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

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) while in the case of Harris and Ramani's work only the fibre/matrix interaction during fabrication would be responsible for the "large positive deviations" in the derived matrix stress–strain curves as compared to the stress–strain curves of the matrix in isolation.

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