Influence of growth temperature on the spin reorientation of Ni/Cu(100) ultrathin films

M. Zheng, J. Shen,[a] P. Ohresser, Ch. V. Mohan, M. Klaua, J. Barthel,[b] and J. Kirschner
Max-Planck-Institute für Mikrostrukturphysik, Weinberg 2, 06120 Halle, Germany

Ni/Cu(100) films were prepared by thermal deposition at room temperature (RT) and 170 K low temperature (LT) separately to study the influence of substrate temperature on the spin reorientation. The critical thickness of the LT grown films is observed to be about 1 ML smaller than that of the RT films. Though both types of films show similar tetragonal distortion and chemical composition, their morphology differs dramatically: the island density of the LT films is significantly higher than that of the RT films. We use this to interpret the different magnetic behavior between the RT and LT films. [S0021-8979(99)64708-9]

I. INTRODUCTION

The magnetic properties of ultrathin films have attracted a great deal of interest these years because of their interesting scientific phenomena as well as potential technological applications. One of the key properties is the magnetic anisotropy, which has been studied widely both in experimental and theoretical articles. Studies of Ni/Cu(100) thin films have been carried out in recent years[1–9] because this system shows unusual magnetic properties: Ni/Cu(100) films exhibit two spin reorientation transitions with increasing Ni film thickness. The first transition is from in-plane to perpendicular magnetization at Ni thickness of about 7–10 monolayers (ML)[1,3] and the second one is from the perpendicular back to in-plane when the Ni film thickness is above 40 ML.[5] The first transition was generally discussed in terms of the competition between a positive strain-induced volume anisotropy (favors perpendicular direction) and a negative surface anisotropy (favors in-plane direction). As the film thickness exceeds the critical thickness for coherent growth (about 13 ML), the gradual relaxation of Ni/Cu(100) lattice mismatch strain diminishes the magnitude of the magnetoelastic anisotropy energy and finally results in the second transition from perpendicular to in-plane.[5]

To verify the relative roles of the strain-induced volume anisotropy and surface anisotropy in the spin reorientation, there are two possible experimental approaches: One may modify the strain-induced volume anisotropy by modifying the strain of the film, e.g., growing Ni films on substrate with larger lattice constant like $Cu_3Au(100)$[10] or smaller lattice constant like $Cu_{1-x}Ni_x$;[11] or an alternative way is to modify the surface anisotropy by changing the surface morphology. In this work we report the modification of surface morphology by low temperature deposition. We will show the morphology effect on the spin reorientation of Ni/Cu(100) films.

II. EXPERIMENT

All experiments were performed in an ultrahigh vacuum (UHV) multichamber system, which includes scanning tunneling microscopy (STM), magneto-optical Kerr effect (MOKE), and an analysis chamber equipped with Auger electron spectroscopy (AES), low-energy electron diffraction (LEED) and facilities for thin film growth. The base pressure of the individual chambers is better than 6×10^{-11} mbar. A polished Cu(100) single crystal was used as a substrate with a miscut less than 0.2°. After cycles of cleaning by Ar^+ sputtering and annealing at 870 K, a clean and atomically flat substrate was achieved. The contamination level was below the detection limit of the Auger system. The evaporation source was a thin Ni wire (99.99%) heated by *e*-beam bombardment. The low temperature (LT) films were prepared at the substrate temperature of 170 K, followed by a subsequent annealing to room temperature. Room temperature (RT) as-grown films have also been prepared for comparison.

After the film preparation, magnetic data of both kinds of films were recorded *in situ* by MOKE measurement. The LEED and *I/V*-LEED measurements were taken after magnetic characterization. Room temperature STM has also been used to examine the morphology of the films.

III. RESULTS AND DISCUSSIONS

Up to 4 ML, both kinds of films show a good layer-by-layer growth, which is consistent with our previous result.[12] Above 4 ML, the islands of both RT and LT films tend to exhibit rectangular shape with island edges along [011] and $[0\bar{1}\bar{1}]$ directions. Figure 1 shows the typical STM images for both types of Ni films at a thickness of 8.5 ML. The morphology of films of other thicknesses is almost the same except that the size of the islands is somewhat smaller (within 30%) when the films are thinner. The height of the islands of both films is shown by the cross sections of the marked lines in the upper left corners of Fig. 1. The average island size of the RT films is almost 15 times bigger than LT films, i.e., the LT-grown Ni films have a higher density of islands than the RT-grown films. The ratio $R_{\mathrm{step/surface}}$ between the number (area) of the step edge atoms and the number (area) of the atoms in the surface plane can be estimated directly from the STM images. The former is virtually the size of an atomic chain along the island edges with a length which equals the sum of the circumferences of all the ex-

[a]Present address: Solid State Division, Oak Ridge National Lab, TN 37831-6057; Electronic mail: Shenj@ornl.gov
[b]Electronic mail: joba@secundus.mpi-halle.mpg.de

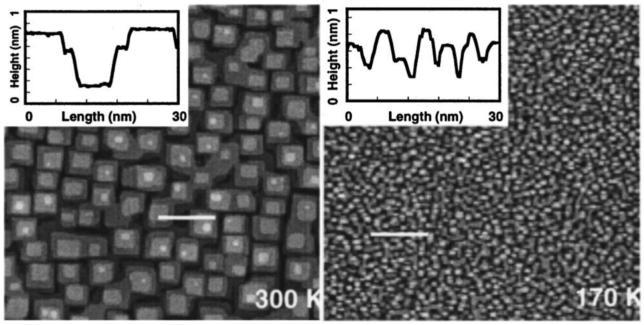

FIG. 1. STM images of 8.5 ML Ni/Cu(100) films grown at room-temperature and 170 K. The insets show the line profiles along the lines marked in the figures. The image size is 150 nm×150 nm.

posed islands in the scanning area, while the latter is the size of the scanning area. The calculated values of $R_{\mathrm{step/surface}}$ are 0.12 ± 0.02 and 0.27 ± 0.04 for the RT and LT films, respectively.

The structure of the films was investigated by LEED. Both kinds of films display a similar (1×1) LEED pattern. I/V-LEED curves were taken in the thickness range up to 21.4 ML. The intensity of the (00) beam was recorded at an incident angle approximately 6° off normal. Figure 2 gives the LEED curves of the (00) beam for both RT and LT/*I/V* films. The solid lines in Fig. 2 mark the peak positions of the spot intensity vs beam energy (*I/V*) curves from the clean Cu substrate, the dashed lines follow the peaks from the Ni films. Within the kinematic theory, the average interlayer distance of the films can be determined from the Bragg peak positions on the wave vector axis. The measured average interlayer distances of both kinds of films as a function of Ni films thickness are displayed in Fig. 3. Within the limits of uncertainty, the average interlayer distances are the same for both kinds

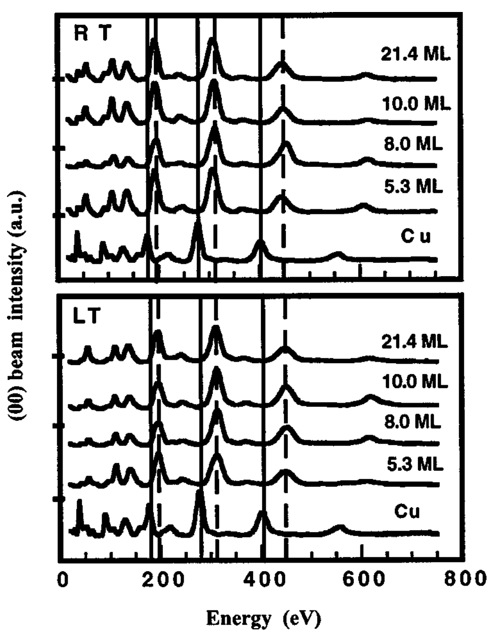

FIG. 2. *I/V*-LEED spectra of the (00) beam for the RT-grown and 170 K-grown Ni/Cu(100) films. The dashed and solid lines indicate the peak positions of the Ni films and of the Cu substrate, respectively.

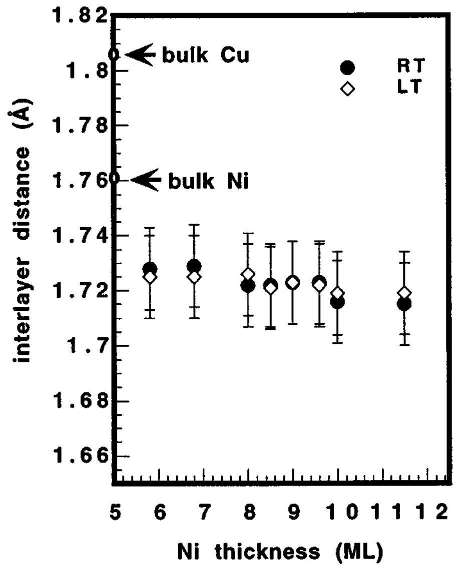

FIG. 3. The interlayer distances calculated from the curves of Fig. 2 for both the RT-grown and 170 K-grown Ni/Cu(100) films.

of films. Assuming pseudormorphic growth, one may obtain the c/a ration of both films from these measurements. We find $c/a=0.97\pm0.01$ in both cases. This result agrees well with a full dynamic *I/V*-LEED analysis indicating a 4% contraction for both 5 and 11 ML films averaging the interlayer distance of topmost four monolayers.[13]

Figure 4 gives the polar MOKE hysteresis loops for both kinds of films measured at 160 K. The magnetic field was applied normal to the film plane and was large enough to saturate all the samples. Both RT and LT films show a spin reorientation from in-plane to perpendicular. The critical thickness of the RT films is about 9.5 ML, below which the Kerr loops are hard-axis-like and above which the Kerr loops become rectangular. This result agrees well with a recent theoretical calculation indicating that the critical thickness is around 9 ML when the films have a c/a ratio of about

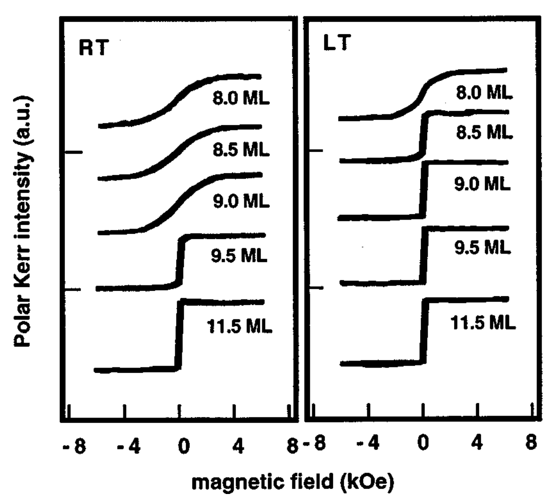

FIG. 4. Polar MOKE hysteresis loops for the RT-grown and 170 K-grown Ni/Cu(100) films. Note that for the RT films the switch from the hard-axis-like loops to easy axis loops occurs at a thickness higher by about 1 ML than for the LT films.

0.974.[14] For the LT-grown films the transition occurs at 8.5 ML, which means the critical thickness for the LT-grown films is about 1 ML smaller than that of RT-grown films.

The effect of growth temperature on the surface anisotropy and the volume anisotropy needs to be investigated. As we pointed out above, the LT films have much more step edge atoms than that of the RT films due to the higher density of islands. From Bruno's calculation,[15] each step atom contributes to a decrease of the surface anisotropy, whatever its sign. Hence the surface anisotropy of LT films should be smaller than that of RT films.

The fact that the LT films have a larger island density compared to that of the RT films raises the question whether the LT films may have a smaller strain than that of the RT films, since the step edge atoms often offer means for the strain relaxation. If this were the case, the LT films would have a smaller strain-induced anisotropy and hence a higher critical thickness of the spin reorientation if everything else remains the same. Obviously, our results are at variance with this hypothesis.

In order to confirm that the smaller critical thickness for LT-grown Ni films is caused by the reduced surface anisotropy, we made a simple evaluation according to Refs. 15 and 16. We divided the contribution of surface anisotropy into two parts, one is the interface anisotropy K_i and the other is the surface anisotropy K_s. So the critical thickness d_c can be determined by:

$$d_c=(K_i+K_s)/K_v\,. \tag{1}$$

For simplicity, we first assume K_i to be the same for both types of films. In view of our finding of c/a being the same for both types of films, we also assume the volume anisotropy K_v to be the same. Taking into account the influence of step edge atoms and the roughness, the difference of the critical thicknesses can be expressed by:

$$d_c^{\mathrm{RT}}-d_c^{\mathrm{LT}}=(K_s^{\mathrm{RT}}-K_s^{\mathrm{LT}})/K_v\,, \tag{2}$$

while, $K_s^{\mathrm{RT(LT)}}=K_s-K_s^{\mathrm{step}}-K_s^r$.
K_s^{step} represents the contribution of step edge atoms, and K_s^r is an effective perpendicular anisotropy caused by the surface roughness.[16] As Bruno pointed out,[15] each step atom contributes half of the surface anisotropy of atoms in the plane, i.e. $K_s^{\mathrm{step}}=R_{\mathrm{step/surface}}\times K_s/2$. K_s^r can be simply calculated using:[16]

$$K_s^r=\tfrac{1}{2}4\pi M_s^2\,\tfrac{3}{4}\,\sigma\{1-f[2\pi(\sigma/\xi)]\}. \tag{3}$$

Here σ and ξ denote the average vertical and lateral length of the islands. From the line profiles of the STM images, σ and ξ are obtained to be 0.5 and 10.8 nm for the RT films, and 0.4 and 4.6 nm for the LT films, respectively.

Taking the 300 K values of K_s and K_v from Ref. 2 and 300 K values of M from Ref. 17, we obtained $d_c^{\mathrm{RT}}-d_c^{\mathrm{LT}}=0.5\,\mathrm{ML}$. Considering K_i and K_v possibly to be different in LT and RT films, and the rough assumption that the $R_{\mathrm{step/surface}}$ values are thickness independent around the critical thickness regime, the calculated value is reasonably close to our experimental result of 1 ML ($\pm$0.3 ML). This means that the reduced surface anisotropy is the main reason for the smaller critical thickness of the Ni films grown at low temperature.

IV. SUMMARY

Smaller critical thickness of spin reorientation was found for 170 K-grown Ni films compared with RT-grown films. High density, smaller rectangular islands were obtained when the Ni films were deposited at low temperature. But the crystalline structure does not change with the different depositing temperature. It is thought that the reduced surface anisotropy caused by the increase of the number of step edge atoms in LT films is the main reason for the critical thickness to be reduced by about 1 atomic layer.

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