Abstract. R.F. Magnetron sputtered CoCr films (79/21 at%) with various thicknesses are magnetically characterized. The domain structure is observed by digitally enhanced Kerr microscopy and depends on the Hc/Hk values of the samples. For low and high coercivity films a comparison is made between the measured VSM hysteresis, domain period and a theoretical domain del. The domain shape is a function of the magnetic history of the sample and the bending created by the deposition process.

INTRODUCTION

The reversal mechanisms of magnetization in sputtered CoCr layers for perpendicular recording has still not been completely clarified. Studying the domain configuration can contribute to the solution of this problem. The magnetic properties of CoCr layers are strongly influenced by variations in microstructure, morphology and chemical homogeneity. Initial layer effects, or segregations at the column boundaries and surface are very important for the final behaviour of the domains. The morphology also depends on the sputtering gas used in the deposition process.

M.O. KERR OBSERVATIONS

The first Kerr observations on CoCr layers with divergent properties are published in [13]. There a specially designed digitally enhanced Kerr microscope with an image processing system is used. The films used have a composition from 17 to 21 at% Cr, thicknesses from 90–620 nm and coercivities from 10–50 kA/m with one sample of 50 kA/m. They all show a more or less stripe-domain structure at remanence. It was concluded from this that all samples reversed by domain-wall movement as a continuous medium. Using the same microscope mentioned above, we came to the conclusion with our preliminary results [14] that we have to divide our CoCr layers into 3 categories namely: low Hc (long stripes), medium Hc with short stripes (80–140 kA/m) and high coercivity films with a dot-like domain structure. The domain behaviour is clearly connected with the shoulder in the perpendicular hysteresis loop. The Kerr observation method has its limitations in the resolution (~0.7 nm) and depth of information (~15 nm). An advantage in relation to transmission Lorentz microscopy is that surface magnetic information can be obtained from all thicknesses without any thining techniques with their consequent disadvantages. A comparison between both observation methods is mentioned in [5], i.e. the Kerr image of a 0.6 μm thick sample (30 at% Cr) and a Lorentz image of 0.1 μm thick (18 at% Cr) both showing a stripe-like domain pattern. The interpretation of these images is not so simple. The differences in thickness, chemical composition and eventually removal from the original substrate will have an important influence on the domain behaviour. Besides, observation of double layered samples is completely impossible with the Lorentz method. Another advantage of Kerr microscopy is the possibility of applying fields up to 800 kA/m. For most of the CoCr layers the domain process through the whole hysteresis loop can be observed [13,14] within the limitation of the resolving power.
A series of samples was made and the measured and calculated results from Kerr observations, VSM and torque magnetometer are given in Table I. The Q-factor of our samples is derived from $Q = k_1/0.5 \mu_0 M_g^2$. For $Q > 1$ the perpendicular axis will thus be the direction of preferred magnetization in the case of a uniform magnetized layer. In the demagnetized state the magnetization is split up into domains. The condition for stable perpendicular magnetization $\mu_0 M_g$ is expressed in the $Q$-factor. For our samples $Q$ varies from 1.30-1.50. The domain shape is not dependent on this $Q$-range, but on the coercivity of the layers. Of course, the $H_c$ is strongly dependent on the reversal mechanism and the magnetic morphology (Curvature-evolution at the column boundaries). In order to explain the observed stripe domains we have used the Kooij and Esh model [13]. The two relevant equations which have to be solved are explicitly given in [4]. The parameters used are: film thickness ($h$), material parameter ($\lambda$), domain period ($P$) and $P = 1 + M_s/2K_1$ which was kept constant at 1.70. Both the equations are solved for several pairs of $h/\lambda$ and $H/M_s$ values. In Fig. 1 the relation $M_s/H$ vs. $H/M_s$ and in Fig. 2 $P/\lambda$ vs. $H/M_s$ are shown. From both plots (see broken lines) $h/\lambda$ is used as a parameter. From VSM measurements we know the hysteresis loop and from this we determine the slope $T = dH/dH$ for $M_s = 10$ or $M_s = 20$. This initial slope is the most relevant parameter. An other relevant parameter $P$ is measured from the Kerr photographs as a function of the applied field.

The measured slope is also in good agreement with the calculated line for $h/\lambda = 00$ in Fig. 2. For this sample $\lambda = 20.8$ nm, $\mu_0 M_g = \mu_0 M_g^2 = 4.2 \times 10^{-3}$ J/m$^2$, $A = (1/K_1)$, $(\mu_0 H/4) = 5.4 \times 10^{-3}$ J/m and $1 = \mu_0 M_g/K_1 = 25$ nm.

Another low $H_c$ film (#2) has also been measured and calculated. Here we found $h/\lambda = 55$ as the best fit. For sample #6 (high $H_c$) we could not fit both curves (see Figs. 1, 2).

We have also compared the size and shape of the domain period of #2 in the remanent state after saturation with perpendicular $H_c$ and demagnetizing fields (Fig. 3).

**Fig. 3 Domain configurations of a low coercivity film in remanent (left) and demagnetized (right) state.**

At remanent state $H_{rm} = 910$ mT and after demagnetization $H_0 = 600$ mT. This minimum energy position of the stripes results in a very good parallel orientation. Domain observations in the field free state show the magnetic history of the sample. Also different domain configurations can be observed after partial saturation and applying a field under an angle [13]. For a high $H_c$ sample (##6) the dot-like domain shape at demagnetizing field can be roughly given as $H_0 = 665$ mT. Typical shoulders in VSM and M.O. Kerr hysteresis loops are not seen in the high $H_c$ films. The calculated model by means of $P$ and $T$ is not in good agreement for each film (see Figs. 1, 2).

We have to realize that the method we used makes the following assumptions: namely: the domain-wall energy is independent of $h$, the wall volume is neglected and, last but not least, the coercivity is zero.

Stresses have a lot of influence on the domain configuration. The bending of the samples is measured. A line scan over the total sample length in two orthogonal directions is made. The curvature (e.g. the strain) is not always isotropic over a certain area and can vary from place to place in the same sample. We found that samples have a different bending in the orthogonal directions showing straight stripe domains (see Fig. 4a). More or less meandered domains are seen in Fig. 4b which is made from a position with equal bending in two directions.

**Fig. 4 Domains of #2 at $H_0 = 268$ kA/m as a function of the bending in two orthogonal directions (xy).**

At the moment we have the impression that stress in our magnetron sputtered films is higher than in our SF sputtered ones [4]. Consequently the stress will have to be studied in more detail in relation to the total anisotropy factor in the films.

**MORPHOLOGY**

Cross-sections were studied with the STEM mode of the electron microscope. The fractal surfaces were prepared by cooling the samples in liquid $N_2$ before breaking them, to obtain a brittle fracture. The columnar-like morphology can be clearly seen in all our films independent of the $H_c$/k$_c$ value.
For samples 3, 5 and 8 respectively the average column size D (measured at the surface) of 40, 45 and 70 nm is determined. Consequently PM/DO values are 26, 21 and 12. From another low (#2) and high (#6) coercivity film we have measured Po (see above) and using these values we came to the conclusion that one domain consists of 4-6 columns for these films. We also have to realize that the column size is a function of the layer thickness. Studying the fractional surfaces it can be seen that the three films show a typical fractured plane close to the substrate. This is interpreted as belonging to the initial layer which is also seen in the in-plane VM graphs of the same samples.

**Coercivity**

The coercivity in ferromagnetic layers with perpendicular anisotropy is determined by two factors.

1. The type of magnetization process involved i.e. is the magnetization reversal by (coherent) rotation or by domain-wall movement?

2. Morphology of the layer i.e. shape and dimensions of the crystallites (columns), nature of the boundaries, surface and initial layer properties.

In fact these two factors are strongly interrelated and their properties are strongly influenced by the deposition parameters such as temperature, Par etc. Based on general knowledge about ferromagnetic films we can conclude that the presence of stripe domains in CoCr must be quite common because the layer splits up into a certain configuration by reducing its demagnetizing energy. Theoretical support for reversed domains in uniaxial thin layers with perpendicular anisotropy is provided by [11,16]. From research on permanent magnets, supporting arguments for the relation coercivity and reversal mechanism are given by [17]. In polycrystalline and/or columnar-like materials the critical domain size is difficult to estimate because it is influenced by the presence of neighbouring particles and their interactions. If the magnetization reversal takes place by coherent rotation Hc must be equal to the Hk. In our case all the layers have Hc<Hk. An explanation for this effect can be made by introducing the so-called fanning, curling and beading nodes. These can in principle, only occur in layers in which the magnetic anisotropy is mainly produced by shape anisotropy. In CoCr the main anisotropy is configurated by magneto crystalline anisotropy. On the other hand experimental work on high coercivity Co or Co-P layers [10-28] shows domain structures independent of the sources of the magnetic anisotropy. The presence of a domain structure and the results of other measurements like hysteresis loss and Hc vs. angle are not completely in line. This could perhaps be explained by the phenomena of interaction domains which were first introduced by [21]. Here particle exchange forces are negligible and the magnetic moments are mutually aligned by magnetoelastic interactions [22]. Recently this model was mentioned as a possible mechanism in CoCr layers [23]. For low Hc the nucleation density is much lower than for high Hc ones. It is clearly shown that nuclei stripe out at the tips of the domain after decreasing the applied field [13,14]. The observable size of the dots (coming from saturation) in high Hc films is determined by the resolving power of the Kerr microscope. Nevertheless these dots do not expand very much due to the very close surrounding of neighbouring dots. The reversal in one column could be, depending on its dimensions, initiated by a reversed domain. These small sizes cannot be resolved by Kerr microscopy and it can be accepted that such domains are no longer present in thinned TEM specimens. The particle behaviour without exchange coupling is suggested by Cr segregation at the column boundary. At the moment only preliminary results are available from Scanning Auger Microscopy (SAM) and wet chemical etch procedures. From the latter we found only qualitatively a higher Cr content at the surface than in the bulk of all our samples. From a SAM measurement of a fractional surface of a high coercivity film we came to the conclusion that there is a Cr segregation at the column boundaries.

**Conclusions**

Different domain configurations can be expected as a function of the coercivity. Factors like stress and magnetic history play an important role. For low Hc material the stripe domain theory is in good agreement with the measured P values from Kerr observations and the hysteresis loop. One remanent domain (dot, stripe) of all our films consists of several columns.

**References**