Hexaferrite thin films grown by chemical solution deposition on differently oriented sapphire substrates

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Strongly oriented ferrite films were deposited on several substrates, and different substrate/seeding layer/ferrite layer architectures were suggested [e.g.1,2]. In previous experiments, the hexaferrite films were deposited on STO ($SrTiO_3$) substrates with orientations (111), (110), (100), and ceramics. In all cases, (00l) textures of Y-phases of different degrees were observed, but only for the (111) substrate strong in-plane texture was found. For other used substrates, significant out-of-plane and in-plane orientations of Y-films were observed only for $LaAlO_3$ substrates and partially also for MgO substrates.

In this work [3], highly oriented $SrFe_{12}O_{19}$ (M-phase) films with different orientations were grown on Al_2O_3 $(11\bar{2}0)$ - *a*-cut, (0001) – *c*-cut, $(10\bar{1}0)$ - *m*-cut and $(1\bar{1}02)$- *r*-cut single crystal substrates. A procedure involving thin $SrGa_6Fe_6O_{19}$ seed layers was adopted to modify the lattice mismatched interface between film and substrate and allow the growth of oriented films. The typical thicknesses of the seed layer and the M-phase layer reached in one deposition cycle were 25 and 40 nm, respectively. The deposition-flash annealing cycle was repeated until the film of desired thickness was obtained. The final thickness of the seeding layer was 100 nm. $SrFe_{12}O_{19}$ film samples with 1700 nm thickness were used for different analyses. The deposition of the bulk M layer always deteriorates the in-plane orientation (while maintaining the epitaxial relationship) of the originally well-oriented M seed layer. The deposition of the bulk layer M usually results in a broadening of the φ scan peaks by about a factor of two in most cases.

The pseudo-epitaxial growth and orientation relation were verified by X-ray diffraction texture analysis and reciprocal space mapping. Significant out-of-plane textures of (*h*00), (00*l*), and (*hh*0) were found on the *a*-, *c*- and *m*-cuts of substrates, respectively. The FWHM of ω-scans were 0.8-1.2°. Similarly, significant in-plane orientations were found for the *a*- and *m*-cuts of the substrates, respectively while for the *c*-cut they were less good indicating a quite large fraction of unoriented crystallites.

For M layers on *r*-cuts of substrates, the orientation is more complex. For characterization, the wide 3D XRD reciprocal space mapping method was also used [4]. Unlike other cases, the M-layer does not align its c-axis parallel to the substrate c-axis, nor does it grow with a low-index plane exactly parallel to the surface. It is clear, that both conditions cannot be simultaneously fulfilled due to lattice parameter mismatch; these tilts are mutually competing. The system adopts a compromise orientation. A common feature of both the M and Al_2O_3 structures is a sublattice of close-packed planes of oxygen anions; in M, these planes have a complex stacking sequence with local similarities to the spinel structure; in Al_2O_3, they have a hexagonal close-packed stacking sequence. Taking into account the different unit cell definitions for the two structures, the orientation relation should preserve the close-packing of oxygen planes across the film-substrate interface, as well as the directions of close-packing within those planes.

A comparison of texture analysis results obtained on sapphire substrates of different cuts shows that the mutual orientations of the crystal structure of the M film and the sapphire substrates in epitaxial structures are always the same (except for a small angle deviation for an *r*-cut). As a result, there is a high probability that thin films of the M hexaferrite film with an easy axis of magnetization at any angle with respect to the substrate can be attained by making an appropriate choice of the orientation of the Al_2O_3 substrate.

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The authors appreciate the support by the grant of the Czech Grant Agency, no. 24-12710S.