Growth Induced Magnetic Anisotropy in Yttrium Iron Garnet Films on Yttrium Aluminum Garnet Substrates

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Abstract

Yttrium iron garnet (YIG) is a ferrimagnetic oxide of importance for a variety of applications including magnetic field sensors and microwave signal processing devices. Most of the applications, however, require a bias magnetic field that impedes miniaturization of YIG based sensors and communication devices. This work is aimed at strain engineering of YIG films to realize an induced magnetic anisotropy field that could act as a built-in bias field. We discuss here pulsed laser deposition (PLD) and characterization of YIG films on yttrium aluminum garnet (YAG) substrates. One expects the ~ 3% lattice mismatch between the film and substrate to result in an in-plane compressive strain and out-of-plane tensile strain. Films with thickness 55 nm to 380 nm were deposited on (100), (110), and (111) YAG substrates and annealed at high temperatures were characterized by structural and magnetic characterization techniques. The uniaxial out-of-plane strain anisotropy field estimated from X-ray diffraction data varied from 0.5 to 4.kOe for the films, showed an increase with increasing film thickness, and the highest values were for films on (100)YAG. Ferromagnetic resonance (FMR) measurements at 5 to 10 GHz also revealed the presence of a growth induced uniaxial anisotropy field as high as 450 Oe in the films, that decreased with increasing film thickness. The FMR line-width varied from 19.5 Oe to 82 Oe with films on (100) YAG showing the smallest values. The PLD YIG films on YAG substrates exhibiting a large perpendicular anisotropy field have the potential for use in sensors and high frequency devices.

Introduction

Yttrium iron garnet (YIG) and rare-earth iron garnets (RIG) are ferrimagnetic oxides with a Curie temperature well above the room temperature [1-3]. Bismuth substituted YIG and several RIG show high Faraday rotation and are useful for applications such as magnetic field sensors [4-6]. YIG is one of the materials with the lowest losses at high frequencies and is used in microwave devices including resonators and filters [7-9]]. Related recent developments in the applications aspects of the ferrimagnetic garnets are the demonstration of YIG-ferroelectric composites for magnetic field sensors and voltage tunable high frequency devices [10-13]. YIG is also of interest for use in devices based on spin torque transfer phenomena [14,15]. In most of these applications the need for a bias magnetic field renders their miniaturization hard to realize. For example, in the case of devices utilizing hexagonal ferrites with a large uniaxial or planar magnetocrystalline anisotropy field it might be possible to eliminate the need for a bias magnetic field, but YIG has a low magnetocrystalline anisotropy field on the order of 50 Oe [16]..

There were several attempts in the past to enhance the anisotropy field H_a in YIG and RIG with substitutions or by controlling the oxygen stoichiometry [17-19]. In the case of thin ferrimagnetic films, it is possible to achieve a significant strain induced enhancement in H_a due to film-substrate lattice mismatch [20-23]. For example, in ultrathin pure and Zn-substituted nickel ferrite films on magnesium aluminate and some gallate substrates H_a on the order 10 to 15 kOe was reported [22,23]. Thin films of YIG on yttrium aluminum garnet (YAG) has attracted considerable interest for achieving a large perpendicular magnetic anisotropy [24-30]. The lattice constants of YIG and YAG are 12.376 Å and 12.009 Å, respectively, and the -3% mismatch is expected to result in a compressive in-plane strain and a tensile

strain perpendicular to the film plane [31]. A recent report predicted a strain induced uniaxial H_a on the order of 2.6 kOe for YIG films on YAG substrates [31].

Past reports on films of YIG on YAG substrates utilized a variety of deposition techniques including chemical vapor deposition (CVD), RF sputtering, and pulsed laser deposition (PLD) [24-30]. Contrary to the expected compressive in-plane strain, early works on YIG films on (111) YAG deposited by CVD techniques reported an in-plane *tensile strain* primarily due to the thermal expansion mismatch between the film and the substrate [25]. Rapid deposition of YIG on YAG by metallo-organic CVD technique was found to result in YIG films with a high magnetization due to the presence of divalent Fe [26]. A recent report on PLD deposition of amorphous YIG on (100) YAG and subsequent recrystallization also resulted in an in-plane tensile strain [28]. Wang, et al., reported an out-of-plane uniaxial H_a that decreased from 1252 Oe in 9.8 nm thick YIG films on (100) YAG to 49 Oe in 72.7 nm thick films prepared by sputtering techniques [30]. One, however, prefers YIG film with thickness at least on the order of 0.5 to 1mm for applications for sensors and microwave devices.

We report here the deposition by PLD techniques of YIG films of thickness 55 nm to 380 nm on (100), (110) and (111) YAG substrates and results of a systematic study on the structural and magnetic properties. The films deposited at 650 C and annealed at 1000 C show an increase in the out-of-plane lattice constant c and a decrease in the in-plane lattice constant a with increasing film thickness. The estimated strain anisotropy H_s from X-ray diffraction data is uniaxial, perpendicular to the film plane, and increased with increasing film thickness. Magnetic force microscopy (MFM) phase images showed a stripe domain structure for all the films. The room temperature magnetization $4\pi M_s$ is in the range 1700 – 2000 G, in general agreement with values for bulk single crystal YIG. Ferromagnetic resonance (FMR) measurements at 5-10 GHz indicate a uniaxial anisotropy of 20 Oe to 450 Oe that generally decreases with increasing film thickness with the maximum values for films on (111) YAG. The FMR line-width DH ranges from 19.5 Oe to 92 Oe with the films on (100) YAG showing the lowest DH . Details on our study are presented in the sections that follow.

Experimental

The target of yttrium iron garnet for the deposition of films was synthesized by traditional ceramic processing techniques from yttrium oxide (Y_2O_3) and iron oxide (Fe_2O_3). First, the oxides were mixed with methanol in the ratio of 2:1 in a ball mill for 12 h and dried for 12 h at room temperature. Then the powder was presintered at 950 °C for 6 h. The sintered powders were ground to fine particles and ball milled for 24 h. Finally, the mixture was dried and pressed into a disk of diameter of 6 cm and sintered at 1525 °C for 15 h for use in the deposition system. Films of YIG were deposited for a duration of 30-120 min on (100), (110), and (111) oriented 0.5 mm thick yttrium aluminum garnet (YAG) substrates. The growth temperature for all the films was 650 °C at an oxygen pressure of 1.0 Pa. The distance between target and substrates was 5 cm and the laser pulse energy was 300 mJ (krypton fluoride, KrF) at a frequency of 5 Hz. All the samples were post annealed at 1000 °C in oxygen atmosphere for 2 h.

The film thickness was measured by cross-section scanning electron microscopy (SEM) images and the film composition was determined by energy dispersive X-ray spectroscopy. X-ray diffraction (XRD) patterns for the YIG target and films were obtained at room temperature using a powder diffraction system (Bruker D8 Advance, Cu K α , $\lambda = 1.5406 \text{ \AA}$). Imaging of the magnetic domains for YIG films was carried out by magnetic force microscope (MFM) using a Basic Atomic Force Microscope (B-AFM). The magnetization measurement was done at room temperature using a Faraday balance. Ferromagnetic resonance (FMR) measurements were performed with the films placed in a coplanar waveguide that was excited with microwave power using an Agilent Vector Network Analyzer (VNA). The DC bias magnetic field was applied parallel to the sample plane. Profiles of FMR were recorded by measuring the amplitude of the scattering matrix parameter S_{12} vs frequency f as a function of bias magnetic field H values and also by recoding the derivative of the power absorbed dP/dH as a function of f for a series of frequencies from 5 to 21 GHz.

Results

Films of YIG were deposited for durations of 30, 60, 90 and 120 min. Figure 1 shows cross sectional SEM images for films deposited on (100) YAG. The deposition rate varied from 1.83 nm/min to 3.17 nm/min for film thicknesses of 55–380 nm. Figure S-1 to S-8 in the supplement shows similar SEM images used for thickness measurements for films on (110) and (111) YAG. Table 1 shows the YIG film thickness as a function of deposition time for all orientations of YAG substrate. Surface topography of the films were probed with an AFM.

Table 1
Thickness of YIG films on (100), (110) and (111) YAG as a function of deposition time.

Film	Deposition time (Sec)	Thickness ($\pm 5\%$) (nm)
YIG/YAG (100)	30	55
	60	120
	90	170
	120	380
YIG/YAG (110)	30	91
	60	145
	90	171
	120	205
YIG/YAG (111)	30	127
	60	171
	90	274
	120	357

Figure 2 shows representative AFM images for films on (100) YAG. The images show smooth surface with a roughness of around 2 nm. Chemical composition for the films was determined by energy dispersive X-ray spectroscopy and was found to be $Y_{3\pm0.02}Fe_{5\pm0.05}O_{12}$, close to the stoichiometric composition (Figure S-9 in Supplement shows representative composition data).

The film crystal structure was studied by X-ray diffraction. Figure 3 shows the XRD $\theta - 2\theta$ plots for the YIG films on YAG substrates for (100), (110), and (111) orientations. All the YIG films were epitaxially grown on YAG substrates showing the (400), (440), or (444) diffraction peaks. The (400) diffraction peaks in Fig. 3(a) shift to the right toward the YAG peaks when the thickness of the films is decreased. A similar shift closer to the YAG peaks with decreasing film thickness is also observed for (440) and (444) peaks for the films on (110) and (111) YAG substrates, respectively. The out-of-plane lattice constants c for the YIG films obtained from the XRD scans using the Bragg formulation are listed in Table 2 and it increases as the film thickness is increased. The lattice constants for bulk YIG a_b is 12.376 Å [30–33]. The in-plane lattice constant a for the films can be estimated from the unit cell volume for bulk YIG and c -values in Table 2 and is given by $a = \frac{\sqrt{12.376^3}}{c}$ Å. Values of a as a function of the film thickness are also listed in Table 1.

Table 2
Estimated out-of-plane lattice constant c , in-plane lattice constant a , the out of plane strain ϵ_{zz} and in-plane strain ϵ_{xx} for YIG films on /YAG (100), (110) and (111) substrates.

Film	Thickness	c (Å)	a (Å)	ϵ_{zz}	ϵ_{xx}
YIG/YAG (100)	55 nm	12.440	12.344	0.52%	-0.26%
	120 nm	12.492	12.318	0.94%	-0.47%
	170 nm	12.544	12.293	1.36%	-0.67%
	380 nm	12.624	12.254	2.00%	-0.98%
YIG/YAG (110)	91 nm	12.405	12.361	0.23%	-0.12%
	145 nm	12.462	12.333	0.69%	-0.35%
	171 nm	12.484	12.322	0.87%	-0.44%
	205 nm	12.524	12.303	1.19%	-0.59%
YIG/YAG (111)	127 nm	12.429	12.355	0.43%	-0.17%
	171 nm	12.464	12.338	0.71%	-0.31%
	274 nm	12.491	12.325	0.93%	-0.41%
	357 nm	12.533	12.304	1.27%	-0.58%

The out-of-plane strain ϵ_{zz} for the YIG films was calculated using the expression $\epsilon_{zz} = \frac{c - a_b}{a_b}$, where a_b is the lattice constant for bulk YIG. The in-plane strain ϵ_{xx} can be estimated from $\epsilon_{xx} = (a - a_b)/a_b$. Values of the estimated in-plane and out-of-plane strains are listed in Table 2 and, as expected, the strains are compressive and tensile, respectively. Consider first the lattice constants and strain values in Table 2 for the films on (111) YAG. The a -value for the 127 nm thick film is close to that of bulk YIG value and it decreases with increasing films thickness. This is accompanied by an increase in the magnitude of compressive ϵ_{xx} from 0.26% for the 127 nm thick film to 0.98% for the 357 nm film. Values of c and the tensile strain ϵ_{zz} both increase with increasing film thickness. These observations are contrary to the expectations that a value for the thinnest film to be close to that of the YAG substrate and relaxation of the both strains and a decrease in their magnitudes with increasing film thickness. The films on (110) and (111) also show a similar dependence of lattice constants and strains on YIG film thickness and the films on (100) YAG substrate show the maximum values for both in-plane and out-of-plane strains.

The variations in the lattice constants and strains with the film thickness in our system is similar to the findings reported in Ref. 28 for films of YIG on (100) YAG. In that study amorphous films of YIG were deposited on unheated YAG substrates and then annealed at 800 C to obtain crystalline films. As in the present study, a similar decrease in the in-plane lattice constant a and an increase in the c -value were reported as the film thickness was increased from 9.6 nm to 56 nm [28]. The authors attributed these

observations to two causes: (i) tilting (rotation) of unit cell of YIG relative to YAG during the film growth and dislocations at the film-substrate interface and (ii) the difference in the linear thermal expansion coefficient α for the film and the substrate that results in a tensile strain at the interface. For YIG $\alpha = 10 \times 10^{-6}/\text{C}$ which is a factor 4 greater than that of YAG, $\alpha = 2.7 \times 10^{-6}/\text{C}$ [34, 35]. In another study, the measured in-plane *tensile strain* in a 4 μm thick YIG film on 1 mm thick YAG substrate grown by CVD techniques at 1250 C was accounted for by the mismatch in α [28]. In the present study (films grown at 650C and annealed at 1000 C) instead of anticipated $\varepsilon_{xx} \sim -3\%$, the observed $\varepsilon_{xx} = -0.26\%$ for the 55 nm film on (100) YAG and subsequent increase in its magnitude to -0.98% for the thickest film could be due to a combination of YIG-YAG compressive strain due to lattice mismatch and a tensile strain due to difference in α -values. Films on other orientations of YAG substrates also show a similar variation in the strains with film thickness. Estimates of the strain induced anisotropy field H_σ are provided in the Discussion Section.

The magnetic domain patterns for the thickest YIG films on (100), (110) and (111) YAG substrates were imaged by magnetic force microscope (MFM) without any applied magnetic field and are shown in Fig. 4. A stripe domain pattern is observed for the YIG films due to non-planar orientation for the magnetization and is indicative of a uniaxial anisotropy field perpendicular to the film plane [36, 37]. The domain width was determined from the MFM images and the width of the bright stripes is 460 nm, 462 nm, and 477 nm for (100), (110), and (111) YIG films, respectively.

The magnetization measurements at room temperature for the YIG films were performed with a Faraday balance. Figure 5 displays M vs H hysteresis loops for H applied in the plane of YIG films. The M - H data for the thickest YIG films on (100), (110), and (111) oriented YAG are shown in Fig. 5(a). The coercive field is the highest for (110) YIG film and the lowest for (100) YIG film. The YIG films have a coercivity of less than 25 Oe. The magnetization $4\pi M$ vs H are also shown in Fig. 5.(b)-(d) for films with thickness 58, 115, and 230 nm on (100) YAG. The saturation induction $4\pi M_s$ for the films are listed in Table 3 and are in the range 1700–1980 G, which is close to the value of 1750 G for bulk YIG [38]. Additional hysteresis loops for H in-plane and perpendicular to the film plane are shown in the supplement [Figure S-10 to S-14].

Table 3
Saturation induction $4\pi M_s$ for YIG films

Film	Thickness ($\pm 5\%$) (nm)	$4\pi M_s$ (kG)
YIG/YAG (100)	55	1.80
	120	1.70
	170	1.80
	380	1.67
YIG/YAG (110)	91	1.70
	145	1.74
	171	1.80
	205	1.67
YIG/YAG (111)	127	1.71
	171	1.77
	274	1.98
	357	1.86

The magnetization values were used for the determination of the growth induced anisotropy field from FMR measurements that are discussed next.

Ferromagnetic resonance measurements were carried out to determine the magnetic parameters for the films. The samples were placed in a coplanar waveguide that was excited with microwave power from a vector network analyzer. FMR profiles were recorded in the frequency range 5 GHz – 21 GHz, and for applied magnetic fields parallel to the film plane. Two types of measurements were done: (i) First derivative of the power absorbed by the sample, dP/dH was recorded at a constant frequency f as a function of H and (ii) the amplitude of the scattering matrix S_{21} was measured as a function of f for constant H . Figure 6 shows representative profiles of dP/dH vs H for the 357 nm thick YIG/(111) YAG for in-plane H and $f = 5\text{--}21$ GHz. The resonance frequency f_r and the field H_r for in-plane applied fields are related by the Kittel equation:

$$f_r^2 = \gamma^2 H_r (H_r + 4\pi M_{\text{eff}}) \quad (1)$$

where γ is the gyromagnetic ratio and the effective magnetization $4\pi M_{\text{eff}}$ is defined as $4\pi M_{\text{eff}} = 4\pi M_s + H_a$. Here H_a is the strain induced magnetic anisotropy field. Our FMR measurements as a function of orientation of in-plane H did not show any measurable change in the resonance frequency or field and is indicative of rather small magnetocrystalline anisotropy field H_c . A recent study on YIG films on YAG reported H_c on the order of 10–60 Oe [28]. We, therefore, did not taken into consideration H_c in Eq. (1).

Data on the resonance frequency and field were fitted to Eq. (1) as shown in Fig. 6 to obtain the parameters $\gamma = 2.79 \text{ GHz/kOe}$ and $4\pi M_{\text{eff}} = 1.579 \text{ kG}$.

FMR profiles of S_{21} vs H for excitation frequencies from 14 to 21 GHz are shown in Fig. 7 for 55 nm YIG film on (110) YAG. The in-plane field H was applied parallel to [001] of the film. The resonance appears as a sharp decrease in S_{21} and the figure also shows the fit to the data on resonance field and frequency. Estimated values of the magnetic parameters are $\gamma = 2.8 \text{ GHz/kOe}$ and $4\pi M_{\text{eff}} = 1.35 \text{ kG}$. The values of γ determined from the FMR data in Figs. 6 and 7 agree with the value of 2.8 GHz/kOe for single crystal YIG [38]. Similar FMR measurements

were also done on YIG/(110) YAG films with H applied parallel to [001]. The parameters γ and $4\pi M_{\text{eff}}$ for all of the films obtained from data as in Figs. 6 and 7 are listed in Table 4. Additional S_{21} vs f profiles as in Fig. 7 for FMR are shown in the Supplement [Fig.S15 – S19].

Table 4
The gyromagnetic ratio γ and the effective saturation induction $4\pi M_{\text{eff}}$ for YIG films on YAG substrates.

Film	Thickness (nm)	γ (GHz/kOe)	$4\pi M_{\text{eff}}$ (kG)
YIG/YAG (100)	55	2.80	1.35
	120	2.79	1.55
	170	2.77	1.64
	380	2.76	1.70
YIG/YAG (110)	91	2.77	1.44
	145	2.79	1.49
	171	2.78	1.60
	205	2.77	1.77
YIG/YAG (111)	127	2.79	1.36
	171	2.79	1.51
	274	2.78	1.53
	357	2.70	1.58

The FMR line-width ΔH for the films was determined from profiles as in Fig. 6. Figure 8 shows the measured peak-to-peak ΔH values for the thickest YIG films on YAG substrates for excitation frequencies from 5 to 10 GHz. The line width increases with increasing frequencies for films of all three orientations on YAG. Films on (100) YAG show the smallest ΔH value that increases from 19.2 Oe at 5 GHz to 22 Oe at

10 GHz. Films on (111) YAG have $\Delta H = 50$ Oe to 60 Oe for the same frequency range. Films on (110) YAG show the highest values for line-width with ΔH increasing from 70 Oe at 5 GHz to 90 Oe at 10 GHz. The line-widths in Fig. 8 for the films are in agreement with $\Delta H = 25$ to 86 Oe reported for films on YIG on (100) YAG [28, 30]. The determination of the growth induced magnetic anisotropy field for the films is considered in the following section.

Discussion

Now we discuss results of this comprehensive study on the structural and magnetic properties of YIG films on the three different orientations of YAG substrates. Films with thickness 55–380 nm were deposited by PLD techniques and, following the growth, annealed at 1000 C in oxygen atmosphere. X-ray diffraction data revealed the expected epitaxial films on all of the substrates. Significant findings from the XRD measurements are as follows. (i) For all three substrate orientations, the thinnest films have in-plane lattice constants a close to that of YIG rather than the anticipated a value for YAG. (ii) With increasing film thickness, a -value decreases and the out-of-plane lattice constant c increases. (iii) The in-plane strain is compressive and the out-of-plane strain tensile and the magnitude of both strains *increase* with increasing film thickness. (iv) Although the $\sim 3\%$ lattice match for YIG films on YAG is expected to be the determining factor in the growth of ultrathin films [30], the thermal expansion between YIG and YAG seems to be the key factor that determines the structure and growth induced strains for films annealed at high temperatures as is the case for our films. This observation agrees with similar findings in previous reports on YIG films on YAG [25, 28].

Next, we estimate the strain induced anisotropy field H_σ from the lattice constants in Table 2. The field H_σ is given by [22]

$$H_\sigma = 3 \lambda \sigma / M_s \quad (2)$$

where λ is the magnetostriction and σ is the effective strain given by

$$\sigma = (a - c) / a_b (1 + \nu) \quad (3).$$

In Eq. (3) a , c , and a_b are the in-plane and out-of-plane lattice constants for the YIG film and the lattice constant for bulk YIG, respectively, E is the Young's modulus and ν is the Poisson's ratio. Using Eqs. (2) and (3) and values of a and c in Table 2, $a_b = 1.2376$ nm, $E = 2 \times 10^{12}$ dynes/cm², $\lambda = -2$ ppm, $\nu = 0.295$, and $M_s = 140$ emu/cc, calculated H_σ values are listed in Table 5. Positive values of H_σ indicate easy plane anisotropy field whereas negative values of H_σ are for uniaxial anisotropy field perpendicular to film plane. The anisotropy fields H_σ in Table 5 are uniaxial and increase with increasing film thickness for all of the YAG substrate orientations. Films on (100) YAG have the highest H_σ value and its magnitude increase from 1 kOe for 55 nm film to 4 kOe

Table 5
The strain anisotropy field H_σ estimated from XRD data
and growth induced anisotropy field H_a determined
from FMR data .

Film	Thickness (nm)	H_σ (kOe)	H_a (kOe)
YIG/YAG (100)	55	-1.02	-0.45
	120	-1.86	-0.15
	170	-2.69	-0.16
	380	-3.96	-0.03
YIG/YAG (110)	91	-0.47	-0.26
	145	-1.38	-0.25
	171	-1.73	-0.20
	205	-2.36	0
YIG/YAG (111)	127	-0.79	-0.35
	171	-1.35	-0.26
	274	-1.78	-0.44
	357	-2.45	-0.28

for the thickest film. Films on (110) and (111) YAG substrates have H_σ of approximately the same value. The thickest films on (100), (110) and (111) YAG have $H_\sigma = 2.4-4.0$ kOe, in very good agreement with recent theoretical predictions [31].

The MFM images in Fig. 4 for the domain structure in the YIG films show stripe domains due to an out-of-plane orientation for the magnetization in the absence of an external magnetic field. Data on the dependence of magnetization on H showed coercive fields H_c less than 25 Oe for in-plane hysteresis loops. The saturation magnetization $4\pi M_s$ measured at room temperature vary from a minimum of 1.67 kG to a maximum of 2 kG. Past studies reported $4\pi M_s$ as high as 2.56 kG in YIG films [26, 39] and were attributed to distribution and/or ionic states of Fe.

The FMR data on the effective magnetization $4\pi M_{eff}$ and the saturation magnetization values in Table 3 can be used to estimate the growth induced anisotropy field H_u . The effective magnetization $4\pi M_{eff} = 4\pi M_s + (H_{mc} + H_a)$, where H_{mc} is the magnetocrystalline anisotropy field. As mentioned earlier, H_{mc} is rather small in YIG and is shown to vary from 42 Oe for 9.8 nm YIG film on (100) YAG to 25.6 Oe for 72.7 nm thick film [30] and can be ignored in comparison to the magnitude of H_a . Thus $H_a = 4\pi M_{eff} - 4\pi M_s$.

Values of H_a determined from the FMR and magnetization data are listed in Table 5. The field H_a is negative and is oriented perpendicular to the film plane. Our FMR data on resonance fields for in-plane fields did not show any significant dependence on the direction of H and was indicative of negligible uniaxial in-plane growth induced anisotropy. Values of H_a in Table 5 range from a maximum of 450 Oe for 55 nm thick film on (100) YAG to a minimum of zero for the 205 nm thick film on (110) YAG and show a general decrease with increasing film thickness. Values of H_a show a significant departure from both the magnitudes and thickness dependence of strain anisotropy field H_σ estimated from XRD data. Whereas H_a shows the expected decrease with increasing film thickness, the strain anisotropy H_σ shows an increase with increasing film thickness. Values of H_a are a factor of 2 to 10 smaller than the H_σ values. The cause of the difference in both the values and thickness dependence of H_a and H_σ need further investigation.

Next, we compare H_a values for our films with past reports for YIG and spinel ferrite films. FMR measurements on films of YIG on (100) YAG with thicknesses 9.8 nm to 72.7 nm were reported to have H_a decreasing from 1252 Oe to 49 Oe [30]. A very recent study on recrystallized PLD films of YIG on (100) YAG reported H_a decreasing from 72 Oe for 9.6 nm film to 30 in 56 nm thick film [28]. Thus, our 55 nm to 380 nm thick YIG films show a modestly high H_a values compared to previous reports. Films of spinel ferrites such as nickel ferrite (NFO) show a much higher H_a than YIG films due to an order of magnitude higher magnetostriction. Films of NFO, 445 nm in thickness, deposited by PLD on magnesium gallate substrate had $H_a = 11.9$ kOe [22]. Zinc and Al-substituted NFO, 15–53 nm in thickness on MgAl_2O_4 substrate, showed $H_a \sim 10$ kOe [23]. Although the YIG films in this study show H_a much smaller than for spinel ferrite films, our YIG films on (111) YAG show one of the highest perpendicular anisotropy filed reported for YIG and are potentially of interest for applications in sensors and high frequency devices.

Conclusions

This work is on the deposition of thin films of YIG on (100), (110) and (111) YAG substrates by PLD techniques and characterization by structural and magnetic measurement techniques. Films of thickness 58–460 nm were annealed at high temperatures. As the film thickness was increased XRD data revealed a decrease in the in-plane lattice constant and an increase in the out-of-plane lattice constant. A growth induced in-plane compressive strain and an out-of-plane tensile strain were inferred from the XRD data. A stripe domain structure for the films was evident from the MFM phase images. Room temperature saturation magnetization values were found to be in general agreement with magnetization for bulk single crystal values. The strain anisotropy field H_σ determined from the XRD data was uniaxial and perpendicular to the film plane and its magnitude increased with increasing film thickness. Films on (100) YAG showed the highest values for H_σ . Ferromagnetic resonance studies at 5 to 21 GHz were performed to determine the growth induced anisotropy field H_a for the films. A decrease in H_a with increasing film thickness was evident for the films on all three orientations of the YAG substrates and estimated values

of H_a ranged from a maximum of 450 Oe for the 55 nm films to a minimum of 30 Oe for 380 nm thick films.

Declarations

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Author Contributions:

Data acquisition: Y.L. P.Z, R. B, and G.S; Analysis: Y.L., P.Z., J.Z., W.Z., M. R. P, and G. S., Funding acquisition: P.Z., W.Z., M.R.P, T. Z., and G.S; Project administration: T.Z., and G.S.; Manuscript draft preparation: Y.L., P.Z., and G.S. All authors contributed to preparation of final version of the manuscript.

Data Availability Statement: -

All data generated or analyzed during this study are included in this published article [and its supplementary information files].

Conflicts of Interest:

The authors declare no conflict of interest.

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Figures

Figure 1

SEM cross-sectional images showing the thickness of YIG films deposited for 30, 60, 90 and 120 min on (100) YAG.

Figure 2

AFM topography images for YIG films on (100) YAG.

Figure 3

(a) XRD patterns for the YIG film on (a) (100), (b) (110), and (c) (111) YAG substrates

Figure 4

(a)-(c) MFM phase images showing the stripe domains in the thickest YIG films on (100), (110), and (111) YAG substrates.

Figure 5

(a) Normalized in plane M-H hysteresis loops for the thickest YIG films grown on (100), (110), and (111) YAG substrates, and $4\pi M$ vs H for films with thickness (b) 170 nm, (c) 120 nm and (d) 55 nm on (100) YAG.

Figure 6

(a) Profiles of dP/dH vs H showing FMR in 357 nm thick YIG/(111) YAG for H applied parallel to film plane. (b) Theoretical fit to data on resonance frequency f_r and field H_r for estimation of magnetic parameters for the film.

Figure 7

(a) Profiles of amplitude scattering matrix S_{21} vs H showing FMR in 55 nm thick YIG/(100) YAG for in-plane H along [001] direction of YAG. (b) Theoretical fit to data on resonance frequency and field for estimation of magnetic parameters for the film.

Figure 8

Dependence of FMR line-width ΔH on film thickness for 380 nm thick YIG film grown on (100) YAG, 205 nm film on (110) YAG, and 357 nm film on (111) YAG substrates.

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