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Full Length Article

Tuning the static and dynamic magnetic properties of c-axis oriented *hcp*-(CoIr) thin films by the addition of Cr

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ABSTRACT

In the present work, we have investigated both static and dynamic magnetic properties of c-axis oriented *hcp*-$(CoIr)_{100-x}Cr_x$ soft magnetic thin films in great detail. We have found that the coercivity decreases much faster than the saturation magnetization with the increasing of Cr content in favor of better soft magnetic properties. Moreover, the in-plane and out-of-plane magnetic anisotropy can be greatly changed as a function of x. Therefore, the microwave magnetic properties can be tuned according to the requirements of different applications.

1. Introduction

Soft magnetic thin films (SMTFs) have a wide range of applications, such as film inductors [1,2], micro-transformers [3] and microwave noise filters [4,5], etc. In general, high saturation magnetization, M_s, is required to get high initial permeability, μ_i, which is prerequisite for microwave applications. In the past, Fe- and Co-based SMTFs with very high M_s, in forms of amorphous and nanocrystalline types, have been studied extensively and proved to have excellent static and microwave properties [6,7]. Magnetic anisotropy is another key factor for the SMTFs to possess good enough soft magnetic properties for practical applications. For SMTFs, their microwave properties are governed by the following equation in the bianisotropy model [8]

$$f_r = \left(\frac{\gamma}{2\pi}\right)\sqrt{(4\pi M_s - H_{grain})\cdot H_u}$$
$$\mu_i = 1 + \frac{4\pi M_s}{H_u} \tag{1}$$

where μ_i, f_r are initial permeability and natural resonance frequency, respectively. γ is the gyromagnetic ratio. $H_{grain} = 2K_{grain}/M_s$ is the effective magnetocrystalline anisotropy field with K_{grain} being the corresponding magnetocrystalline anisotropy constant. $H_u = 2K_u/M_s$ is the effective in-plane uniaxial anisotropy field with K_u being the corresponding anisotropy constant. For traditional Fe- or Co-based SMTFs, H_{grain} is close to zero. Therefore, one can not increase f_r, by increasing H_u, without decreasing μ_i due to the so-called Acher limit [9].

To overcome the above limit, from Eq. (1), it is clear that only negative values of K_{grain} can increase f_r without affecting μ_i. In fact, in our recent work [10–15] we have already shown that $Co_{1-x}Ir_x$ based SMTFs with negative K_{grain} have superior magnetic properties for future applications. For example, the critical film thickness for the Néel type domain wall changing into Bloch type has been greatly enhanced [12] and micrometer thick SMTFs with its magnetic moments restricted strictly in-plane can be grown in favor of practical use [15]. However, some properties of these films still need to be improved before they can be employed in real electronic devices. For example, these films still have coercivities, H_c, in the range of several mili-tesla which is unfavorable for their application in actual devices. Therefore, further reducing the coercivity and adjusting the anisotropy property of these films become utmost important now.

Usually, the H_c of SMTFs can be reduced by fine tuning the growth conditions or post-annealing process which shall minimize the defects and/or internal residual stress of the final films [16–20]. Alternatively, H_c can be adjusted by adding a third component during the film growth procedure [21–25]. For the mostly investigated Fe/Co- system, the usually used doping elements include B [26,27], Si/SiO_2 [24,25], or simply Fe/Co-oxide [22]. These doped elements or oxides usually reside at the interstitial site or fill at the interspace between individual grains which usually leads to refined crystallite grains and optimized exchange coupling between different grains. However, these doping usually result in a rapid decrease of M_s due to the nonmagnetic nature of the dopants. Therefore, it is better to choose a 3*d* transition metal as the doping element for our *hcp*-CoIr film. Cr has a finite magnetic moment and not too close to the Co in the periodic table. Apart from the interspace between different grains, we also expect that the Cr atom might enter into the crystal structure and tune the intrinsic magnetocrystalline anisotropy of our *hcp*-CoIr film. Therefore, in this work, we present our study of *hcp*-$(CoIr)_{100-x}Cr_x$ films. Microstructure, magnetic anisotropy

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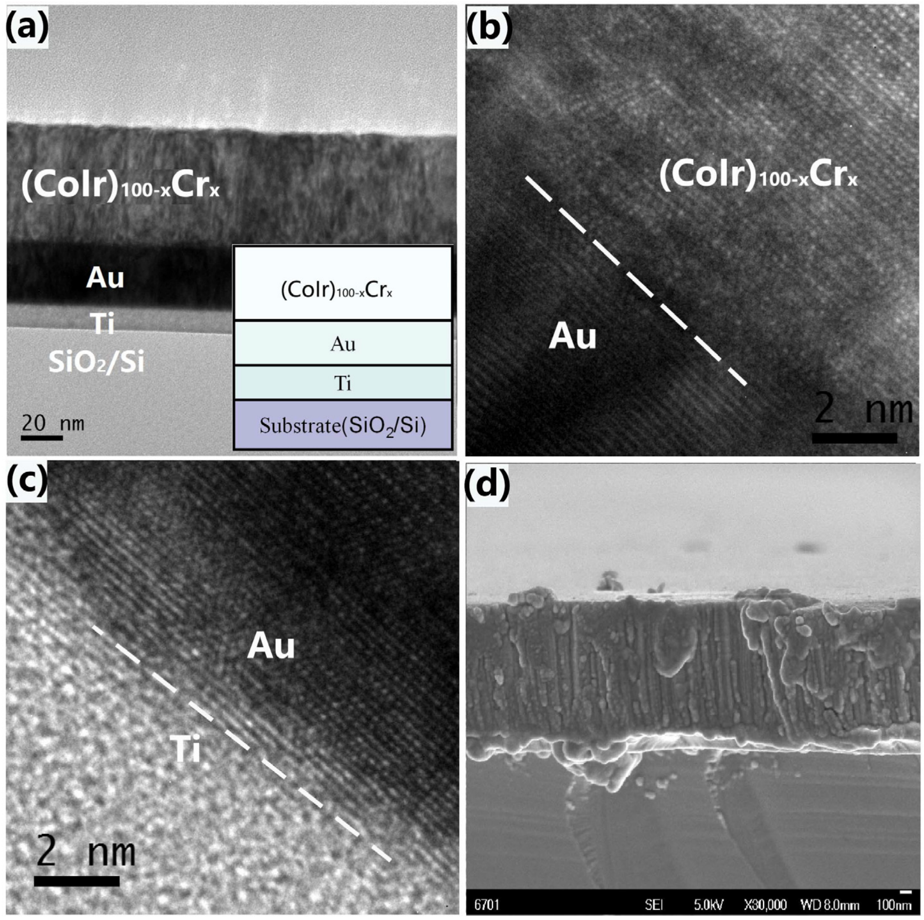

Fig. 1. Typical TEM image of our thin film system showing the total cross section of the layered structure in (a), the c-axis oriented growth of $(CoIr)_{100-x}Cr_x$ on the c-axis oriented Au-seed layer in (b) and the oriented growth of the Au-seed layer on top of the amorphus Ti layer in (c). (d) Typical SEM image of the cross section of a 950 nm thick film with x = 0 which, together with the TEM image in (a), show that the columnar type growth of our $(CoIr)_{100-x}Cr_x$ thin film can survive to very thick films. Inset in (a) shows the schematic layer structure of our sample.

and exchange interaction between the columnar type grains and their effects on the coercivity reduction have been investigated extensively as a function of Cr content x. We have found that by adding the third element Cr, the soft magnetic properties of our CoIr-film can be greatly tuned for future applications.

2. Experiment

All magnetic thin films studied in this work were prepared by DC magnetron sputtering technique similar to our earlier work [15] with a layered structure of substrate/Ti/Au/$(CoIr)_{100-x}Cr_x$ as shown in Fig. 1(a). Si wafer with (1 0 0) surface orientation was used as substrate, and the seed layer of Ti(8 nm)/Au(25 nm) was deposited first in order to induce the c-axis orientation of the *hcp*-$(CoIr)_{100-x}Cr_x$ magnetic soft layer. The amorphous Ti layer provides a clean and flat surface for the (1 1 1)-plane oriented Au layer to grow. As Au has a cubic crystal structure with lattice constant a = 4.07 Å, then the lattice spacing within the (1 1 1)-plane amounts to about 2.88 Å. This value is about 12% larger than the lattice spacing of *hcp*-(CoIr) with a~2.58 Å which means that the obtained c-axis oriented *hcp*-(CoIr) film should be under tensile stress. For the growth of the seed layer, pure Ar with a pressure of 0.25 Pa was used as the sputtering working gas. As for the growth of the magnetic layer, 0.3 Pa Ar gas was used. The soft magnetic layer was deposited using a Co target with certain Ir and Cr chips symmetrically placed on top of it. The Cr content can be changed by changing the distance between the Cr chips and the erosion race-track of the Co target while keeping the position of the Ir chips unchanged. In order to obtain films with better high frequency properties, an in-plane uniaxial anisotropy was induced by inclining the substrate with an angle of ~ 5^o against the sputtering target during the whole sputtering process for all the investigated films.

Chemical compostion of the prepared thin films have been characterized by using energy dispersive spectrometer (EDS) equipped within a scanning electron microscope (SEM). The actual composition was determined to be $(Co_{84}Ir_{16})_{100-x}Cr_x$. For simplicity, we will use $(CoIr)_{100-x}Cr_x$ to denote all our samples in the following text. Grain morphology of the film was measured using transmission electron microscope (TEM) and crystal structure of our sample was characterized by X-ray diffraction (XRD) with Cu $K_{\alpha1}$ radiation. Static magnetic properties of our films were measured with a vibrating sample magnetometer (VSM) and dynamic magnetic properties were measured with an electron spin resonance spectrometer (ESR) to determine the intrinsic magnetocrystalline anisotropy constant of our films [15]. For the microwave property, our films were check by a vector network analyzer (VNA) using the shorted microstrip method. All these measurements were performed at room temperature and the sample were stored under vacuum except the measurements procedure.

3. Results and discussion

Fig. 1(a) present the typical TEM image showing the layered structure of our film as described by the schematic diagram in the inset of Fig. 1(a). The successful growth of the c-axis oriented *hcp*-$(CoIr)_{100-x}Cr_x$ layer and the Au-seed layer can be clearly seen from