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TEM-based study of oxidation of FeCo high moment alloy at elevated temperatures

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ABSTRACT

FeCo (60%-70% Fe, 30%-40% Co at.%) alloys have been used as write transducer materials due to their large saturation magnetization at room temperature ($\mu_0 M = 2.45$ T). In this research, we investigate the influence of oxidation on the magnetic moment of a $Fe_{70}Co_{30}$ material at elevated temperatures in the application of heat assisted magnetic recording. We find that the moment of Fe70Co30 decreases after oxidation for all temperatures explored and that Co does not take part in the oxidation process. The moment loss is mainly due to the formation of ferrimagnetic Fe₃O₄ or γ-Fe₂O₃ as well as the change of the Fe/Co ratio in the FeCo alloy. A 2 nm Cr underlayer has been employed to change the FeCo texture from (110) to (002). This also increases the grain size of the FeCo thin film. In the case of the (002) textured sample, voids form between the FeCo film and the oxide layer, which could be the direct result of Fe cations diffusion.

1. Introduction

In heat assisted magnetic recording (HAMR), a ferromagnetic write transducer is placed in close proximity to a metal-based plasmonic nearfield transducer (NFT), especially near the air-bearing surface (ABS) [1-3]. A thin diamond-like-carbon (DLC) layer protects the pole-tip of the magnetic transducer from direct air exposure at the ABS. However, heat generated by the NFT often leads to DLC damage and leaves the magnetic pole-tip exposed to air. In such cases, oxidation of the magnetic pole tip becomes a likely event, especially when the surrounding temperature becomes significantly elevated.

The magnetic write transducer, or write head, is usually made of high moment FeCo (Fe 60%-70%, Co 30%-40% at.%) alloys with saturation magnetization around $\mu_0 M_S = 2.45T$ at room temperature. Oxidation of these alloys at elevated temperatures may have a potentially negative effect on the FeCo alloy's magnetic properties, which would lead to the degradation of write head performance. Although oxidation of FeCo nanoparticles has been reported previously [4,5], the exploration of thin film oxidation of FeCo alloys is still limited.

In the first part of this paper, we examine the influence of oxidation on Fe₇₀Co₃₀ thin films with different exposure temperatures and exposure times. X-ray diffraction (XRD), Transmission Electron Microscopy (TEM) images as well as magnetic properties are presented to display the properties and structure of oxidized and unoxidized samples. In the second part of the paper, we discuss the effects of inserting a very thin Cr underlayer to change the texture of FeCo from (110) to (002). By fixing the exposure temperature at 240 $^\circ\text{C}\textsc{,}$ comparison between the samples with/without a Cr underlayer is illustrated. Chemical analysis is also performed to achieve an in-depth understanding of the O2 influence on samples with/without a Cr underlayer.

2. Experimental procedure

Fe₇₀Co₃₀ thin films were deposited at room temperature on Corning 7059 glass substrates using RF sputtering in a Leybold Heraeus Z-400 sputtering system. The thickness of the as-deposited Fe₇₀Co₃₀ films was 100 nm. After deposition, the samples were heated in the chamber without breaking the vacuum. The heating curve with respect to the time and current source was calibrated. In situ oxidation was performed

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Fig. 1. XRD patterns for (a) as deposited $Fe_{70}Co_{30}$ (b) $Fe_{70}Co_{30}$ with O_2 exposure at 120 °C (c) $Fe_{70}Co_{30}$ with O_2 exposure at 240 °C (d) $Fe_{70}Co_{30}$ with O_2 exposure at 370 °C.

at 120 °C, 240 °C and 370 °C after the sample temperature became stable. The O₂ pressure was 1.8 mTorr with the O₂ flow rate being 5 SCCM (Standard Cubic Centimeters per Minute). The oxidation time varied from 1 to 90 min for each temperature. For samples exposed at 240 °C, a 2 nm thick Cr underlayer was deposited before the deposition of FeCo using RF sputtering.

The crystallographic characterization of the sample was obtained from the Panalytical X'Pert Pro MPD X-Ray Diffractometer using Cu K_α radiation with 2 θ scan from 25° to 90°. Magnetization vs field curves were measured by a vibrating sample magnetometer (SQUID-VSM) from -7 T to 7 T. The microstructure of the samples with and without oxidation was observed using a FEI Tecnai F20 Transmission Electron Microscope. Chemical analysis was done using the TEM Energy-Dispersive X-ray spectroscopy (EDX) mode. Depth profiling was performed by X-ray photoelectron spectroscopy (XPS) with 3 keV Ar⁺ sputtering and iteration, in which Al K_α radiation source with photoelectron energy of 1486.6 eV was used, while Fe 2p, Co 2p and O 1 s peaks were used for composition quantification. During depth profiling, a square etching area of 1 mm² was used and the X-ray spot was centered on a 200 µm diameter region inside the etching area to minimize the crater-edge effect.

3. Results and discussion

3.1. Composition and crystal structure analysis

Fig. 1 shows the XRD scan for as-deposited $Fe_{70}Co_{30}$ samples as well as for samples with O_2 exposure at 120 °C, 240 °C and 370 °C. For the

100 nm Fe₇₀Co₃₀ thin film directly deposited on the substrate, the XRD shows a good FeCo (110) texture indicating that Fe₇₀Co₃₀ has either the BCC or the B2 structure with a lattice constant around 2.86 Å (Fig. 1(a)) [1]. The bump between 25° and 35° is mainly due to the glass substrate. For samples prepared at 120 °C with O₂ exposure, only the FeCo (110) peak is observed (Fig. 1(b)), which is due to poor diffusivity of oxygen in FeCo at low temperatures. At 240 °C with O2 exposure, an Fe-oxide (311) peak is observed with the increase of exposure time (Fig. 1(c)), which could be attributed to Fe_3O_4 ($2\theta_{(311)} = 35.42^\circ$) or γ -Fe₂O₃ $(2\theta_{(311)} = 35.61^{\circ})$. The two oxides have the same FCC Bravias lattice and similar lattice constants ($a_{Fe3O4} = 8.3941$ Å, $a_{\gamma-Fe2O3} = 8.352$ Å) [6] so it is difficult to differentiate them using XRD. At 370 °C with O2 exposure, however, there is hardly any FeCo (110) peak left. Instead, the oxide (311) peak becomes dominant in this situation together with some other oxide peaks, which demonstrates that the original FeCo was completely oxidized at these temperature and exposure conditions. It's worth noting that there is no FeO or Co based oxide at the three exposure temperatures, which corresponds to the predictions from the Ellingham diagram [7].

3.2. Microstructural analysis

Fig. 2 shows the TEM bright field image for as-deposited $Fe_{70}Co_{30}$ film as well as those for samples oxidized at 120 °C, 240 °C and 370 °C. The as-deposited $Fe_{70}Co_{30}$ film shows large grain size with grain diameters around 20–30 nm. The TEM bright field image also shows a native oxide layer on top of the FeCo layer with a thickness less than 2 nm (Fig. 2 (a)). The initial oxidation layer could be the result of a contact



Fig. 2. TEM images for as deposited $Fe_{70}Co_{30}$ and $Fe_{70}Co_{30}$ samples with O_2 exposure at different temperatures. (a) Bright field TEM for as deposited $Fe_{70}Co_{30}$ film. (b) Bright field TEM for $Fe_{70}Co_{30}$ with $120 \degree C O_2$ exposure for 30 min. (c-e) Bright field TEM for $Fe_{70}Co_{30}$ with $240\degree C O_2$ exposure for 1, 30, 90 min, respectively. (f) Bright field TEM for $Fe_{70}Co_{30}$ with $370\degree C O_2$ exposure for 30 min. The scale bar in (a-e) is 50 nm, scale bar in (f) is 100 nm.

potential difference between the metal and the absorbed oxygen, enabling the outward diffusion of metal ions and leading to an oxide layer less than a few nanometers as described by Cabrera-Mott theory [8,9]. For samples at 120 °C oxidation, there was not much oxidation compared with the as-deposited sample (Fig. 2 (b)). Even if we increased the exposure time from 30 min to 90 min, the thickness of the oxide layer did not change. During 240 °C exposure, however, even 1 min exposure resulted in a 30 nm thick oxide layer and a rough surface (Fig. 2 (c)). With increasing exposure time, the thickness of the oxide grew from 47 nm to 55 nm and the interdiffusion between the FeCo layer and the substrate occurred (Fig. 2 (d) and (e)). For 370 °C exposure, nearly complete oxidation occurred even after only 30 min of O2 exposure with an oxide thickness of more than 150 nm. In addition, voids formed across the boundary between the oxide layer and the remaining FeCo layer (Fig. 2 (f)). The void formation mechanism can be explained by the Smigelskas Kirkendal effect [10,11] as follows: the relative high temperature provides enough thermal energy for outward bound diffusion of Fe cations, while the promotion of the inward bound diffusion of large oxygen anions is not as easy. As a consequence, the amount of outward Fe cations greatly exceeds the amount of inward oxygen anions, resulting in the formation of vacancies between oxide and FeCo layer. The accumulation of vacancies can in turn promote the formation of the void volumes. The TEM images provide a proof that exposure temperature is a dominant factor determining the oxidation of Fe₇₀Co₃₀ thin films. Also, FeCo thin films can be oxidized more easily than the FeCo nanoparticles discussed by K. N. Collier et al., [4]. FeCo nanoparticles do not have grain boundaries and the surface is covered with a thin passivation oxide layer. On the other hand, oxygen could diffuse along the grain boundary of the FeCo thin film, making it easier to be oxidized.

3.3. Magnetization and coercivity measurements

The M-H loops for all samples were measured using a vibrating sample magnetometer (SQUID-VSM) for both the in plane and the out of plane directions of the films. The applied fields varied between -7 T to 7 T and the magnetization was normalized by the sample area with the substrate signal subtracted from the measurement. Fig. 3 shows the M-H loop for as deposited $Fe_{70}Co_{30}$ and $Fe_{70}Co_{30}$ oxidized at 120 °C, 240 °C and 370 °C for 90 min, respectively. The Fe70Co30 oxidized at 120 °C (Fig. 3 (b)) and 240 °C (Fig. 3 (c)) show similar M-H loops compared with the as-deposited Fe₇₀Co₃₀ (Fig. 3(a)), indicating the overall magnetic performance is still governed by the unoxidized Fe70Co30 layer. M-H loop for the Fe70Co30 oxidized at 370 °C (Fig. 3(d)) is much different because the dominant phase in the system becomes oxide as indicated in the TEM image. Fig. 4 shows the change of the magnetization at 7 T (Fig. 4 (a)) and the in plane coercivity (Fig. 4 (b)) with respect to the oxidation time and temperature. It was found that the as deposited Fe70Co30 film has a saturation magnetization of 2.2 T when the film plane was the easy plane (Fig. 3 (a)). The saturation magnetization is smaller than the literature bulk value of 2.45 T, which is probably the result of finite size effects [1]. The exchange bias (Fig. 3 (a) inset) observed is due to the native ferrimagnetic oxide layer on top of FeCo [6]. For samples



Fig. 3. Magnetic properties for $Fe_{70}Co_{30}$ with/without O_2 exposure. The insets in (a), (b) and (c) are the in-plane M–H loop with small range of field. (a) M–H loop for as deposited $Fe_{70}Co_{30}$. (b) M–H loop for $Fe_{70}Co_{30}$ oxidized at 120 °C for 90 min. (c) M–H loop for $Fe_{70}Co_{30}$ oxidized at 240 °C for 90 min. (d) M–H loop for $Fe_{70}Co_{30}$ oxidized at 370 °C for 90 min.



Fig. 4. (a) Magnetization at 7 T vs exposure time for $Fe_{70}Co_{30}$ without/with oxidation at different exposure temperature. (b) In plane coercivity vs exposure time for $Fe_{70}Co_{30}$ without/with oxidation at different exposure temperature. The area denoted by dashed rectangle is amplified by the inserted graph.



Fig. 5. XRD pattern for $Cr/Fe_{70}Co_{30}$ film without/with O_2 exposure at 240 °C.

oxidized at 120 °C, though we did not observe a thick oxide layer from TEM image, an obvious moment decrease is found for this set of samples. For samples oxidized at 240 °C and 370 °C, the moment drop is even more obvious compared with samples oxidized at 120 °C (Fig. 4 (a)). The moment drop of FeCo is mainly due to the formation of ferrimagnetic Fe oxide, which cancels out some of Fe moments. Also, the diffusion of Fe cations from FeCo to Fe oxide can contribute to the composition change of the original FeCo film, which according to the Slater Pauling curve, decreases the saturation magnetization for the remaining $Fe_{70-x}Co_{30+x}$ film. It is also worth noting that samples with the oxide layer may not be fully saturated magnetically even at 7 T, [12–14] which is also shown in Fig. 3 (d) since the magnetization is still slowly increasing as the magnetization at 7 T and averaged the in plane and out of plane magnetization to diminish the sample size effect.

Fig. 4 (b) shows the in plane coercivity for this set of samples. It is observed that for samples oxidized at 120 °C and 240 °C, the coercivity can be lower than that of the as-deposited $Fe_{70}Co_{30}$ layer while for samples exposed at 370 °C, the coercivity is much larger than that of the

as-deposited sample. Since the unoxidized FeCo still dominates the magnetic behavior for the samples oxidized at 120 °C and 240 °C, the ordering of FeCo layer during the heat treatment in the oxidation process reduces the coercivity [15]. Samples oxidized at 240 °C show lower coercivity than samples oxidized at 120 °C due to the temperature increase. For samples oxidized at 370 °C, the phenomenon can be partially explained by Hoffmann's ripple theory [16,17] as shown below:

$$H_C = \frac{\beta \sqrt{2} \sqrt{DS}}{\pi L M_S} \left(\frac{K_U}{A}\right)^{1/4}$$

where β is a fitting factor, D is the average grain size, S is a structural factor proportional to grain size. L is the coupling length, M_S is the saturation magnetization, A is the exchange constant and K_U is the magnetocrystalline anisotropy constant. For samples oxidized at 370 °C, the reduction of the saturation magnetization is obvious but on the other hand, the dominant phase for 370 °C oxidized samples is mainly nanosized Fe₃O₄ (or some amount of γ -Fe₂O₃), whose magnetocrystalline anisotropy constant could be one order larger than the bulk value of $K_1 = 0.135 \times 10^5 J/m^3$ [18]. Compared with the magnetocrystalline anisotropy constant for FeCo reported as $K_1 = 15KJ/m^3$ for Fe 65% Co 35% [19], the increase in K_U would also result in the increase of coercivity. In addition to the above explanation, some other factors such as the defects and the crystalline quality of the film may also be taken into account to fully understand the change of coercivity.

3.4. Oxidation of Cr/Fe₇₀Co₃₀ with (002) FeCo texture

We also performed similar experiments on FeCo thin films that were deposited on a 2 nm Cr underlayer. For samples with oxidation, the exposure temperature was fixed at 240 °C. The XRD data and TEM images for this set of samples are presented below. The differences between magnetization and coercivity for samples with/without the Cr underlayer were compared and analyzed. The XRD data (Fig. 5) shows that the texture of $Fe_{70}Co_{30}$ changes from the (110) to (002) of BCC or B2 lattices with the Cr underlayer. Also, with the increase of exposure time, different oxide peaks begin to appear, which could be attributed to an oxide (400) peak. The reason why FeCo (002) promotes oxide (400) is due to the epitaxial growth of oxide on FeCo across (001) planes [6]. It is worth noting that the FeCo (002) peak slightly shifts to the right after the oxidation, indicating a lattice constant decrease from 0.2875 nm to



Fig. 6. (a) Bright field TEM image for $Cr/Fe_{70}Co_{30}$ film without O_2 exposure (b) Bright field TEM image for $Cr/Fe_{70}Co_{30}$ with 90 min O_2 exposure at 240 °C (c) High resolution TEM image for void area in (b). The scale bar in (a) and (b) is 100 nm, scale bar in (c) is 10 nm.



Fig. 7. (a) Magnetization at 7 T for $Fe_{70}Co_{30}$ with/without a Cr underlayer (b) In plane coercivity for $Fe_{70}Co_{30}$ with/without a Cr underlayer, label 1 is the as deposited $Fe_{70}Co_{30}$ without oxidation and label 2 is the $Fe_{70}Co_{30}$ with 1 min oxidation at 240 °C.



Fig. 8. Plane-view TEM image for (a) Fe₇₀Co₃₀ without a Cr underlayer and (b) Fe₇₀Co₃₀ with a Cr underlayer.

0.2852 nm. Since we preheated the sample before applying the oxygen exposure, the crystalline quality of the FeCo would improve (less vacancies/dislocations), resulting in the change of the lattice constant. It is also possible that the slower ordering of FeCo under the heating changed the FeCo structure from BCC to B2, which reduced the lattice constant. A similar effect is also observed for the FeCo film without the Cr underlayer.

Fig. 6 shows the microstructure of $Cr/Fe_{70}Co_{30}$ without O_2 exposure (Fig. 6 (a)) and $Cr/Fe_{70}Co_{30}$ with 90 min O_2 exposure at 240 °C (Fig. 6 (b) and (c)). For the sample without O_2 exposure, the bright field TEM image shows FeCo with large grains (Fig. 6 (a) and Fig. 8 (b)), which is different from the observations of Xiaoxi Liu *et al.*, [19]. Similarly, a native oxide layer formed on top of FeCo with thickness less than 2 nm. For samples with 90 min oxidation at 240 °C, the TEM image shows some void volumes as denoted by the red circles in Fig. 6 (b). Fig. 6 (c) shows that the actual void volumes are between the FeCo layer and the oxide layer, as denoted in the purple circle. Neither a crystal structure or amorphous area could be observed in this region. The formation of voids in this case can be mainly attributed to the larger grain size, which results in less grain boundary density. With less grain boundary density, it is harder for large oxygen ions to move inward through the film while the diffusion of Fe ions outward dominates, resulting in the formation of

voids. The void formation mechanism for these thin films is similar to the case for Fe nanoparticle oxidation, which is a direct result of the Fe outward diffusion [4].

The comparison of the magnetization and coercivity between Fe₇₀Co₃₀ with/without Cr underlayer is shown in Fig. 7. For samples with O₂ exposure, the exposure temperature was fixed at 240 °C. It can be seen that with Cr underlayer, the magnetization at 7 T is larger than that of the samples without Cr underlayer (Fig. 7 (a)). Also, the in plane coercivity for samples with the Cr underlayer is larger than samples without the underlayer (Fig. 7 (b)). The two observations can both be explained by the change in the microstructure. As seen in the plane-view TEM images, the as-deposited FeCo thin film has fine polygon-shaped grains (Fig. 8 (a)) while for the FeCo thin film with the Cr underlayer, coarse grains are observed (Fig. 8 (b)), which results in the less grain boundary density and reduces the moment loss due to the surface moments. The difference between the microstructure could result from the change of the surface energy introduced by the Cr underlayer, which could change the nucleation density for the FeCo grains. In addition, according to Hoffmann's theory mentioned above, larger grains at this range of grain size, usually leads to higher coercivity, which explains the results in Fig. 7 (b).



Fig. 9. (a) STEM for Fe₇₀Co₃₀ film with 90 min oxidation at 240 °C (b-d) EDX mapping results for O, Fe and Co content for Fe₇₀Co₃₀ film with 90 min oxidation at 240 °C.



Fig. 10. (a) STEM for Cr/Fe₇₀Co₃₀ film with 90 min oxidation at 240 °C (b-e) EDX mapping results for O, Fe, Co and Cr content for Cr/Fe₇₀Co₃₀ film with 90 min oxidation at 240 °C.



Fig. 11. TEM EDX line scan for (a) Fe₇₀Co₃₀ (b) Cr/Fe₇₀Co₃₀ with 90 min oxidation at 240 °C and XPS results for (c) Fe₇₀Co₃₀ (d) Cr/Fe₇₀Co₃₀ with 90 min oxidation at 240 °C.

3.5. Chemical analysis

In this section, we report the chemical analysis that was performed on $Fe_{70}Co_{30}$ samples with/without Cr underlayer. Fig. 9 shows the STEM image as well as EDX mapping results for $Fe_{70}Co_{30}$ film with 90 min oxidation at 240 °C. The mapping results show that Co did not take part in the oxidation process. Also, a third oxide layer emerged between the FeCo layer and the substrate, which corresponds to the results in TEM bright field image. The formation of this oxide layer was mainly due to the reaction between Fe and alkali oxides in the Corning 7059 substrate. Fig. 10 shows the STEM image and EDX mapping results for Cr/Fe₇₀Co₃₀ film with 90 min oxidation at 240 °C. It is interesting to note that there are several "black holes" in the STEM image, while EDX results show there are hardly any elements in these areas. These "black holes" correspond to the voids mentioned above, which is the direct result of Fe outward diffusion.

Cross section TEM images only show a small region of the samples that were oxidized, for oxidation samples, however, the oxidation extent in the thickness direction can sometimes appear to be uneven. To better illustrate the oxidation process on a large scale, X-ray photoelectron spectroscopy (XPS) depth profiling was performed and results were compared with the TEM EDX line scan (Fig. 11). The boundary between the substrate and the film is roughly denoted by the dashed line expressed in Fig. 11 (c) and (d). This shows that XPS results are in good accordance with the TEM EDX line scan results. In addition, XPS results show that for $Fe_{70}Co_{30}$ sample, the atomic fraction of oxygen increases much even before reaching the substrate. However, for $Cr/Fe_{70}Co_{30}$ sample, the atomic fraction of oxygen only has a small increase before

reaching the substrate. It is worth noting that the increase could be the result of non-uniform sputtering and some extent of atom intermixing between layers caused by energetic ions, which caused a seemingly more gradual film-substrate boundary in XPS depth profiles in Fig. 11 than that shown in STEM-EDS in Fig. 9 and Fig. 10. This could also be due to the faint Cr signal, which overlaps with the shifted peak of oxygen when approaching the substrate. The results reveal that the Cr underlayer acted as a barrier layer which prevented the inter-diffusion between the Fe and the glass substrate. XPS also provides a solid evidence that the composition of FeCo gradually changes as we go deep into the film, which is another reason for the reduction of magnetization.

4. Conclusion

In this paper, the influence of oxidation on Fe₇₀Co₃₀ high moment material was investigated and discussed. We show that for all temperatures selected, the moment after oxidation decreases compared with the asdeposited sample. The temperature plays an important role in the oxidation: very little oxidation occurs at temperatures under 120 °C O₂ exposure while for 370 °C exposure, the samples are almost completely oxidized even after 30 min. The mechanism of moment reduction is mainly due to the formation of ferrimagnetic oxide (Fe₃O₄ or γ -Fe₂O₃). On the other hand, the change of Fe/Co ratio in oxidized samples is another reason for the reduction of sample moment. For all cases above, Co does not take part in the oxidation process. By inserting a thin Cr underlayer, the texture of Fe₇₀Co₃₀ as well as its grain size has been changed, the (002) texture Fe₇₀Co₃₀ film with larger grain size shows higher magnetic moment after O₂ exposure, which suggests better oxidation resistance behavior.

5. Data availability

The data that support the findings of this study are available from the corresponding author upon request.

CRediT authorship contribution statement

Tianxiang Du: Conceptualization, Methodology, Investigation, Writing - original draft. B.S.D.C.S. Varaprasad: Methodology, Investigation. Zhitao Guo: Investigation. Andrew J. Gellman: Writing review & editing. Jian-Gang (Jimmy) Zhu: Writing - review & editing. David E. Laughlin: Writing - review & editing.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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References

- [1] G. Scheunert, O. Heinonen, R. Hardeman, A. Lapicki, M. Gubbins, R.M. Bowman, A review of high magnetic moment thin films for microscale and nanotechnology applications, Appl. Phys. Rev. 3 (1) (2016) 011301, https://doi.org/10.1063/ 1.4941311.
- [2] D. Weller, O. Mosendz, G. Parker, S. Pisana, T.S. Santos, L10 FePtX–Y media for heat-assisted magnetic recording, Phys. Status Solidi A Appl. Mater. Sci. 210 (7) (2013) 1245–1260.
- [3] D. Weller, G. Parker, O. Mosendz, A. Lyberatos, D. Mitin, N.Y. Safonova, M. Albrecht, Review Article: FePt heat assisted magnetic recording media, J. Vac. Sci. Technol. B, Nanotechnol. Microelectronics: Mater., Proc., Measur. and Phenomena 34 (6) (2016) 060801, https://doi.org/10.1116/1.4965980.

- [4] K.N. Collier, N.J. Jones, K.J. Miller, Y.L. Qin, D.E. Laughlin, M.E. McHenry, Controlled oxidation of FeCo magnetic nanoparticles to produce faceted FeCo/ ferrite nanocomposites for rf heating applications, J. Appl. Phys. 105 (7) (2009) 07A328, https://doi.org/10.1063/1.3054376.
- [5] N.J. Jones, K.L. McNerny, A.T. Wise, M. Sorescu, M.E. McHenry, D.E. Laughlin, Observations of oxidation mechanisms and kinetics in faceted FeCo magnetic nanoparticles, J. Appl. Phys. 107 (9) (2010) 09A304, https://doi.org/10.1063/ 1.3334197.
- [6] N.J. Jones, A study of the oxidation of Fe1-xCox alloys and their resulting magnetic properties, Doctoral thesis, Carnegie Mellon University, Pittsburgh, Pennsylvania, 2011.
- [7] H.J.T. Ellingham, Reducibility of oxides and sulphides in metallurgical processes, J. Soc. Chem. Ind. (London) 63 (5) (1944) 125, https://doi.org/10.1002/ ictb.5000630501.
- [8] C.M. Wang, D.R. Baer, L.E. Thomas, J.E. Amonette, J. Antony, Y. Qiang, G. Duscher, Void formation during early stages of passivation: Initial oxidation of iron nanoparticles at room temperature, J. Appl. Phys. 98 (9) (2005) 094308, https://doi.org/10.1063/1.2130890.
- [9] N. Caberra and N. F. Mott, Rep. Prog. Phys. 12, 163 1948–1949.
- [10] A.D. Smigelskas, E.O. Kirkendall, Zinc diffusion in alpha brass, Trans. AIME 171 (1947) 130–142.
- [11] F. Seitz, On the porosity observed in the Kirkendall effect, Acta Metall. 1 (3) (1953) 355–369.
- [12] T. Hibma, F.C. Voogt, L. Niesen, P.A.A. van der Heijden, W.J.M. de Jonge, J.J.T. M. Donkers, P.J. van der Zaag, Anti-phase domains and magnetism in epitaxial magnetite layers, J. Appl. Phys. 85 (8) (1999) 5291–5293.
- [13] M. Watanabe, S. Abe, Out-of-plane magnetic moment and lattice distortion in sputtered Ge added Fe3O4 thin film, J. Nanosci. Nanotechnol. 16 (2016) 2509–2516.
- [14] M. Watanabe et al., Structural and Magnetic Properties in Sputtered Iron Oxide Epitaxial Thin Films - Magnetite Fe3O4 and Epsilon Ferrite e-Fe2O3-, Determinations Nanomed Nanotechnol. 1(1). DNN.000502. 2019.
- [15] R.H. Yu, S. Basu, Y. Zhang, J.Q. Xiao, Magnetic domains and coercivity in FeCo soft magnetic alloys, J. Appl. Phys. 85 (8) (1999) 6034–6036, https://doi.org/ 10.1063/1.369073.
- [16] H.S. Jung, W.D. Doyle, S. Matsunuma, Influence of underlayers on the soft properties of high magnetization FeCo films, J. Appl. Phys., Vol. 93, No. 10, Parts 2 & 3, 15 May 2003.
- [17] H. Hoffmann, T. Fujii, The wall coercivity of soft magnetic films, J. Magn. Magn. Mater. 128 (3) (1993) 395–400.
- [18] D. Caruntu, et al., Magnetic properties of variable-sized Fe₃O₄ nanoparticles synthesized from non-aqueoushomogeneous solutions of polyols, J. Phys. D: Appl. Phys. 40 (2007) 5801–5809.
- [19] X. Liu, H. Kanda, A. Morisako, The effect of underlayers on FeCo thin films, J. Phys. Conf. Ser. 266 (2011) 012037, https://doi.org/10.1088/1742-6596/266/1/ 012037.