-plane magnetization of 8 ML Ni on smooth Cu(100) substrate grown at T_g=400 K. The magnetization at zero external field is the remanence. The external field at zero net magnetization is called the coercive field or coercivity.

remanence and coercivity as functions of Ni film thickness from 4 to 20 ML with the three different processing conditions. The results are shown in Figs. 2 and 3.

First we will examine the data for Ni films grown on a smooth Cu surface at the optimum growth temperature of 400 K. For film thicknesses below 8 ML, there are relatively weak Kerr ellipticities in both the parallel and perpendicular directions (Fig. 2). The magnitudes of the ellipticities cannot be directly compared because the polar and longitudinal Kerr effects have different sensitivities. The sensitivity for the polar Kerr effect is roughly ten times that of the longitudinal Kerr effect for our experimental setup, so the magnetization in the parallel direction is about ten times larger than the magnetization in the perpendicular direction. This is consistent with ferromagnetic resonance studies of Ni films on Cu(100) which showed the anisotropy energy favors in-plane magnetization below 7 ML.[25] The magnitudes of both components are comparable to those of films grown at 300 K indicating there is not a significant amount of alloying with the Cu substrate since this would result in a lower Kerr ellipticity for the 400 K growth temperature films.

The most striking difference between the films grown at 300 and 400 K on a smooth substrate is the behavior of the coercive field at low thickness (Fig. 3). The coercive field stays nearly constant for 400 K growth temperature films where it increases monotonically for the 300 K growth temperature films. This shows the films have different domain structure with the 400 K films exhibiting a behavior characteristic of single domain films and the monotonically increasing coercivity for the 300 K growth temperature films characteristic or multidomain structure with an increasing domain size as the film thickness is increased.

At 8 ML film thickness, there is an abrupt increase in both magnetizations. This abrupt increase in magnetizations is accompanied by a sudden change in the coercive fields as well. The coercive field in the perpendicular direction drops to half its initial value indicating that the easy axis of magnetization is now in the perpendicular direction. The coercive field in the parallel direction increases sharply above the switching thickness, ultimately going above the maximum field attainable with our experimental apparatus. The slope of the perpendicular coercive field versus film thickness above the switching thickness is 1.2 Oe/ML. The abrupt change in the easy axis of magnetization is characteristic of a magnetic reorientation phase transition where the easy axis of magnetization changes from in plane to perpendicular at the switching thickness of 8 ML. This effect has been observed with x-ray magnetic circular dichroism measurements at a film thickness of 10 ML.[26]

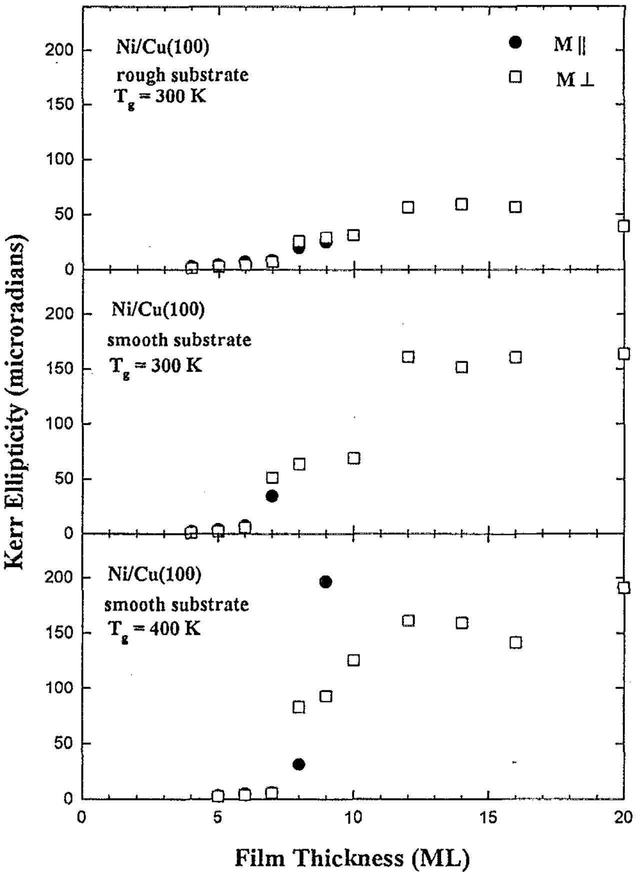

FIG. 2. Remanence vs Ni film thickness for three different film/substrate systems measured at temperature T=200 K. For each of the plots there is an abrupt increase in both parallel and perpendicular magnetizations at approximately 8 ML. Because of the limitation of our maximum applied magnetic field, the magnetization of the films could not be saturated in the parallel direction above 10 ML. This implies that the parallel component could also be increasing above this thickness.

We do not attribute the magnetization reorientation phase transition to any gross structural change since the LEED spots are similar both below and above the switching thickness for films grown at 400 K. However a subtle structural change occurs at thicknesses greater than 10 ML where the Ni(100) film transforms from a strained face centered tetragonal structure at low thickness with an in-plane lattice constant of the Cu substrate to a relaxed face centered cubic structure with an in-plane lattice constant identical to bulk Ni.[16] Small differences in the thickness where the structural strain relaxation change occurs may be due to different film preparation procedures or different levels of carbon and oxy-

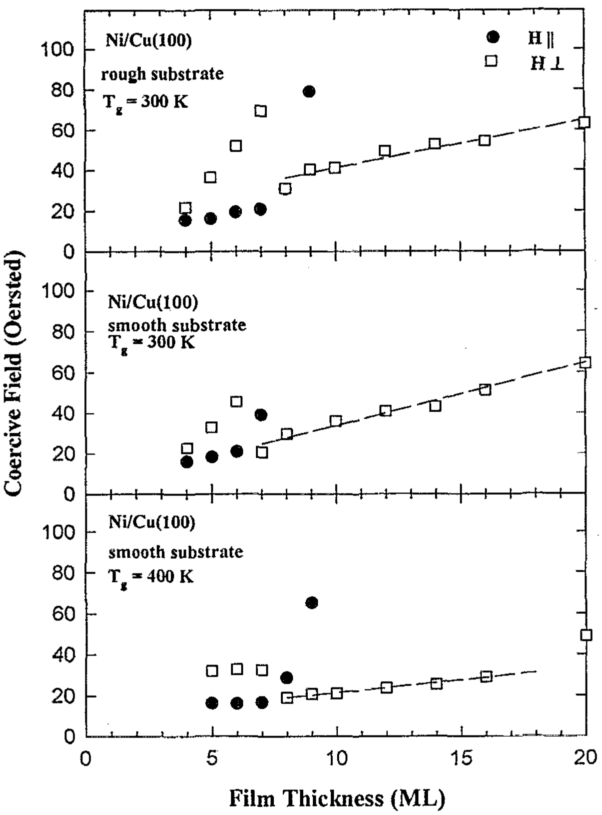

FIG. 3. Coercivity vs Ni film thickness for three different film/substrate systems measured at temperature T=200 K. For each of the systems, the coercivity in the perpendicular direction drops at the switching thickness indicating that the easy axis of magnetization is now in the perpendicular direction. The coercive field in the parallel direction increases sharply above the switching thickness, ultimately going above the maximum field attainable with our experimental apparatus.

gen contamination. This slight structural relaxation change nevertheless affects the anisotropy through the spin-orbit interaction It is sensitive to small changes in interatomic nearest-neighbor distances and by the different symmetry properties of the overlapping wave functions for the two structures. A recent calculation of the magnetocrystalline anisotropy energy showed that as a general trend, decreasing the atomic spacing will increase the energy term which prefers a perpendicular magnetization.[13,14]

In order to study the effect of interfacial strain on the switching thickness, we performed similar measurements for films deposited on a rough substrate. Interfacial roughness increases the amount of vertical strain and decreases the uniaxial anisotropy in films with perpendicular spontaneous magnetization.[27,28] The Kerr intensity for the films deposited on the rough substrate is significantly lower than that of the films deposited on a smooth substrate which shows that the magnetization of the rough films is reduced. Surprisingly, the switching thickness remains the same at 8 ML film thickness where there is a rapid decrease in the polar coercive field. However the polar Kerr ellipticity is three times weaker and the coercive field is twice as large as the films grown at 400 K. This reflects the different domain structure of the rough films and suggests that the structural change still occurs at 8 ML film thickness.

All these measurements show that the Ni films grown on Cu(100) have a predominantly in-plane magnetization for films below 8 ML thickness. Above a switching thickness the magnetization develops a strong out-of-plane component. We attribute this anomalous behavior to both structural and magnetic micromorphology of the Ni thin films. Ultrathin films of Ni have a multidomain magnetic microstructure which is a combination of perpendicular and in-plane domains, with the in-plane domains dominating below 8 ML. At a critical film thickness there is a structural transition and a perpendicular magnetization develops. The surface roughness can affect the magnetic strain anisotropy in a complicated fashion because the Ni layers can expand in both the in-plane and perpendicular direction. In the thinnest films, an in-plane interfacial strain anisotropy component is dominant. However, with increasing thickness, an out-of-plane component develops. The measurements indicate a mix of in-plane and out-of-plane magnetization domains, the latter shows a sudden increase at an onset thickness of 8 ML. At much greater film thickness, Ni/Cu(100) films show a transition back to in-plane magnetization.[29]

Funding for this research was provided by Grant No. NSF-DMR 9121736.

[1] D. Pescia and V. L. Pokrovsky, Phys. Rev. Lett. **65**, 2599 (1990).
[2] A. Kashuba and V. L. Pokrovsky, Phys. Rev. Lett. **70**, 3155 (1993).
[3] D. K. Morr, P. J. Jensen, and K. -H. Bennemann, Surf. Sci. **307–309**, 1109 (1994).
[4] Z. Q. Qiu, J. Pearson, and S. D. Bader, Phys. Rev. Lett. **70**, 1006 (1993).
[5] D. P. Pappas, C. R. Brundle, and H. Hopster, Phys. Rev. B **45**, 8169 (1992).
[6] C. Chappert, D. Renard, P. Beauvillain, J. P. Renard, and J. Seiden, J. Magn. Magn. Mater. **54–57**, 795 (1986).
[7] O. Eriksson, B. Johansson, R. C. Albers, A. M. Boring, and M. S. S. Brooks, Phys. Rev. B **42**, 2707 (1990).
[8] J. M. Leger, C. Loriers-Susse, and B. Vodar, Phys. Rev. B **6**, 4250 (1972).
[9] F. Brailsford, *Physical Principles of Magnetism*, edited by F. Brailsford (Van Nostrand, London, 1966).
[10] J. G. Gay and R. Richter, J. Appl. Phys. **61**, 3362 (1987).
[11] J. G. Gay and R. Richter, Phys. Rev. Lett. **56**, 2728 (1986).
[12] P. Bruno, Phys. Rev. B **39**, 865 (1989).
[13] D.-S. Wang, R. Wu, and A. J. Freeman, Phys. Rev. Lett. **70**, 869 (1993).
[14] D.-S. Wang, R. Wu, and A. J. Freeman, Phys. Rev. B **47**, 14932 (1993).
[15] F. Huang, M. T. Kief, G. J. Mankey, and R. F. Willis, Phys. Rev. B **49**, 3962 (1994).
[16] J. Zhang, Z. L. Han, S. Varma, and B. P. Tonner, Surf. Sci. **298**, 351 (1993).
[17] J. W. Matthews and J. L. Crawford, Thin Solid Films **5**, 187 (1970).
[18] A. R. Miedema and J. W. F. Dorleijn, Surf. Sci. **95**, 447 (1980).
[19] F. R. de Boer, R. Boom, W. C. M. Mattens, A. R. Miedema, and A. K. Niessen, *Cohesion in Metals*, edited by F. R. de Boer and D. G. Perrifor (North-Holland, Amsterdam, 1988), Vol. 1, p. 676.
[20] P. M. Ossi, Surf. Sci. **201**, L519 (1988).
[21] M. Hansen and K. Anderko, *Constitution of Binary Alloys* (McGraw-Hill, New York, 1985).
[22] F. Huang, G. J. Mankey, and R. F. Willis, Surf. Sci. Lett. **297**, L79 (1993).
[23] M. T. Kief, Ph.D. thesis, Pennsylvania State University, 1991.
[24] S. D. Bader, J. Magn. Magn. Mater. **100**, 440 (1991).
[25] B. Schulz, R. Schwarzald, and K. Baberschke, Surf. Sci. **307–309**, 1102 (1994).
[26] W. L. O'Brien and B. P. Tonner, Phys. Rev. B **49**, 15370 (1994).
[27] P. Bruno and J.-P. Renard, Appl. Phys. A **49**, 499 (1989).
[28] A. Yamaguchi, S. Ogu, and R. Yamamoto, J. Magn. Magn. Mater. **126**, 272 (1993).
[29] G. Bochi, C. A. Ballentine, H. E. Inglefield, S. S. Bogomolov, C. V. Thompson, and R. C. O'Handley, J. Appl. Phys. **75**, 6430 (1994).