Magnetic parameters of ferrite inclusions observed in magnesium oxide substrates used in nickel ferrite film growth

In a recent electron diffraction study of the oxidation of thin films of $NiFe_2$ produced by sputtering onto single crystal MgO substrates it was reported that, at temperatures of the order of 1200° C, nickel and iron ions diffuse into the substrate [1]. At room temperature these ions come out of solution to form coherent octahedral-shaped ferrite precipitates with a probable nickel ferrite composition. In an attempt to determine the nature of the precipitates from their magnetic parameters these same films have now been investigated using ferrimagnetic resonance techniques.

The films were studied in a standard Q-band spectrometer operating at room temperature and a frequency of the order of 33 GHz. Using small 3 mm diameter disc specimens, observations were made of the position of the main ferrimagnetic resonance as a function of the angle between the external magnetic field (H_z) and crystal axes of the precipitated crystallites. This was done for orientations such that the crystallite [1 0 0] and [1 1 0] type axes were perpendicular to H_z.

The resonance conditions for ferrimagnetic samples possessing magnetocrystalline anisotropy are well known (for example [2]). Since the crystallite separation is large compared to their size the conditions appropriate to disc shaped films are not applicable to the precipitate case under consideration. This should be treated as a matrix of coherent crystallites each with its [0 0 1] axis perpendicular to the plane of the substrate. If the magnetocrystalline anisotropy is taken to have cubic symmetry and the average shape anisotropy of the octahedra to differ little from that of spherical crystallites then the resonance conditions, provided $H_z \gg K_1/M$, are

$$H_r = \frac{\omega}{\gamma} - \frac{K_1}{2M}\left\{\frac{3}{2} + \frac{5}{2}\cos 4\phi\right\} \quad (1)$$

for the case when the static and radiofrequency fields are both in the (1 0 0) plane and

$$H_r = \frac{\omega}{\gamma} - \frac{K_1}{M}\left\{-\frac{3}{16} + \frac{5}{4}\cos\theta + \frac{15}{16}\cos 4\theta\right\} \quad (2)$$

when the static and radio frequency fields are in the (1 1 0) plane, ϕ and θ are the angles between H_z and the [0 0 1] axis in the two cases and H_r is the magnitude of H_z at resonance.

Typical results are shown in Fig. 1 for films orientated with a [1 1 0] axis perpendicular to H_z together with the corresponding theoretical curve from Equation 2. These indicate hard and easy directions of magnetization along the [0 0 1] and [1 1 1] axes respectively, consistent with $K_1 < 0$. The results shown are representative of all the experimental data; the fit between experiment and theory is particularly good indicating that the resonance Equations 1 and 2 are applicable to the samples investigated. From these results it is relatively simple to obtain values of the parameters g and K_1/M for the ferrite concerned and the best fit is given by

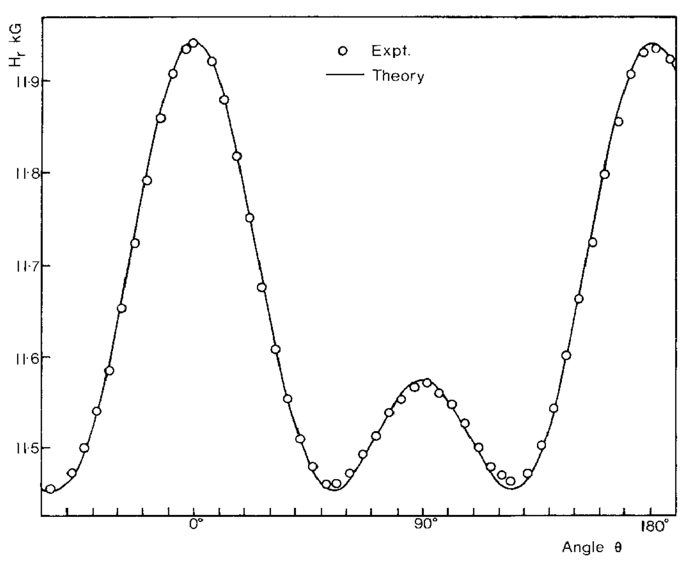

Figure 1 The dependence of the resonance field H_r on the angle θ between H_z and the [0 0 1] axis in the (1 1 0) plane.

$$\frac{K_1}{M} = -144 \pm 3\,\mathrm{G}, \quad g = 2.000 \pm 0.004.$$

These values are not consistent with nickel ferrite.

Unfortunately, the limited data [3–5] in the literature do not permit sufficiently detailed comparison with the above to allow a definite suggestion to be made as to the precise composition of the inclusions.

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Comments on "Fracture and fatigue of discontinuously reinforced copper/tungsten composites"

In a recent paper [1], Harris and Ramani reported some stress strain observations on the composite system comprising copper containing 380 μm tungsten fibres. The authors found "large positive deviations" in the matrix behaviour in composite *vis a vis* that in isolation. In discontinuous fibre composites (25% V_f) they observed yield drops while in continuous fibre composites (V_f's = 6, 8.6 and 16.2%) they did not observe this yield drop phenomenon. It is not clear from their paper whether this enhancement in the matrix stress–strain behaviour was observed in discontinuous fibre composites only, or in continuous fibre composites as well. In their discussion the authors suggested that the occurrence of yield drop indicated that the plastic constraint due to the Poisson ratio difference between Cu and W, envisaged by Kelly and Lilholt [2] in their work on Cu/W system, was responsible for this. Does the absence of this yield drop phenomenon in the continuous fibre composites indicate an absence of this plastic constraint in them? In any case, I think that the authors' explanation of matrix strength enhancement as due to the plastic constraint on the matrix is unfounded for the following reasons. Kelly and Lilholt used very small diameter tungsten wires (10 and 20 μm) and fibre volume fractions that gave them interfibre spacings of a few tens of microns. The small interfibre spacing was chosen, precisely, to enhance the fibre/matrix interaction during straining [2, 3]. For a given V_f, the smaller the fibre diameter, the larger will be the volume of matrix affected by plastic constraint. In Harris and Ramani's case, assuming uniform fibre distribution, the minimum interfibre spacing (which corresponds to 25% V_f) was 380 μm; much too large to give rise to any pronounced plastic constraint. An important interaction between fibre and matrix that leads to enhanced dislocation density in matrix in the as-fabricated stage is the one pointed out by Chawla and Metzger [4], namely, that the thermal stresses due to expansion coefficient mismatch between Cu and W during the fabrication lead to plastic deformation of matrix and thus the matrix in the composite has higher dislocation density to start with; which would result in higher derived matrix strength levels and higher derived matrix strain hardening rates [5, 6]. Chawla and Metzger [4] worked with large diameter W wires (228 μm) and small V_f's (< 15%) in Cu single crystal matrix. The manufacturing process used by Harris and Ramani, namely vacuum hot-pressing, would result in a matrix with a considerably high dislocation density. This interaction between fibre/matrix during processing will be extremely severe in the case of small diameter fibres (and, therefore, small interfibre spacings) and relatively large volume fractions. This is precisely the case wherein the fibre/matrix interaction during straining is maximized too. So, I think in Kelly and Lilhot's work both these effects were present (although Lee and Harris [6] failed to observe the drop in matrix stress–strain curves beyond the fibre yield strain)