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Current perspectives

Anisotropy dispersion in $(CoCrPt)_{1-x}(SiO_2)_x$ perpendicular recording media

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ABSTRACT

The distribution of easy axis orientation in perpendicular media is of technological importance because it affects the value of S^* (see Fig. 1), which quantifies the switching field distribution (SFD) and hence partially determines the data density achievable on a given medium. The distribution is controlled by the crystallographic orientation of grains and factors such as intergranular exchange and dipolar coupling. Due to strong demagnetising fields in the perpendicular orientation, traditional measurements of remanence as a function of angle are difficult to interpret and have required the use of large-scale computational models. In this work we have utilised the variation of coercivity H_C with angle, which has the advantage that at H_C the global demagnetising field is zero. Additionally, since such materials follow essentially the Stoner–Wohlfarth mode of reversal, the variation of H_C with angle, $H_C(\theta)$, is much greater than that for the remanence. We find that for $(CoCrPt)_{1-x}(SiO_2)_x$, where the level of exchange coupling is controlled, the distribution of magnetic easy axes is narrower when the exchange coupling is reduced, but dipolar coupling between the grains is strong and affects the magnetisation reversal significantly.

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1. Introduction

The recent paradigm shift in magnetic information storage from longitudinal to perpendicular orientation of recorded bits has required the developments of new measurement techniques to characterise the storage medium. This is due to the large demagnetising field H_D from the film, $H_D = -4\pi M_S$ [1]. Using this value for H_D the loops are generally overcorrected due to the exchange coupling between the grains. Generally in perpendicular recording media a high moment, magnetically soft underlayer (SUL) is grown beneath the recording layer but separated from it by an exchange break layer. The SUL is generally circa 20 nm thick and the exchange break layer, which prevents noise in the signals, is circa 5 nm or chromium of a chromium alloy. In this case the 20 nm SUL of high-magnetisation material generates a magnetic moment, which swamps that from the recording layer. The solution to these difficulties is the application of polar MOKE magnetrometry, but unfortunately rotational experiments are not possible [2]. This means that measurements of the anisotropy axis dispersion in the medium cannot be made using standard techniques such as the variation of remanence with angle [3].

In perpendicular media the dispersion of the magnetic easy axes of the grains is of critical importance as small deviations from perfect orientation lead to large variations in the switching fields, hence generating a broader switching field distribution (SFD), which widens the bit-to-bit transition region, lowering data density. Perpendicular media are highly oriented single-domain systems and the reversal closely follows Stoner–Wohlfarth behaviour [4]. Assuming that there are no demagnetising effects, coherent reversal of spins in the aligned case occurs at the anisotropy field H_K at $T = 0$. For any deviation from the perfectly aligned case, a torque is exerted on the moment by the field. This leads to a rotation of the moment until the point at which the magnetostatic energy from the field overcomes the anisotropy energy and the moment reverses. This has a dramatic effect on the switching field and for example, a deviation from perfect alignment by 10° gives rise to a reduction in the switching field of 30% [4]. Of course in a real material at finite temperatures thermal fluctuations always result in the presence of a torque and hence naturally broaden the SFD.

In the recent past there have been a number of attempts to measure both the physical and magnetic easy axis dispersion from X-ray studies and magnetic measurements. Wu et al. [5] used

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measurements of transverse susceptibility to determine the magnetic easy axis dispersion modelled using a micromagnetic simulation. These were compared with X-ray rocking curves and it was found that the dispersion determined magnetically was narrower than the physical distribution. This was attributed to stacking faults in grains of the film. However, the transverse susceptibility technique is not widely available and Voznyuk et al. [6] obtained the magnetic dispersion via measurements of remanence and again used computer modelling to deconvolve the effects of the exchange interaction and demagnetising fields. A similar technique involving iterative correction for the demagnetising effect was undertaken by Wu et al. [7], who determined hysteresis loops with an approximately corrected total internal field.

There are clearly experimental difficulties and theoretical limitations in attempts to correct for the demagnetising field. It is possible to correct only for the global field, whereas local variations in the demagnetising field, as regions are aligned either up or down, make this a first approximation. Furthermore, the presence of strong intergranular exchange coupling invariably distorts the applicable value of the demagnetising field, necessitating the use of computer models. All models of magnetic systems are limited by the boundary conditions and assumptions that are used in the calculations. These include features such as: a uniform exchange coupling between the grains, single values for anisotropy constants, etc. In particular, measurements of remanence as a function of angle are prone to significant variations because of demagnetising effects. Also, due to the high degree of orientation and the intergranular exchange coupling, it is found that the variation of remanence with angle is relatively small, leading to large errors in the measurements.

In this work, we have used the nature of the Stoner–Wohlfarth model in an attempt to overcome some of these difficulties. Rather than measure the remanence we have measured the variation of coercivity with angle. This has the advantage that the global demagnetising field is then zero and the variation of the coercivity with angle is significantly greater. However, it must be born in mind that careful control over experimental conditions is required due to the nature of magnetic viscosity in all ferromagnetic materials, which has a peak at the remanent coercivity [8].

2. Experimental studies

In this work we have studied a set of films consisting of $(CoCrPt)_{1-x}$ $(SiO_2)_x$ with varying levels of SiO_2 [9]. $Co_{85}Cr_{10}$, Pt_5 and SiO_2 were cosputtered onto 2.5-in glass discs to form the (CoCrPt)–(SiO_2) media using UHV-magnetron sputtering. The percentage of SiO_2 on CoPtCr was varied between 19% and 34% by controlling the deposition rate. The structure of the resulting media was glass/Pt(10)/Ru(20)/[$(CoCrPt)_{1-x}(SiO_2)_x$](10 nm). By varying the levels of SiO_2 we are able to evaluate the effect of the exchange coupling on the distribution of the anisotropy axes. The films were prepared without the presence of an SUL, which enabled the use of conventional magnetometry techniques to obtain the measurements. The measurements were made at room temperature using an ADE Model 10 VSM with a maximum applied field of 2 T and a noise base of around 1×10^{-6} emu. The fully automated, high resolution magnet rotation capability of this system made it ideal for these types of measurement.

Hysteresis loops were measured by varying the angle, θ, between the applied field and the normal to the film surface. We were able to control the angle of field to the sample reproducibly to a resolution of 1°. The increase of SiO_2 content reduces the intergranular RKKY coupling, lowering the degree of cooperative reversal and increasing H_C. The hysteresis loops and derived parameters are shown in Figs. 1 and 2 for $\theta = 0$. The SFD is quantified via the parameter S^* shown for one sample in Fig. 1. Note that S^* increases as the SFD narrows since it reflects the squareness of the loop. In practice S^* is an engineering parameter that provides a guide to likely recording performance via the Williams and Comstock [9] model. The origin of S^* in terms of material parameters involves a complex combination of grain size distribution, anisotropy distribution and easy axis orientation. Hence the variation of S^* with SiO_2 concentration shown in Fig. 2 is somewhat erratic.

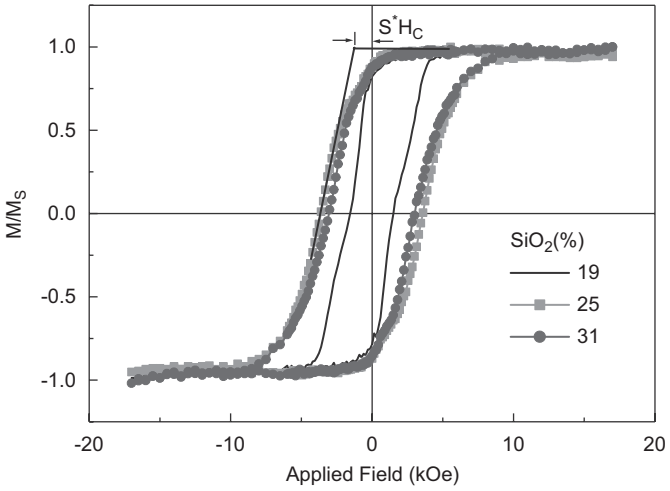

Fig. 1. Hysteresis loops measured with the applied field perpendicular to the film. Only three loops are shown for clarity.

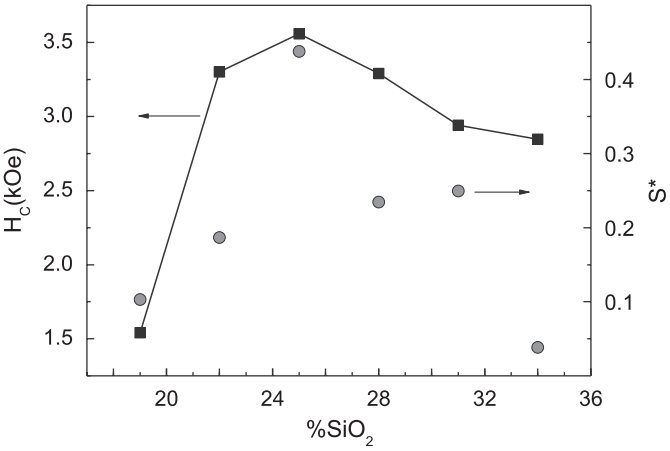

Fig. 2. Variation of coercivity with SiO_2 content measured with the applied field perpendicular to the film.

Above 25% SiO_2 the size of the grains reduces, hence reducing the energy barriers to reversal, giving a reduction of H_C [10]. The reduction in the slope of the hysteresis loops at H_C reflects the weakening of the intergranular exchange coupling with increasing SiO_2. The role of exchange coupling on the hysteresis loop properties is complex. Initially the coupling keeps grains aligned parallel, increasing the squareness. However, a reversal by a single grain, perhaps having low anisotropy, leads to cooperative reversal as the grains are coupled. This increases the squareness of the loop (S^*), implying a narrower SFD, but lowers H_C as the reversal is dominated by grains having the lowest energy barriers. Also exchange coupling will align the moments of the grains and maintain them parallel at any angle close to the easy axis until the

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