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LPE GROWTH AND MAGNETIC ANISOTROPY OF Ni(Fe, Al)\(_2\)O\(_4\) FILMS

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Results are reported about the LPE growth of Ni(Fe,Al)\(_2\)O\(_4\) films grown onto (100), (110) and (111) MgO and on (111) ZnGa\(_2\)O\(_4\), Zn(Ga,Al)\(_2\)O\(_4\) and Zn(Ga,Fe)\(_2\)O\(_4\) substrates using a PbO–B\(_2\)O\(_3\) flux. For small film thicknesses and small lattice mismatches elastically deformed films, showing no stress relief, can be grown on the spinel substrates. Interdiffusion between film and substrate occurs to a small extent when spinel substrates are used. For films grown in tension, positive uniaxial anisotropies are found and stripe domains are observed at the film surface. It is shown that the anisotropy can be explained by the stress in the film and the negative magnetostriction constant of NiFe\(_2\)–\(x\)Al\(_x\)O\(_4\). Crack-free films with a uniaxial anisotropy energy exceeding the demagnetization energy are obtained when the misfit is lower than 0.03 \(A\) and 0.50 < \(x\) < 0.65. For those films the stripe domains can be contracted to bubbles with the use of a bias field.

1. Introduction

A stress-induced anisotropy \(K^3\) is present in spinel epitaxial films when a stress \(\sigma\) is present in the spinel film at room temperature. When \(\lambda\) is the magnetostriction constant of the film, the stress-induced anisotropy can be calculated as

\[ K^3 = \frac{3}{2} \lambda_{111} \sigma, \]

for films grown on (111) substrates. The stress \(\sigma\) at room temperature [1] can be calculated as

\[ \sigma = \left(1 - \eta\right) \frac{a_s - a_f}{a_f} + \eta(\alpha_f - \alpha_s) \Delta T \frac{E}{1 - \nu}, \]

where \(a_s, a_f\) = bulk room temperature lattice parameters of substrate and film respectively, \(\alpha_s, \alpha_f\) = thermal expansion coefficients of substrate and film respectively, \(\Delta T\) = difference between deposition temperature and room temperature, \(E\) = Young’s modulus, \(\nu\) = Poisson ratio, and \(\eta\) = fractional stress-relief (0 < \(\eta\) < 1).

For small misfit \((a_s - a_f)\) values and thin films, the film is strained elastically to bring the lattices of film and substrate into register at the interface \((\eta = 0)\). For higher misfit values and thicker films it is assumed that the film takes on its equilibrium free lattice constant away from the interface, so that it is essentially unstressed at the deposition temperature. The misfit stress is relieved by the formation of misfit dislocations. Since the strain energy increases with increasing film thickness, there is a critical thickness, \(h_c\), above which it becomes energetically favourable to form misfit dislocations [2]:

\[ h_c = \frac{b(1 + \nu) a_f}{8\pi(a_s - a_f)} \ln(r/b) + 1, \]

where \(b\) is the Burgers vector and \(r\) the dislocation strain-field radius. When stress relief occurs at the deposition temperature \((\eta = 1)\) and when the film is considered to behave elastically on cooling to room temperature, a stress develops due to differences between \(\alpha_f\) and \(\alpha_s\). Intermediate situations (0 < \(\eta\) < 1) are also possible.

In a previous article [3] a stress-induced anisotropy was reported in LPE grown Ni(Fe,Al)\(_2\)O\(_4\) films. On films, grown in tension \((a_s > a_f)\) upon ZnGa\(_2\)O\(_4\) substrates, a serpent-like domain pattern could be made visible with Bitter fluid. Since for these films \(\sigma > 0\) and \(\lambda_{111} < 0\), a positive stress-


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induced anisotropy results. In this article we report about the growth and characterization of Ni(Fe,Al)$_2$O$_4$ films grown upon MgO, ZnGa$_2$O$_4$, Zn(Ga,Al)$_2$O$_4$ and Zn(Ga,Fe)$_2$O$_4$ substrates.

2. The growth of Al- and Fe-substituted zinc gallate substrates

In order to grow compressively and tensively stressed Ni(Fe,Al)$_2$O$_4$ films, substrates are needed with different lattice constants. The substrates are based on ZnGa$_2$O$_4$ ($a_0 = 8.336$ Å) which can be grown from a ZnO–Ga$_2$O$_3$–Pb$_2$P$_2$O$_7$ melt by slow cooling [4]. Octahedrally shaped crystals with \{111\} facets of dimensions up to 6 mm along the edge could be obtained. When Al$_2$O$_3$ or Fe$_2$O$_3$ is added to the melt, doped zinc gallates are obtained with lattice constants respectively lower and higher than 8.336 Å. (Table 1). The lattice constants of the crystals were determined by X-ray diffractometry by measuring the Bragg angles from planes parallel to the \{111\} facet face. Since from ceramic standards a linear relation was found between composition and lattice constants in the Al- and Fe-substituted ZnGa$_2$O$_4$ systems, the composition near the surface of the crystals could be determined from the lattice constants. Although the surface compositions of crystals obtained from one melt composition are rather close to each other, we always find some composition range. Concentration gradients in the melt during the growth or slightly different growth rates must be the cause of this. When the surface compositions of the crystals are compared with the melt composition, we find for Fe a 60% higher and for Al a 40% lower concentration (fig. 1). Also at the growth of Ni(Fe,Al)$_2$O$_4$ single crystals [5] a lower concentration of Al was found. The effect was more pronounced with a Pb$_2$P$_2$O$_7$ flux (−50%) than with a PbO–B$_2$O$_3$ flux (−30%).

We have performed microprobe analyses on crystals cut through the centre. Up to about 1 mm from the surface, the composition remained rather constant and was in excellent agreement with the composition determined from the lattice constant of the crystal. However, more towards the centre of the crystals a distinct concentration gradient was found. For the Zn(Ga,Fe)$_2$O$_4$ crystals the Fe content decreases towards the centre, while for the Zn(Ga,Al)$_2$O$_4$ crystals the Al content increases. In a large ZnGa$_2$–Fe$_2$O$_4$ crystal (7 mm) grown from a melt corresponding with $x = 0.38$ we find a value of $x = 0.60$ near the surface and a value of $x = 0.31$ in the centre.

The concentration gradients in the crystals must be due to a strong temperature dependence of the segregation coefficients for Fe and Al, defined as the molar ratios:

$$k_M = [M/(M + Ga)]_{crystal}/[M/(M + Ga)]_{melt}, \quad (4)$$

where M = Fe or Al respectively.

Since the crystals are grown from the melt by slow cooling with a rate of approximately $1^\circ$C/h from 1300–800$^\circ$C, the material in the centre of the largest crystals will be grown at high temperatures, while the outer part of the large crystals and the entire small crystals are grown at lower temperatures. In the high temperature region $k_{Al}$ must be >1 and $k_{Fe}$ <1; the reverse will be true in the low temperature region.

<table>
<thead>
<tr>
<th>$x$</th>
<th>Pb$_2$P$_2$O$_7$</th>
<th>ZnO</th>
<th>Ga$_2$O$_3$</th>
<th>Fe$_2$O$_3$</th>
<th>Al$_2$O$_3$</th>
<th>$a_0$ (Å)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>0.40</td>
<td>0.48</td>
<td>0.120</td>
<td>–</td>
<td>–</td>
<td>8.336</td>
</tr>
<tr>
<td>0.20</td>
<td>0.40</td>
<td>0.48</td>
<td>0.108</td>
<td>0.012</td>
<td>–</td>
<td>8.348–8.351</td>
</tr>
<tr>
<td>0.38</td>
<td>0.40</td>
<td>0.48</td>
<td>0.097</td>
<td>0.023</td>
<td>–</td>
<td>8.360–8.367</td>
</tr>
<tr>
<td>0.76</td>
<td>0.40</td>
<td>0.48</td>
<td>0.074</td>
<td>0.046</td>
<td>–</td>
<td>8.392–8.397</td>
</tr>
<tr>
<td>0.35</td>
<td>0.40</td>
<td>0.48</td>
<td>0.099</td>
<td>–</td>
<td>0.021</td>
<td>8.310–8.300</td>
</tr>
<tr>
<td>0.70</td>
<td>0.40</td>
<td>0.48</td>
<td>0.078</td>
<td>–</td>
<td>0.042</td>
<td>8.286–8.270</td>
</tr>
</tbody>
</table>
When spinel substrates are used, the films are grown directly on the (111) facets of the substrates with no other precautions than cleaning of the substrates. MgO substrates are cut from a boule of MgO using an internal diamond cutting blade. Substrates having (100), (110) and (111) orientations, cut within 0.5°, were used. Cutting damage was removed by polishing with SiC and syton.

The LPE grown Ni(Fe,Al)₂O₄ films are characterized using three parameters: the Al content, the thickness and the lattice constant difference between film and substrate (misfit).

From microprobe data all films were shown to be pure spinels with a Ni content of 0.99 ± 0.02 atoms Ni and a Pb content less than 0.02 atoms Pb per spinel formula unit. When the PbO, B₂O₃ and NiO content of the melt is kept constant, the Al content in the films increases when the Al₂O₃ : Fe₂O₃ ratio in the melt is increased. When the B₂O₃ content in the melt is increased, keeping the other components constant, the saturation temperature of the melt and the Al content in the film decrease. In fig. 2 the relation between the Al content of the film and the B₂O₃ content in the melt is plotted. Saturation temperatures are determined by extrapolating to zero growth rate. Undercoolings up to 60°C are used. When the melt composition is kept constant, the Al content in the film increases by about 10% when decreasing the undercooling from 60 to 10°C.

3. The growth of Ni(Fe, Al)₂O₄ films

For the LPE growth we have used a melt composed of 1 PbO, 0.13 B₂O₃, 0.013 NiO and 0.11 Fe₂O₃ (moles) for the growth of pure NiFe₂O₄ films. For the growth of Al-substituted films 0.05 moles Al₂O₃ were added to the melt per mole PbO and the B₂O₃ content ranged from 0.13 to 0.45 moles B₂O₃ per mole PbO. The flux constituents were mixed and premelted in a 75 ml Pt crucible; the total load was about 300 g. Actually 1 wt% of the PbO was replaced by PbO₂ for protection of the crucible. The crucible was placed in a vertical three-zone resistance furnace having a large isothermal area. The film growth is performed by dipping a substrate for a certain time in the supersaturated melt at a constant temperature. Afterwards, the substrate–film combination is cleaned in a hot dilute acetic–nitric acid solution.

Fig. 1. Surface compositions of Zn(Ga, Fe)₂O₄ and Zn(Ga, Al)₂O₄ crystals versus the compositions calculated from the Ga/Fe and Al/Fe ratios in the melts.

Fig. 2. Aluminum content in Ni(Fe, Al)₂O₄ films versus the amount of B₂O₃ (moles) in the melt per mole PbO. Undercooling during the growth of the films between 10 and 60°C.
The film thickness can in principle be determined by grinding a steel sphere, coated with diamond paste, through the film. This method, however, can only be applied successfully when the film thickness is reasonably constant over the substrate surface. Films grown on (100) and (110) MgO were rough with surfaces consisting respectively of pyramids and ridges composed of (111) facets. Films grown on (111) MgO substrates are reasonably flat, but due to the slight misorientation (<0.5°) of the substrates, small terraces can be observed on the surface of the films. Obviously, only LPE growth on the habit face (111) results in a smooth surface. For films on MgO substrates the film thickness could be determined within 10%.

When films are grown on the natural (111) facets of flux grown spinel substrates, hillocks are observed on the film surfaces with heights up to 4 μm and base cross sections up to 200 μm. The hillocks are observed over the entire undercooling range used (10—60°C). Since for spinels (111) is a habit face, growth may occur by two-dimensional nucleation or by a spiral growth mechanism. The hillocks grow by means of the spiral growth mechanism, in which a screw dislocation, with its Burgers vector ending on the surface, acts as a continuous step source. In literature, growth hillocks are reported in (111) spinel [6] and (110) garnet [7,8] films, orientations corresponding to habit faces of spinel and garnet respectively. Since surface irregularities seem to act as step sources, the density of the hillocks, especially when low undercoolings are used, could be reduced by boiling the substrates for a short time in H₃PO₄, but we did not succeed in removing them completely. Determination of the average film thickness is very inaccurate when large hillocks are present at the film surface. When the surface of the film is illuminated with thallium light (λ = 0.54 μm), thickness variations are made visible. In figs. 3a—3c, thallium light photographs are shown for films grown from the same melt with the same dipping time, grown with increasing undercooling and as a consequence with increasing average film thickness. At the highest undercooling cracks can be observed, while at the lowest undercooling individual small hillocks are present.

The stress in the film results from the misfit between film and substrate, as described by eq. (2).
X-ray diffractometry, from the Bragg angles of planes parallel to (111).

When \( a_\text{f} \) and \( a_\text{s} \) are the bulk room temperature lattice constants of film and substrate, respectively, and \( a_\text{f}^\perp \) is the measured lattice constant of the film on the substrate, determined from a film plane parallel to the film surface, it can be shown that for \( \eta = 0 \):

\[
a_\text{f}^\perp - a_\text{s} = \frac{1 + \nu}{1 - \nu} (a_\text{f} - a_\text{s}).
\]

Since for spinels the Poisson ratio is about 0.3, it follows that when \( \eta = 0 \), the measured misfit \( \Delta a^\perp = 1.86 \Delta a \), in which \( \Delta a \) is the bulk misfit.

When 100% stress relief occurs at the deposition temperature (\( \eta = 1 \)), a stress develops due to \( \Delta a \) and \( \Delta T \), which also causes a deformation of the spinel film. Using high temperature diffractometry, we have determined the thermal expansion coefficients for MgO, NiFe\(_2\)O\(_4\), NiFe\(_{1.5}\)Al\(_{0.5}\)O\(_4\), ZnGa\(_2\)O\(_4\) and ZnGa\(_{1.5}\)Al\(_{0.5}\)O\(_4\) (see table 2).

For the substrate–film combinations used in this study, \( \alpha_\text{f} - \alpha_\text{s} \) is about +0.05 \( 10^{-5}\text{C}^{-1} \) (spinel substrates) or about \(-0.3 \times 10^{-5}\text{C}^{-1} \) (MgO substrates). With \( \Delta T = 900^\circ\text{C} \) and \( E = 1.6 \times 10^{12} \text{dyn/cm}^2 \) [9], a stress in the film of respectively +0.1 \( \times 10^8 \) and \(-0.6 \times 10^9 \) N/m\(^2\) is calculated (\( \eta = 1 \)), resulting in a lower, respectively higher, value of \( a_\text{f}^\perp \) compared with \( a_\text{f} \).

In conclusion: when the misfit is not too small, the state of stress can be followed, for a film with a fixed composition, grown upon a fixed substrate, by measuring \( a_\text{f}^\perp \) and knowing the values for \( \eta = 0 \) and \( \eta = 1 \).

In fig. 4 experimental and theoretical relations between \( a_\text{s} \), \( a_\text{f} \) and \( a_\text{f}^\perp \) are presented, clearly showing the change of \( a_\text{f}^\perp \) with changing \( \eta \).

<table>
<thead>
<tr>
<th>Compound</th>
<th>( \alpha \times 10^6 (\text{°C}^{-1}) )</th>
</tr>
</thead>
<tbody>
<tr>
<td>MgO</td>
<td>+1.30 ± 0.10</td>
</tr>
<tr>
<td>ZnGa(_2)O(_4)</td>
<td>+0.95 ± 0.05</td>
</tr>
<tr>
<td>ZnGa(<em>{1.5})Al(</em>{0.5})O(_4)</td>
<td>+0.90 ± 0.05</td>
</tr>
<tr>
<td>NiFe(_2)O(_4)</td>
<td>+1.00 ± 0.05</td>
</tr>
<tr>
<td>NiFe(<em>{1.5})Al(</em>{0.5})O(_4)</td>
<td>+0.95 ± 0.05</td>
</tr>
</tbody>
</table>

4. The domain structure of the films

When a stress is present in the film at room temperature, a uniaxial anisotropy energy results according to eq. (1). From magnetostriction measurements performed on NiFe\(_{2-x}\)Al\(_x\)O\(_4\) single crystals (for \( x = 0 \) to 0.3) by means of laser interferometry [5,10], it is shown that \( \lambda_{111} \) is negative and \( |\lambda_{111}| \) decreases with increasing Al content. For pure NiFe\(_2\)O\(_4\), \( \lambda_{111} = -22 \times 10^{-6} \) and from linear extrapolation it is found that \( \lambda_{111} \) approaches zero for \( x \sim 0.8 \). Knowing \( \lambda_{111} \) as a function of the Al content \( x \) in the film, we are able to calculate the stress-induced anisotropy energy \( K_a^\lambda \) as a function of \( x \) for different misfit values assuming \( \eta = 0 \). The results are presented in fig. 5 with the lattice constant of the substrate as a parameter. From this figure it can be seen that when the misfit is high enough, the stress induced-anisotropy energy exceeds the demagnetization energy \( 2\pi M^2 \) (\( M = \text{satisfaction magnetization} \)). With increasing Al content in the film, the minimum misfit
Fig. 5. Theoretical values for the stress induced anisotropy versus the aluminum content and the lattice parameter of Ni(Fe, Al)$_2$O$_4$ with the lattice parameter of the substrate as parameter (under-lined values in Å). Also the relation between $\lambda_{111}$ and the aluminum content in Ni(Fe, Al)$_2$O$_4$ is given. The dashed parts of this relation are the values obtained by extrapolation. The values between brackets represent the misfits required to achieve $K_u > 2\pi M^2$.

required to achieve this situation decreases.

When Ni(Fe, Al)$_2$O$_4$ films are grown on MgO ($2a_0 = 8.42$ Å), no domain pattern could be observed. Obviously stress relief occurs at the deposition temperature due to the very large misfit and a compressive stress is present in the film at room temperature, resulting in a negative stress-induced anisotropy. On spinel substrates, however, where a closer lattice fitting is obtained, serpentine-like domain patterns are observed when films are grown in tension ($K_u > 0$), while for films grown in compression ($K_u < 0$) no domain pattern is observable anymore. When cracks are present in the film, the stress is relieved along the crack and the domain pattern vanishes (fig. 6). In all NiFe$_{2-x}$Al$_x$O$_4$ films with $x$ ranging from zero up to 0.65, domain patterns are observed provided the films were grown in tension and not too large misfit values were used ($\Delta a < 0.05$ Å). In general the width of the magnetic domains decreases with decreasing film thickness and with increasing saturation magnetization of the film (figs. 7a and 7b).

When a magnetic field is applied perpendicular to the plane of the film, the width of the domains with its magnetization in the same direction of the field increases, while the width of the other domains decreases with increasing field. When $0 < K_u < 2\pi M^2$, the domain pattern becomes diffuse and finally vanishes when the field is increased above a certain value. When $K_u > 2\pi M^2$, however, the domain pattern starts unwinding and the domains with their magnetization opposite to the direction of the field contract to cylindrical bubbles with increasing field. When the bias field is increased further, the bubble diameter decreases by about 25% and finally the bubble collapses (fig. 8). Given a certain film, the field at which the bubble collapse is not very well defined. When some serpentine like domains are still contracting towards bubbles, other domains are already in the bubble state, while some bubbles have already collapsed. Some bubbles only collapse when much higher field strengths are used. When, after complete saturation of the film, the magnetic field is reduced to zero, the serpentine-like domains

Fig. 6. Bitter domain pattern observed on a NiFe$_{2-x}$Al$_x$O$_4$ film with $x = 0.11$ grown onto a ZnGa$_2$O$_4$ substrate, showing the vanishing of the pattern near the crack.
become visible again, but in some areas they are completely absent. When the film is demagnetized by heating for a short time above its Curie temperature or by moving a demagnetization pen over the film surface, the original domain pattern is obtained again. In general, crack-free NiFe$_{2-x}$Al$_x$O$_4$ films with $K_u > 2\pi M_s^2$ are obtained when $0.50 < x < 0.65$, and the films are grown in tension with a misfit lower than about 0.03 Å.

5. Anisotropy measurements

The uniaxial anisotropies in the Ni(Fe,Al)$_2$O$_4$ films are measured using a torque magnetometer. Therefore, after the growth of the films on the (111) facets of flux grown substrates, the crystals were shaped in such a way that one substrate facet with a film on one side was left.

The anisotropy is calculated as

$$K_u = \left( \frac{KV}{M_s V} \right) M_s + 2\pi M_s^2.$$  \hfill (6)

Theoretically eq. (6) also contains a contribution of the magnetocrystalline anisotropy $K_1$ [11]. This contribution, however, is very small and falls within the limits of inaccuracy.

$KV$ is obtained from the torque measurement; $M_s V$ is determined using a Faraday balance and $M_s$ can be calculated when the volume of the film is known. For films grown on spinel substrates, however, inaccuracies up to 25% are present in the calculated $M_s$ values caused by inaccuracies in the determination of the film thickness (growth hillocks). When the composition of the film is known, $M_s$ can be taken from literature.

In calculating $K_u$ from eq. (6), the small dimensions of our spinel substrates result in large inaccuracies. A 2 μm thick film, grown on a triangular facet measuring 5 mm along the edges, has a volume of about $2 \times 10^{-5}$ cm$^3$. When films are grown in tension with a misfit of 0.02 Å, $KV$ ranges from about $-2$ erg for NiFe$_2$O$_4$ up to about +2 erg for NiFe$_{1.4}$Al$_{0.6}$O$_4$ films. The inaccuracy of the torque magnetometer is 2% for a 100 erg sample, but is 10% for a 10 erg sample. The lower limit is about 0.2 erg. Inaccuracies of 10–20% are to be expected.

For polycrystalline NiFe$_2$O$_4$, $M_s = 258$ Gauss according to Blasse and Gorter [12]. For polycrystalline NiFe$_2$O$_4$, sintered at 1250°C in oxygen, we obtained 265 ± 5 Gauss. For NiFe$_2$O$_4$ single crystals, however, grown from a PbO–B$_2$O$_3$ flux [5], a value of 229 ± 5 Gauss was obtained measured over more than 10 crystals. Since the PbO–B$_2$O$_3$ flux is strongly oxidizing, we believe that the higher $M_s$ values found for polycrystalline samples are due to some Fe$^{3+}$ in
Fig. 8. Bitter domain patterns observed on a NiFe$_{2-x}$Al$_x$O$_4$ film with $x = 0.52$ grown onto Zn(Ga,Al)$_2$O$_4$ without (a) and with an external magnetic field $H$ perpendicular to the film; (b) $H = 95$ Oe, (c) 190 Oe, (d) 290 Oe.

substitution of Zn$^{2+}$ and Ga$^{3+}$ in the spinel film occurs on tetrahedral sites and will result in an increase of $M_s$ of the film when small amounts of Zn$^{2+}$ and Ga$^{3+}$ are present. In a previous paper [13] we have shown that interdiffusion occurs when lithium ferrite films are grown on MgO substrates above 1000°C. Robertson et al. [14] have shown that very severe interdiffusion occurs when (Mn, Zn) Fe$_2$O$_4$ films are grown into Zn$_2$TiO$_4$ substrates, even at temperatures below 900°C, while on MgO substrates no evidence for interdiffusion was found. Interdiffusion results in broadened and/or asymme-
μm NiFe₂O₄ film grown for 30 min at 870°C onto a ZnGa₂O₄ crystal, showed that the film composition can approximately be given as (1−x)NiFe₂O₄−xZnGa₂O₄. At the film–air surface x is about zero and reaches a value of about 0.03 at 2 μm from the film–substrate interface. The average film composition corresponds with x = 0.01. Using polycrystalline (1−x)NiFe₂O₄−xZnGa₂O₄ samples, sintered at 1250°C in oxygen, we have determined \( M_s \) as a function of x (fig. 9). Starting with 265 Gauss for x = 0, \( M_s \) reaches its maximum of 360 Gauss for x is about 0.15. In view of this result, the average \( M_s \) value of the 12 μm film is about 5% higher than the value for pure NiFe₂O₄.

An improvement in the determination of the film volume was achieved by using X-ray fluorescence. The film grown on a (111) facet was dissolved in lithium borate and the amounts of Ni and Fe were determined. Since the film thickness can be assumed as a constant, when interdiffusion occurs, the volume in the (1−x)NiFe₂O₄−xZnGa₂O₄ system. of the film can be calculated from the density of NiFe₂O₄. The obtained accuracy is about 15% resulting in \( M_s = 260 \pm 40 \) Gauss for the film magnetization, which is still too inaccurate to calculate reliably the anisotropy in NiFe₂O₄ films. For NiFe₂O₄ films on MgO 235 Gauss and on spinel substrates 245 Gauss is used in the calculations of \( K_u \).

Torque measurements are performed on NiFe₂O₄ films grown on MgO and on spinel substrates with lattice constants lower and higher than NiFe₂O₄.

### Table 3
Anisotropy data in \( 10^4 \) erg/cm³ for NiFe₂−xAlₓO₄ films; the values of \( M_s \) used in the calculations of \( K_u(\text{exp}) \) are expressed in Gauss

<table>
<thead>
<tr>
<th>Substrate</th>
<th>x (film)</th>
<th>( M_s )</th>
<th>( \eta )</th>
<th>( K^s_\mu ) (( \eta = 0 ))</th>
<th>( K^s_\mu ) (( \eta = 1 ))</th>
<th>( K_\mu(\text{exp}) )</th>
<th>( a_8 - a_f ) (Å)</th>
</tr>
</thead>
<tbody>
<tr>
<td>MgO</td>
<td>0</td>
<td>235</td>
<td>1</td>
<td>−18 ± 4</td>
<td>(−5) − (−12) ± 3</td>
<td>~+0.09</td>
<td></td>
</tr>
<tr>
<td>Zn(Ga, Al)₂O₄</td>
<td>0</td>
<td>245</td>
<td>0.5−1</td>
<td>−30 ± 4</td>
<td>+4 ± 2</td>
<td>(−9) − (+6) ± 5</td>
<td>−0.035 ± 0.005</td>
</tr>
<tr>
<td>ZnGa₂O₄</td>
<td>0</td>
<td>245</td>
<td>0</td>
<td>−4 ± 2</td>
<td>+3 ± 2</td>
<td>(−9) − (+1) ± 5</td>
<td>−0.002 ± 0.001</td>
</tr>
<tr>
<td>Zn(Ga, Fe)₂O₄</td>
<td>0</td>
<td>245</td>
<td>0.2−0.5</td>
<td>+24 ± 5</td>
<td>+3 ± 2</td>
<td>(+26) − (+12) ± 5</td>
<td>+0.027 ± 0.005</td>
</tr>
<tr>
<td>ZnGa₂O₄</td>
<td>0.10−0.12</td>
<td>200</td>
<td>0−1</td>
<td>+9 ± 2</td>
<td>+3 ± 2</td>
<td>(+15) − (+3) ± 5</td>
<td>+0.011 ± 0.001</td>
</tr>
<tr>
<td>ZnGa₂O₄</td>
<td>0.16−0.18</td>
<td>175</td>
<td>0−1</td>
<td>+16 ± 2</td>
<td>+2 ± 2</td>
<td>(+15) − (−2) ± 4</td>
<td>+0.020 ± 0.002</td>
</tr>
<tr>
<td>Zn(Ga, Al)₂O₄</td>
<td>0.16−0.18</td>
<td>175</td>
<td>0−1</td>
<td>−11 ± 2</td>
<td>+3 ± 2</td>
<td>(−11) − (−3) ± 4</td>
<td>−0.014 ± 0.002</td>
</tr>
<tr>
<td>Zn(Ga, Al)₂O₄</td>
<td>0.50−0.55</td>
<td>50</td>
<td>0−0.3</td>
<td>+6 ± 2</td>
<td>+1 ± 2</td>
<td>(+8) − (+4) ± 3</td>
<td>+0.016 ± 0.005</td>
</tr>
<tr>
<td>Zn(Ga, Al)₂O₄</td>
<td>0.55−0.60</td>
<td>35</td>
<td>0−0.5</td>
<td>+7 ± 1</td>
<td>+1 ± 2</td>
<td>(+9) − (+2) ± 3</td>
<td>+0.023 ± 0.005</td>
</tr>
<tr>
<td>Zn(Ga, Al)₂O₄</td>
<td>0.60−0.65</td>
<td>25</td>
<td>0.2−0.7</td>
<td>+7 ± 1</td>
<td>+1 ± 2</td>
<td>(+5) − (+2) ± 3</td>
<td>+0.030 ± 0.005</td>
</tr>
</tbody>
</table>
LPE growth of Ni(Fe, Al)2O4 films (compressively and tensively stressed films), and on Ni(Fe, Al)2O4 films grown in compression as well as in tension on Zn(Ga, Al)2O4 substrates. The results are presented in table 3 and discussed in the following section.

6. Results and discussion

The experimentally found uniaxial anisotropy data (Ku(exp)) presented in table 3 are obtained using eq. (6). From this equation it can be seen that the accuracy of Ku decreases with increasing Ms for a fixed accuracy of KV/MsV. In general the obtained accuracy of KV/MsV increases with increasing Ms. The theoretical anisotropy data (Ku(η = 0) and Ku(η = 1)) are calculated from eqs. (1) and (2) for η = 0 and η = 1. The values of η in the table are deduced from aFe and aFe.

6.1. NiFe2O4 films

Since NiFe2O4 films on MgO substrates are stress relieved at the deposition temperature [3], the stress in the film results from Δa and ΔT on cooling to room temperature. Ku is found to be in the range of −5 \times 10^4 to −12 \times 10^4 erg/cm^3 with an inaccuracy of about 3 \times 10^4 erg/cm^3, compared with (−18 \pm 4) \times 10^4 erg/cm^3 theoretically, taking inaccuracies in Δa and ΔT into account. The difference between Ku(exp) and Ku(η = 1) may be due to some stress relief in the film, caused by cracking. The same difference was found between experimental and theoretical anisotropy data for copper ferrite films [15] grown on MgO.

In 1973, Kanturek et al. [16] have reported about cylindrical magnetic bubbles in NiFe2O4 films grown onto MgO substrates by means of chemical vapour deposition. The induced anisotropy was attributed to the misfit between film and substrate. The existence of bubbles, as well as the given explanation, is in contradiction with the results reported in this article. A severe interdiffusion between film and substrate, resulting in a solid solution between MgFe2O4 and NiFe2O4 and a considerable amount of stress, perhaps results in a positive uniaxial anisotropy exceeding the demagnetization energy of the film. Interdiffusion between NiFe2O4 and MgO and/or the presence of Fe2+ in the film can be concluded from the lattice parameters of “NiFe2O4” films grown by means of CVD [9,17,18], which are in the range of 8.38—8.42 Å compared with about 8.34 Å for pure NiFe2O4. When NiFe2O4 films on MgO are annealed at temper-
temperatures of 1000°C and higher in oxygen, an increase of the film lattice parameter is indeed observed with increasing annealing time.

On flux-grown spinel substrates, NiFe$_2$O$_4$ films are grown between 870 and 900°C (fig. 10). Assuming a saturation magnetization $M_s$ of 245 Gauss, the average film thickness could be determined from $M_sV$ and surface area. In fig. 11 the obtained $K_u$ values, for NiFe$_2$O$_4$ films grown in tension on Zn(Ga, Fe)$_2$O$_4$ substrates, are plotted against the film thickness with the misfit as parameter. It can be concluded that the best agreement between $K_u$(exp) and $K_u(\eta=0)$ is obtained for the thinnest films and the lowest misfits values. The fractional stress relief $\eta$ was found as 0.2 for 0.5 μm films up to 0.5 for 4 μm films. When ZnGa$_2$O$_4$ substrates are used as substrates, a very small misfit is present, and up to about 6 μm film thickness, $\eta$ is about zero. Although, due to the low misfit, the relative inaccuracy in $K_u$(exp) is rather large, the obtained anisotropy data are in reasonable agreement with $K_u(\eta=0)$. A rather high misfit is present when NiFe$_2$O$_4$ films are grown onto Zn(Ga,Al)$_2$O$_4$ substrates resulting in a value of $\eta$ between 0.5 and 1 and $K_u$(exp) between $K_u(\eta=0)$ and $K_u(\eta=1)$.

6.2. Ni(Fe$_{4-x}$Al$_x$)$_2$O$_4$ films

NiFe$_{2-x}$Al$_x$O$_4$ films with $x=0.10–0.12$ are grown in the temperature range of 890–935°C onto ZnGa$_2$O$_4$ substrates. Depending on the film thickness, $\eta$ ranges from zero to 1. Positive anisotropies are obtained, covering the range between $K_u(\eta=0)$ and $K_u(\eta=1)$.

Films with $x=0.16–0.18$ are grown between 928 and 975°C. For dipping times of 10 min, $\eta$ decreases from 1 to zero with decreasing undercooling (fig. 4). The relation between growth temperature and measured uniaxial anisotropy is plotted in fig. 12, showing an excellent agreement between measured and calculated $K_u$ values for films grown in tension as well as in compression.

Uniaxial anisotropies exceeding the demagnetization energies are obtained in films with $x$ between 0.50 and 0.65. The very high inaccuracies in $KV/M_sV$ are compensated by the low value of $M_s$ in the determination of $K_u$. The experimental $K_u$ values are found in reasonable agreement with the calculated stress-induced anisotropy data.

7. Conclusions

Ni(Fe$_{4-x}$Al$_x$)$_2$O$_4$ films are grown successfully by means of liquid phase epitaxy on MgO, ZnGa$_2$O$_4$, Zn(Ga,Al)$_2$O$_4$ and Zn(Ga,Fe)$_2$O$_4$ substrates from a PbO–B$_2$O$_3$ flux, using undercoolings of up to 60°C. Smooth and shiny films are obtained only on (111) substrates. At the surface of the films, grown onto the (111) facets of flux-grown substrates, many growth hillocks are present. The large defect density of the substrate will be one of the main causes for the occurrence of the hillocks.

The compositions of the films are very close to stoichiometry and the Pb content is lower than 0.02 atoms Pb per spinel formula unit. Interdiffusion between film and substrate is observed to a small extent when films are grown on the spinel substrates. It is shown that the saturation magnetization in the NiFe$_2$O$_4$–ZnGa$_2$O$_4$ system increases rapidly with increasing ZnGa$_2$O$_4$ content and a maximum is found at approximately 15 mole% ZnGa$_2$O$_4$. 
Depending on the film thickness and the misfit, cracked or crack-free films are obtained. NiFe$_2$O$_4$ films grown onto MgO substrates are stress-relieved at the deposition temperature. For small film thicknesses and small lattice mismatches, elastically deformed films can be grown without stress relief on the spinel substrates.

From torque measurements it is shown that compressive films have a negative anisotropy, while positive anisotropies are found for films grown in tension. On films with positive uniaxial anisotropies, serpentine-like domain structures are observed, using Bitter fluid. Crack-free NiFe$_{2-x}$Al$_x$O$_4$ films with a uniaxial anisotropy energy exceeding the demagnetization energy are obtained when the misfit is lower than 0.03 Å and 0.50 < x < 0.65. In such films the serpentine-like domains can be contracted to cylindrical magnetic bubbles, with the use of an external magnetic field applied perpendicular to the plane of the film. Evidence is presented that the uniaxial anisotropy results from the misfit stress in contribution with the negative $\lambda_{111}$ magnetostriction constant of the film. The experimental anisotropy data are found to be in reasonable agreement with calculated values of the stress-induced uniaxial anisotropy.

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**References**