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# Micromagnetic properties of ultrathin cobalt films

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The properties of the magnetic domain structures of ultrathin fcc cobalt films epitaxially grown on Cu (001) have been examined using an ultrahigh vacuum surface magneto-optic Kerr effect instrument. The evolution of magnetic behavior is observed for film thicknesses ranging from 1.4 to 7.5 monolayers. The coercivity is sensitive to film growth temperature and thermal cycling history. The coercivity decreases with diminishing film thickness and falls to very low values for the thinnest layers. The results are discussed in terms of Néel domain-wall micromagnetics for ultrathin films.

## INTRODUCTION

There has been much recent experimental and theoretical work aimed at a clearer understanding of the properties of ultrathin ferromagnetic layers.[1] In particular, much of this interest has focused on the decrease in Curie temperature[2] with decreasing thickness and the critical behavior of these "two-dimensional" magnetic systems.[3] However, much less is known concerning their domain structure and related hysteresis behavior. Scanning electron microscopy with secondary-electron polarization analysis has been employed to image domains of ultrathin bcc Fe films epitaxially grown on a Ag(001) substrate[4] and ultrathin fcc Co films epitaxially grown on Cu(001) substrates.[5] These studies have revealed that for film thicknesses ranging from 3 to 19.5 monolayers, the domains are very large, typically several hundred microns.[5] In the "as-grown" state, the films show a single domain configuration with small 180° edge domains. Demagnetizing the films *in situ* with an ac magnetic field breaks up the single domain state, giving rise to a multidomain structure even in the thinnest films studied (3 monolayers). In the case of the fcc cobalt films on Cu(001) between 3 and 19.5 monolayers, 90° Néel domain walls dominate over 180° domain walls,[4,5] with the magnetization aligned in-plane along the ⟨110⟩ directions. The same behavior is observed for much thicker films with thicknesses exceeding 100 Å.[6] Another interesting characteristic of the domain structure is the very irregular shape of the domains and the extremely rough nature of the domain walls in these ultrathin films. This has been interpreted[5] as an indication of a vanishing magnetic field energy contribution to the Néel wall energy.

The domain-wall surface energy for Néel walls in films of thickness $t$, which is small compared to the wall width, is generally written as the sum of two terms:[7]

$$\gamma_t = \gamma_w + 2\pi t M_{\text{eff}}^2. \tag{1}$$

The first term contains exchange and anisotropy energies. The second term accounts for the effective magnetic field energy associated with the moment rotation across the domain wall. The field energy decreases linearly with film thickness until the wall energy is dominated by the first term for thicknesses of a few monolayers. The consequence is that in very thin films the wall energy term alone does not fix the wall orientation along any particular direction. Domain-wall roughening is promoted by local variations in the exchange and anisotropy energies due to interface and surface roughness effects. In addition, it has been suggested that domain-wall roughening may also result from a lowering of the free energy by including the configurational entropy.[5]

In this paper we report a detailed study of the hysteresis behavior of ultrathin films of fcc cobalt grown epitaxially on Cu(001) single-crystal substrates for thicknesses from 1.4 to 7.5 monolayers. Specifically, we observe a strong dependence of the micromagnetic properties on temperature, film thickness, thermal cycling, and film growth conditions which lend strong support to the previously reported domain structure[4,5] observations and an interpretation based on Néel domain-wall energetics.

## EXPERIMENT

Cobalt grown epitaxially on Cu(001) is ferromagnetic in a fcc phase. The Cu(001) substrate was prepared by repeated 500-eV $Ar^+$ bombardment and annealing to 1000 K. The electron bombardment evaporation sources were emission current stabilized to produce a constant flux at the sample equivalent to 1-monolayer deposition in 3 min. A quartz crystal microbalance located at a quarter of the source/substrate separation monitored the flux. The flux was calibrated with reflection high-energy electron diffraction (RHEED) oscillations and thicknesses checked by Auger electron spectroscopy (AES) peak intensity ratios. The background pressure during evaporation was $<5\times10^{-10}$ mbar and film purity was better than 97% as measured by AES. Crystallographic order was monitored with low-energy electron diffraction (LEED).

Magnetization vs magnetic field hysteresis loops were obtained with an ultrahigh vacuum (UHV) surface magneto-optic Kerr effect (SMOKE) instrument.[8] Longitudinal SMOKE measurements were made in-plane along the ⟨100⟩ and ⟨110⟩ directions. The typical applied magnetic field was cycled ±125 G at a frequency of 0.1 Hertz. Hysteresis loops of 800 data points were collected by averaging over 10 sequential scans. Typical hysteresis loops along the ⟨100⟩ direction for 1.5 and 7.5 monolayers are shown in Fig. 1. Least-square fitting over appropriate loop segments are used to extract data on the saturation magnetization, remnant magnetization, and coercive field $H_c$. Coercivities presented in this paper were obtained from the loops measured along the ⟨100⟩ direction and have an

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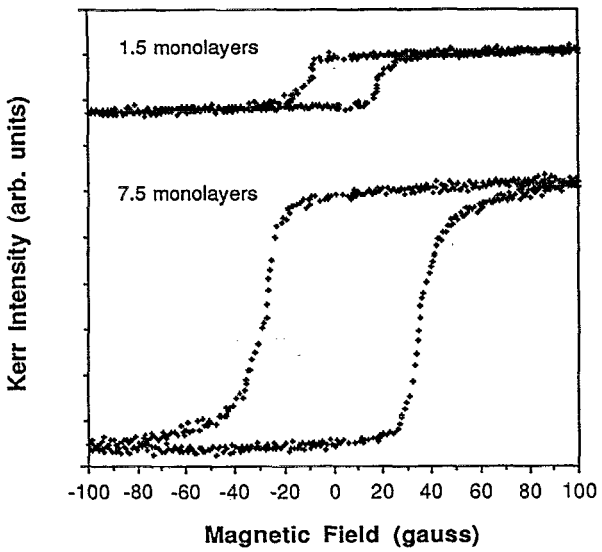


FIG. 1. Hysteresis curves for 1.5- and 7.5-monolayer thick films of fcc Co(001) epitaxially grown on Cu(001) at 300 K and measured at 150 K.

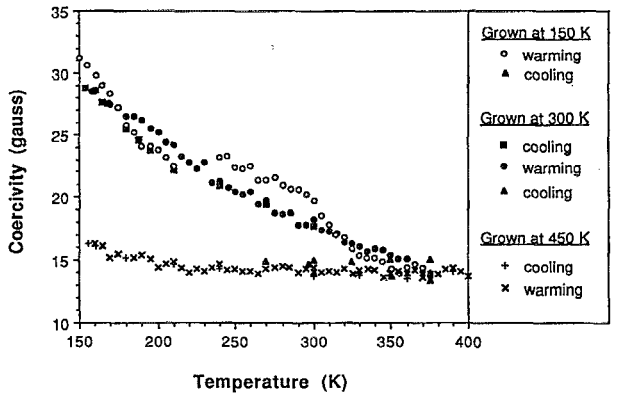


FIG. 2. Variation of coercive field with temperature for 2.0-monolayer Co/Cu(001) films grown at 3 different substrate temperatures.

estimated error of 0.5 G. For thicknesses above 2 monolayers coercivities along the ⟨100⟩ direction are typically a few gauss smaller than along the ⟨100⟩ direction indicating the development of an in-plane anisotropy. Films up to 7.5 monolayers show no significant change in Kerr intensity from 100 to 300 G indicating that they are saturated in both directions, but further verification of this with higher fields is planned.

## RESULTS AND DISCUSSION

In a previous publication[9] we have determined that the cobalt films grown on Cu(001) at 150, 300, and 450 K have a very different microstructure. LEED and RHEED patterns show the 150-K deposited films to be rough on a microscopic scale. Films grown on 300-K substrates produce good $p(1\times1)$ LEED patterns with a low background intensity. Films prepared at 450-K substrate temperature have sharp $p(1\times1)$ LEED indistinguishable from that of a well-prepared Cu(001) substrate crystal. RHEED oscillations are observed from both the 300- and 450-K films indicating layer-by-layer growth. Hydrogen gas titration experiments[9] indicate, however, that the 450-K films are capped with an ordered Cu(001) overlayer due to substrate surface segregation during film growth at this highest temperature. These data strongly suggest that the optimal growth temperature to obtain uniform ultrathin fcc Co(001) films is 300 K. At this growth temperature the cobalt covers the Cu(001) substrate at the equivalent of 1.2 monolayers.[10,11]

The differences in film microstructure are reflected in the micromagnetic behavior of the coercive field with temperature, Fig. 2, but not in an obvious way. For example, Fig. 2 compares the $H_c$ vs temperature behavior of a 2-monolayer film deposited at the above three temperatures and shows that the largest discrepancy in behavior is the film deposited at 450 K, for which $H_c$ remains almost independent of temperature except for a slight upturn below 250 K. In contrast, both the "rough" 150-K film and the "smooth" 300-K film anneal irreversibly down to the 450-K film value with increasing temperature. Cooling the as-grown 300-K film to 150 K and warming back to 400 K follows the general behavior of the 150-K annealing curve. This behavior suggests that the magnetic structure in the 150- and 300-K films anneals to some equilibrium value, $H_c \approx 15$ G, largely independent of the microstructural roughness of the lower growth temperature films.

The magnetization and the Curie temperature ($T_c$) of these films is a sensitive function of film thickness. SMOKE measurements of the magnetization versus temperature were used to determine the film Curie temperature. 1.0-monolayer films have a $T_c$ below 150 K. 1.5-monolayer thick films have a $T_c$ near 350 K which is nearly independent of the film growth temperature. Films thicker than 2 monolayers have a $T_c$ above 450 K, the precise value being sensitive to any copper interdiffusion.

Variations of the coercive field with film thickness and temperature are shown in Fig. 3. Films of varying thick-

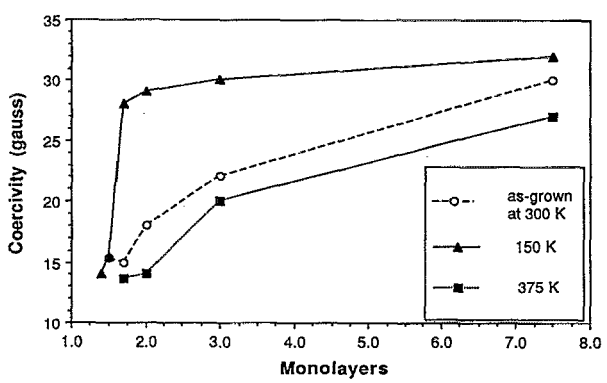


FIG. 3. Variation of coercive field with film thickness showing the behavior at different film temperatures.

ness were grown at the optimal temperature of 300 K. SMOKE measurements were then performed successively at 300, 150, and 375 K. Films measured at 300 and 375 K show a gradual increase in $H_c$ with film thickness. In contrast, the coercivities measured at 150 K stay close to an asymptotic "bulk" value above about 2.0 monolayers, but below 2 monolayers collapse to the same value as the 300-K measurements, $H_c$ about 15 G.

We interpret this behavior in terms of a decreasing magnetostatic energy contribution to the total Néel wall energy [second term in Eq. (1)]. The behavior of the domains for film thicknesses of a few monolayers is dominated by the wall energy contribution, $\gamma_w$, in agreement with previous work.[5] The wall boundaries are no longer restricted to orient along distinct directions that keep the perpendicular magnetization to the wall constant. In addition, wall motion is promoted by decreasing in-plane anisotropy and an increasing importance of the configurational entropy. These effects increase with decreasing film thickness such that $H_c$ decreases, as is observed in Fig. 3. Lowering the temperature arrests wall diffusion, while raising the temperature accelerates wall movement; as shown in Fig. 3, the largest change being observed for temperatures between 150 and 300 K. For films $<2$ monolayers, we observe only low $H_c$ values, about 15 G, again consistent with a negligible magnetostatic energy contribution and a lowered activation energy to wall diffusion.

## CONCLUDING REMARKS

We have observed thickness and temperature-dependent magnetic hysteresis behavior in ultrathin ferromagnetic films which can be explained in terms of the domain structures and Néel wall micromagnetic behavior recently observed using spin-polarized domain imaging methods.[4,5] Film between 2.0- and 7.5-monolayer thick show an increased sensitivity of the coercivity on temperature. Film thicknesses below 2 monolayers show little variation in coercivity down to 150 K. The observed differences can be explained in terms of micromagnetic structure and energetics which reflect a stronger dependence on film thickness rather than detailed microstructure. In these two-dimensional layers the vanishing magnetostatic field energy contribution to the total wall energy implies that we can no longer neglect the configurational entropy associated with the rough domain walls. This becomes a dominant contribution to the free energy so that wall boundaries can more easily reconfigure, resulting in lower values of the coercive field, particularly for films thicknesses below 2 monolayers.

## ACKNOWLEDGMENT

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