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MAGNETIC ANISOTROPY AND LATTICE STRAIN IN Co/Pt MULTILAYERS

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Abstract

We report on the correlation between perpendicular anisotropy and in-plane lattice strain in Co/Pt multilayers. $(Co_x/Pt_y)_n$ samples, where x, y are the thickness of the individual Co and Pt layers and n is the number of repeats were prepared by Molecular Beam Epitaxy and studied by means of polar Magneto-Optic Kerr effect and transmission electron microscopy. Kerr rotation data and electron diffraction experiments show that the largest perpendicular anisotropy and square hysteresis loop occur when x = 3Å while the Pt layers are subjected to about -2% in-plane strain. As Co thickness increases, Co and Pt layers gradually lose coherency and the magnetic anisotropy goes from perpendicular to planar. This is accompanied by a relaxation of lattice strain in both Co and Pt layers. The close relationship between magnetic anisotropy and lattice strain can be explained as magneto-elastic anisotropy or stress anisotropy effect due to lattice mismatch between the adjacent epitaxial layers.

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Introduction

To satisfy the demands of the information storage industry a lot of effort in recent years has been put into the development of erasable, long life-time, high density magneto-optical (MO) recording media The magnitude of the read out signal in MO recording is related to the optical reflectivity R and the Kerr-rotation $\varnothing_k$. The latter is particularly enhanced if the thin film media exhibits anisotropy perpendicular to the film plane. It has been well established that nanometer scale multilayers comprising of alternating ferromagnetic and non-magnetic layers, under certain conditions, can exhibit strong perpendicular anisotropy [1,2]. However, the origin of the magnetic anisotropy of such multilayer films is an unanswered question and remains a lively and challenging subject of investigation among material scientists and physicists. Although, in principle, it was pointed out long ago that the change of symmetry at a surface (or interface) might result in anisotropy energy different from the bulk value[3], an improvement of the interface-induced anisotropy and a better understanding of the attendant mechanisms are needed for making this type of magnetic film useful in future.

Experimental Details

All samples in this study were prepared by molecular beam epitaxy (MBE). Multilayers of $(Co_x/Pt_y)_n$ were grown on the As terminated [111] surface of GaAs substrates with a 200 Å Ag buffer layer. Six samples with (x/y)=(3/18), (6/18.5), (10.6/18), (13.5/19.1), and n=15, as well as $(Co_{50}/Pt_{50})_4$ and $(Co_{4.4}/Pt_{4.4})_{15}$ were examined. The Kerr-rotation measurements were performed using a polarized laser beam with a wavelength of 633 nm. TEM samples were prepared by mechanically thinning and ion-milling. The lattice strain was measured using the position and displacement of the diffraction spots from the Co and Pt layers and calibrated using the standard reflections from the GaAs substrate.

Results and Discussion

The effective anisotropy energy, K_{eff}, of the samples $(Co_x/Pt_y)_{15}$, x=3, 6, 10.6 & 13.5; y=18, 18.5, 18, 19.1 decreases monotonically as Co thickness increases. The perpendicular anisotropy (K_{eff} positive) persists upto a Co layer thickness of 20 Å[4] and becomes in-plane for the sample $(Co_{50}/Pt_{50})_4$, i.e. K_{eff} negative. The measured Kerr-rotation curves are shown in Fig.1. The area within the hysteresis loops, from (a) to (e), decreases and equals zero at a Co thickness of 50 Å in $(Co_{50}/Pt_{50})_4$. The measured intrinsic coercivity values are 4.083×10^3 Oe for $(Co_3/Pt_{18})_{15}$, 1.096×10^3 Oe for $(Co_6/Pt_{18.5})_{15}$, 0.208×10^3 Oe for $(Co_{10.6}/Pt_{18})_{15}$, and 0.159×10^3 Oe for $(Co_{13.5}/Pt_{19.1})_{15}$.

Examining the relationship between the properties and the microstructure of these samples, we found that in the sample $(Co_3Pt_{18})_{15}$ which has the highest K_{eff}, both Pt and Co layers are severely strained in the film plane due to the 10% lattice mismatch between the two layers. There is a ~2% compressive strain in the Pt layers while the Co layers are under tension along <220> crystallographic directions. The Co monolayers are difusive and a single crystal [111] zone diffraction pattern, indicative of intermixing, is shown in Fig. 2a. As the Co layers are grown thicker, though the fcc stacking remains the same[5], relaxation occurs. The (220) type reflections in Fig. 2b corresponding to the sample $(Co_6/Pt_{18.5})_{15}$ indicate elongation along <220> directions. When Co thickness reached 10.6 Å - almost 5 MLs, in sample $(Co_{10.6}/Pt_{18})_{15}$, diffraction spots of Co are strong and clearly separated from those of Pt. These are shown in Fig. 2c with an average strain of -1.7% (Pt) and +1.8% (Co). This trend is sustained in the sample $(Co_{13.5}/Pt_{19.1})_{15}$ (Fig. 2d). Finally, both Co and Pt layers in $(Co_{50}/Pt_{50})_4$ relax to average zero strain (Fig. 2e) and the anisotropy is completely in-plane as shown in Fig.1e. The variation of lattice strain as a function of Co thickness is plotted in Fig. 3. Moreover, lattice strain in the film growth direction measured using cross-sectional samples of $(Co_{10.6}/Pt_{18})_{15}$ indicate that the Co layers are negligibly strained along [111], whilst the Pt layers are strained by -2.4%. We can thus conclude that overall, the Pt layers within the film experience 3-D compression and Co layers are in 2-D tension. This specific strain configuration, especially the in-plane strain in the Co layers, has a direct ramification on the effective anisotropy energy.

Based on the above observations, we predicted that even if we make a sample with 1-2 monolayer of Co and Pt, as long as the Co and Pt layers are strained in the film plane, we should be able to get fairly good perpendicular anisotropy. Therefore, a sample of $(Co_{4.4}/Pt_{4.4})_{15}$ was made. The elastic strain measured from the Co layer is +6% and the Pt layer is -1.3%. It is interesting to note that the elastic strain of the Co layer as well as the anisotropy energy of the new sample stays in between samples Co_3/Pt_{18} and $Co_6/Pt_{18.5}$. Fig. 4 shows the diffraction pattern and the Kerr-rotation. The coercivity of the sample is 1.248 kOe. In the diffraction pattern, the two sets of (220) reflections from the Co and Pt layers are clearly resolved. The inner spots are from the Pt layers and outer spots from the Co layers.

There is evidence for interface mixing in these multilayers and this comes from x-ray photoelectron diffraction[6], x-ray diffraction[7] and HREM[8]. However, the exact composition of the interface alloy phase, if any, has not been conclusively determined. Even though the synchrotron grazing incident x-ray diffraction data has suggested the presence of an ordered $CoPt_3$ phase at the interface, this has not been supported by our data. There is no observable intensity at the (110) type positions of the

[111] diffraction patterns arising from the ordered $CoPt_3$ phase even for long time exposures.

The anisotropy energy, K_{eff} vs. Co thickness, t can be phenomenologically described as[9]

$$K_{eff}\ t = K_v\ t\ +\ 2\ K_s \qquad (1)$$

where K_s refers to the interface anisotropy per unit area and K_v, the contribution per unit volume of Co, is generally written as

$$K_v\ =\ -1/2\ \mu_0 M_s^2 + K\ -\ \text{constant}\ \lambda\ \sigma \qquad (2)$$

where the first term is the demagnetization energy, K is the anisotropy energy of the bulk Co and the last magnetoelastic term is the contribution due to stress (σ) and the related magnetostriction (λ). It is common practice to write equation (1) as $K_{eff} = K_v + 2\ K_s/t$, assume K_v to be a constant and consider K_{eff} to be inversely proportional to t. However, as we would like to emphasize the role of the in-plane strain in creating and stabilizing the easy axes of magnetization in this paper, the third term in equation (2) can become important.

Spin-orbit coupling which determines the crystal anisotropy is modified by the sign of the lattice stress and the magnetoelastic interactions in thin films affect their magnetic behavior. In simple terms, the axis of stress is an easy axis if the magnetoelastic contribution to the anisotropy constant ($\lambda\ \sigma$) is positive. On the other hand, if ($\lambda\ \sigma$) < 0, the stress axis is a hard axis and the favoured direction of magnetization lies in a plane perpendicular to the stress axis. It has been reported[10] that the residual stress after plastic deformation induces uniaxial anisotropy in pure Ni and the maxima of the torque curve can be 10 times higher than the annealed sample. Since the saturation magnetostriction of Ni is negative in all principle directions the residual compressive stress creates an axis of easy magnetization parallel to the stress axis and a residual tension creates an easy axis perpendicular to the stress axis.

A similar argument should also be valid for the Co layers that persist in the fcc structure up to a thickness of 50Å in our multilayer stacks[5]. The magnetostriction of Co_{fcc} on the close packed (111) planes can be considered to be similar to the one on the hexagonal planes of Co_{hcp}, i.e. $\lambda < 0$[11]. In addition, the Co layers are under tension, i.e. $\sigma > 0$, in the film plane due to its lattice parameter being smaller than Pt. Therefore, for (111) growth, the product ($\lambda\ \sigma$) is negative and a uniaxial anisotropy perpendicular to the in-plane stress axis is favoured. Furthermore, magnetoelastic anisotropy can contribute to both K_v and K_s terms in equation (1) because of the stress gradient or periodic stress variation coupled with chemical variation along the film normal. At the interfaces the lattice strain exceeds the measured average value and this will contribute to the interface anisotropy. Towards the center of the Co layers the lattice strain will be less than the average value. The average strain-stress field within the Co layer will contribute to K_v. Even though other factors such as the overall microstructure of the multilayer stack[8], atomic mixing or compound formation[4,8] or the polarization of the Pt layers[2] at the interface may contribute to the perpendicular anisotropy, magnetoelastic effects discussed in this paper are quite important in interpreting the anisotropy of magnetic multilayers.

Conclusions

We have measured the in-plane lattice strain in a series $(Co_x/Pt_y)_n$ samples and related it to the anisotropy energy. We propose that the magnetoelastic or stress anisotropy plays an important role in inducing perpendicular anisotropy in these multilayer magnetic films.

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Captions

Figure 1. Polar Kerr-rotation curves from the samples a) $(Co_3/Pt_{18})_{15}$, b) $(Co_6/Pt_{18.5})_{15}$, c) $(Co_{10.6}/Pt_{18})_{15}$, d) $(Co_{13.5}/Pt_{19.1})_{15}$, e) $(Co_{50}/Pt_{50})_4$.

Figure 2. Plan-view diffraction pattern from the samples a) $(Co_3/Pt_{18})_{15}$, b) $(Co_6/Pt_{18.5})_{15}$, c) $(Co_{10.6}/Pt_{18})_{15}$, d) $(Co_{13.5}/Pt_{19.1})_{15}$, e) $(Co_{50}/Pt_{50})_4$ showing the gradual splitting of the (220) type of reflections. The inner spots are due to twinning and double positioning.

Figure 3. In-plane lattice strain of Co and Pt layers vs. the thickness of Co in each layer. The data for Co in Co_3Pt_{18}, ~8%, is not physically representative due to the possible intermixing of the Co monolayers and the consequent difficulty in the evaluation of the strain.

Figure 4. a) Polar Kerr-rotation curves of sample $(Co_{4.4}/Pt_{4.4})_{15}$ with a coercivity of 1.248 kOe; b) plan-view diffraction pattern showing the distinct (220) reflections from Co and Pt, as well as Ag buffer layers.

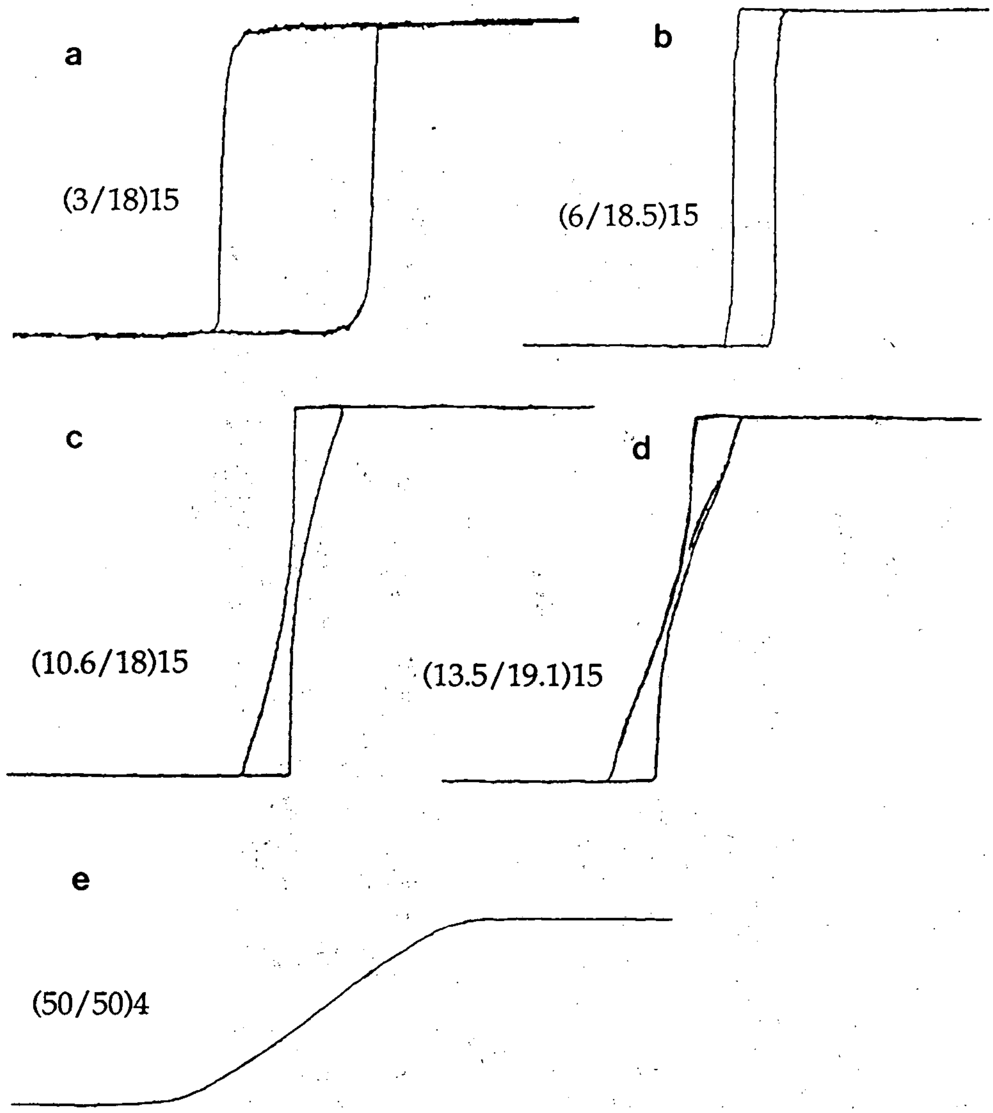
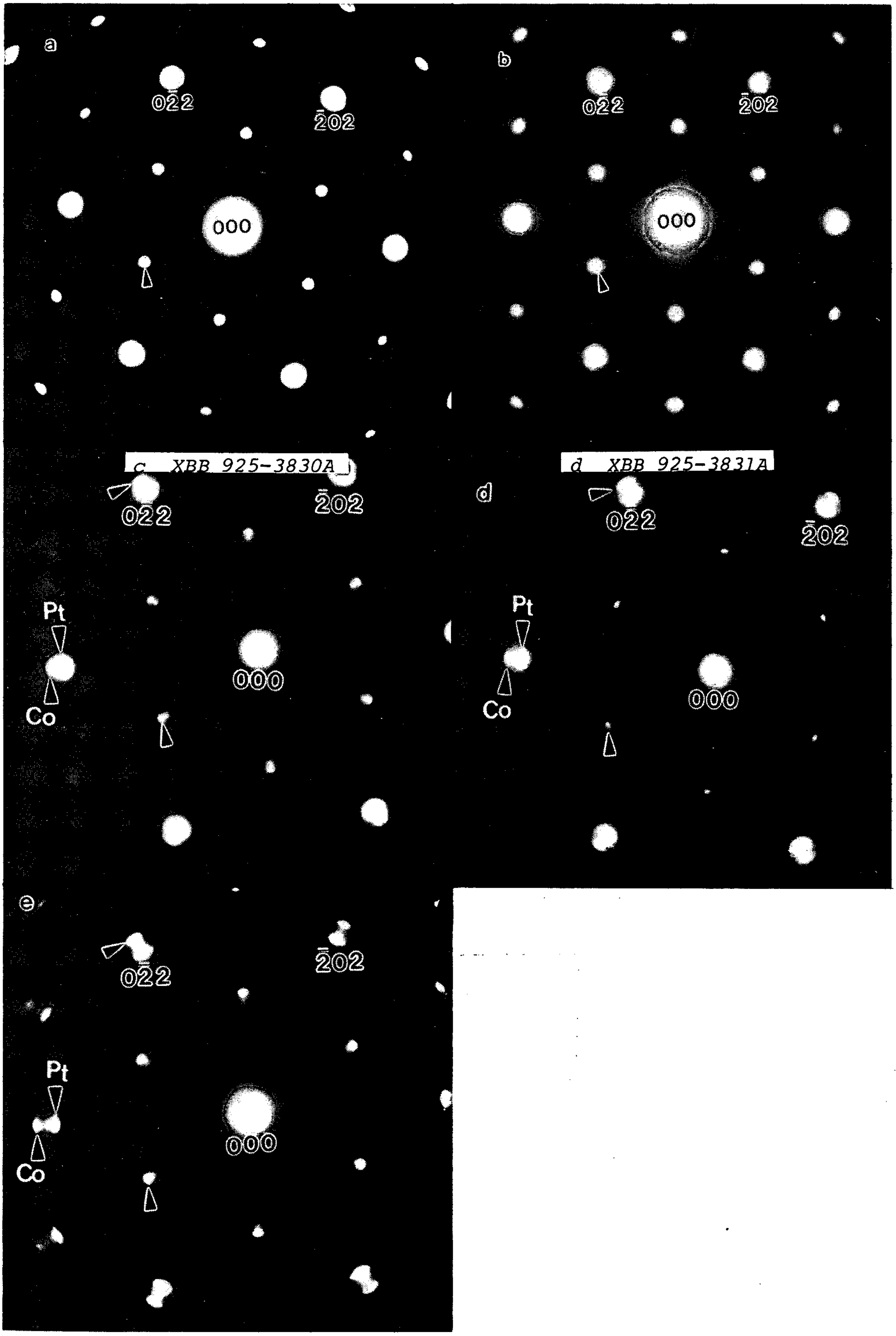


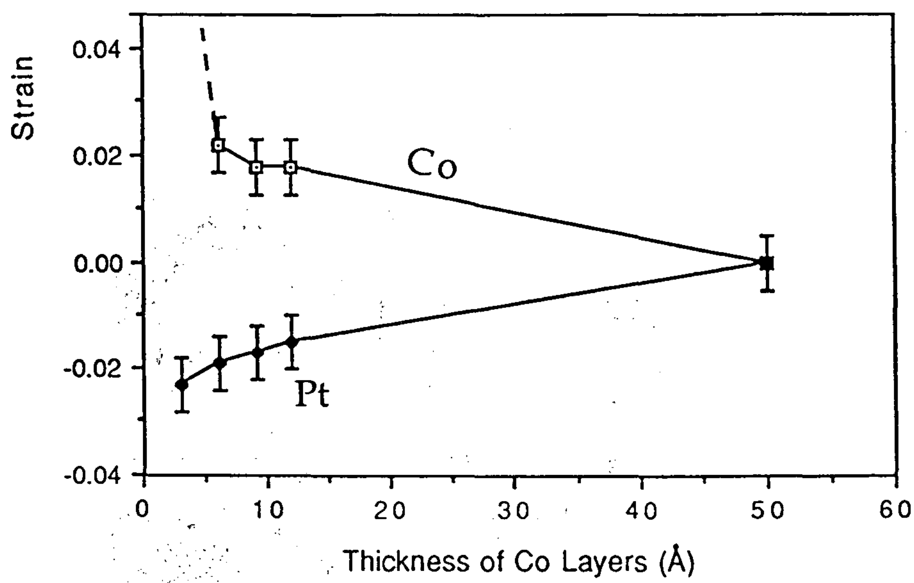
Figure 1

a XBB 925-3828 b XBB 925-3829



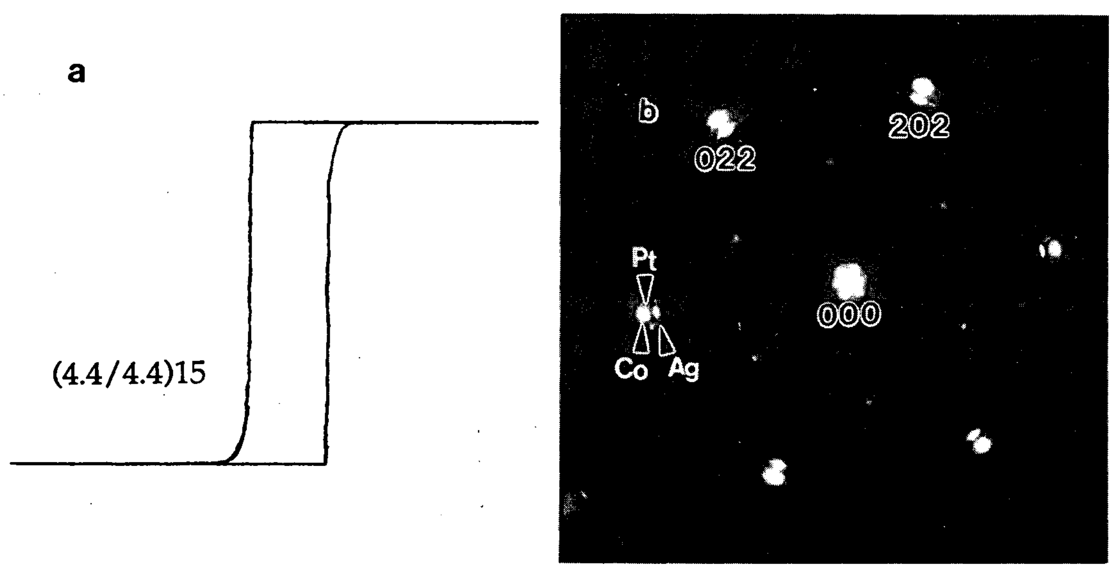
e XBB 925-3832A

Figure 2



Strain
0.04
0.02
0.00
-0.02
-0.04
Co
Pt
0
10
20
30
40
50
60
Thickness of Co Layers (Å)

Figure 3



XBB 927-7440

Figure 4

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