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From bulk to film magnetostrictive actuators

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Abstract

Bulk 3d–4f intermetallic compounds exhibiting a giant magnetostriction at room temperature have been developed since 1975 by A.E. Clark. Until recently, attempts to obtain thin films with high magnetoelastic coupling had failed, due to the sperimagnetic character of amorphous R-Fe_2 alloys. Two ways have been recently proposed for overcoming this problem, namely preparing (i) R-Fe_2 nanocrystalline alloys, and (ii) R-Co_2 amorphous alloys. Both methods have shown to be promising while still further improved performances have been reached using composite materials as for example multilayer systems (similar to the spring-magnets), which exhibit giant magnetostrictivity (exceeding 1000 MPa T^{-1} under only 2 mT). We shall review these new materials of interest for microsystems. © 1998 Elsevier Science S.A.

Keywords: Giant magnetostriction; Rare-earth intermetallic compounds; Thin film actuators

1. Introduction

Magnetostrictive 4f metals (like Tb, Dy, Sm) are known to have very large magnetoelastic energy densities, but their strong magnetic anisotropy and their low Curie temperature prevent their use as actuators at room temperature. The goal has been to develop 4f-based alloys with a magnetostrictivity ($\partial\lambda/\partial H$) as large as possible at room temperature, which implied increasing the Curie temperature and decreasing the magnetic anisotropy.

A first approach has been proposed by A.E. Clark [1]: alloying the low-T_c highly magnetostrictive terbium with high-T_c iron, he discovered the so-called Terfenol alloy ($TbFe_2$), a cubic material with a very large magnetostriction at room-temperature; however, this alloy exhibited a fairly large magnetic anisotropy. Then he succeeded in reducing markedly the anisotropy by alloying Terfenol with $DyFe_2$ since both alloys exhibit 4th order magnetic anisotropies of opposite signs, while their magnetostrictions are positive: hence, the Terfenol-D alloy ($Tb_{0,3}Dy_{0,7}Fe_2$) still exhibits a fairly large magnetostriction ($\lambda_S=1.5\times10^{-3}$) at relatively low magnetic induction (0.3 T) near the room temperature. A few types of bulk actuators using Terfenol-D will be briefly reviewed in the second section of this paper.

An alternative approach for obtaining a vanishing magnetic anisotropy consists in preparing amorphous alloys by film deposition on non-magnetic substrates, a promising technique since it is compatible with existing silicon technologies and will favour the integration of mechanical functions (microactuators) into electronic circuits.

However, due to the geometry of films, the principles of actuation are quite different from those developed with bulk samples: Section 3 deals with the specific problems related to the two-dimensional character of the magnetostrictive films.

Various attempts for preparing such amorphous alloys have been published since 1973: magnetostriction of ‘amorphous Terfenol’ has been found to be one order of magnitude lower than that of its crystalline counterpart [2]. Recent magnetic studies have shown that such amorphous Fe-4f alloys could never offer promising magnetoelastic performances, due to the distribution of the signs of the Fe-Fe interactions and their sperimagnetic character. Alternative solutions have been suggested, namely (i) preparing nanocrystalline $TbFe_2$ alloys and (ii) substituting cobalt for iron. Both these approaches will be presented in Section 4 and Section 5 respectively. Composite materials are known to exhibit sometimes outstanding properties which cannot be predicted from the constituent behaviour: actually, a very high magnetostrictivity has been recently observed in TbCo/FeCo and TbFe/Fe multilayers. The basic idea of

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these original materials is somewhat similar to that of 'Spring Magnets' and it offers new opportunities for surface magnetoelastic devices: they will be described in Section 6. As a conclusion, some possible applications of these giant magnetostrictive thin films will be briefly discussed.

2. Bulk magnetostrictive actuators

Most magnetostrictive actuators are based on the use of Terfenol-D alloy, while only a few specific applications, e.g. liquid helium valves, have employed 4f alloys or 4f-Zn compounds which exhibit still larger magnetostrictive strains, but a very low Curie temperature. In this section, we shall recall first the microscopic origin of magnetostriction in 4f metals and alloys, and the main properties of giant magnetostrictive alloys. We shall then illustrate the principles of magnetoelastic actuation with a few specific examples.

2.1. Microscopic origin of magnetoelastic coupling in 4f metals and alloys

The magnetoelastic coupling energy represents this part of magnetic anisotropy energy induced by the strains, and its leading component is the lower even order term, mainly associated in 4f systems with the α_J Stevens coefficient. Even in isotropic substances or in cubic crystals where the second order anisotropy vanishes due to symmetry, this magnetoelastic energy exists, since it represents the derivative of anisotropy energy with respect to strains: a function may be null, while its derivative is not.

The 4f atoms which are the best candidates for large magnetoelastic coupling are given in Table 1. However, the α_J coefficient is not the only parameter to be considered and the magnetoelastic coupling coefficients B result also from other contributions, e.g. arising from symmetry, or from specific properties of the conduction band. For instance, in cubic R-Fe$_2$ alloys, the same α_J coefficient leads to a very large $B^{\varepsilon,2}$ coefficient, while the $B^{\gamma,2}$ coefficient is negligible. On the other hand, $B^{\gamma,2}$ dominates $B^{\varepsilon,2}$ in R-Zn alloys.

Moreover, magnetostriction can be derived formally as $\lambda = -B\langle O\rangle/c$, where O is an angular momentum operator, which varies markedly with the actual L and S values of the metallic ion, and c is an elastic coefficient, which is strongly structure dependent. $\langle O\rangle$ is temperature and field dependent. Finally, economical parameters are also relevant to the choice of bulk actuators: for instance Sm or Tb will be preferred to Tm which is more expensive.

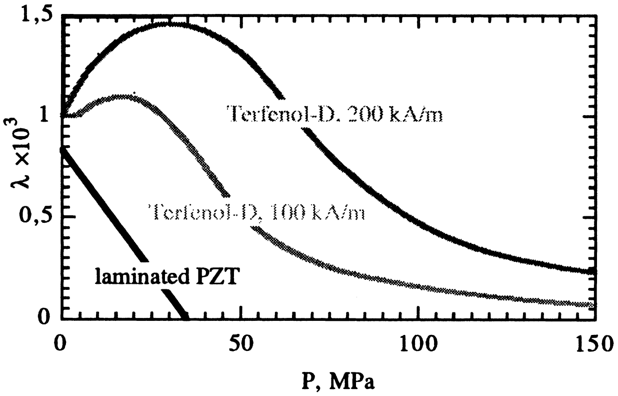

Fig. 1. Dependence of the available magnetostrictive strain $\lambda_{\parallel}$ on the pre-stress P applied to a cylindrical terfenol-D rod; comparison with PZT.

2.2. Physical properties of giant magnetostrictive materials

The main features of Terfenol-D are a large magnetoelastic coupling coefficient $B^{\varepsilon,2}$, a low elastic coefficient c^{ε} and hence a high magnetostriction $\lambda^{\varepsilon,2}$ along the [111] direction. This alloy is fragile, its tensile strength is poor (28 MPa), and it is necessary to apply a moderate compressive stress (a few MPa) along the direction of actuation (Oz) to prevent it from break. Such a stress creates a magnetic easy plane perpendicular to Oz, and increases the available magnetostrictive strain. The magnetic anisotropy is relatively low for this alloy near 300 K but varies rapidly with temperature, and increases strongly below 270K: hence, optimum performances are expected with this alloy only at room temperature.

2.3. The principles of magnetostrictive actuation with bulk materials

Bulk actuators designed with Terfenol-D alloy consist usually of active rods, a few centimeters long, submitted to a bias magnetic field (which may be generated by a permanent magnet) and a driving field created by a solenoid. A moderate pre-stress (0 to 40 MPa) may be applied for increasing the total strain available, as one can see in Fig. 1. Evidently, saturation will necessitate in this case a larger magnetic field to be applied due to the increase of magnetic anisotropy induced by this stress. For comparison, the $\lambda(P)$ characteristic for laminated PZT is

Table 1
Stevens α_J coefficients for some 4f ions

Ion	Ce^{3+}	Pr^{3+}	Nd^{3+}	Sm^{3+}	Tb^{3+}	Dy^{3+}	Ho^{3+}	Er^{3+}	Tm^{3+}
$10^2\alpha_J$	−5.71	−2.10	−0.64	+4.13	−1.01	−0.63	−0.22	+0.25	+1.01

given: the energy density of this piezoelectric material is much lower.

2.4. Some examples of bulk magnetostrictive actuators

A number of magnetostrictive actuators have been designed using terfenol-D alloys [3]. The simplest ones consist of a simple active rod fixed at one end, the length of which may be modulated by the driving magnetic field either quasi-statically for controlling the position of the other end, or dynamically for generating sound waves (sonar) or performing active damping of vibrations.

More sophisticated devices may combine two or more rods, for generating complex motions: for instance, a rotary motor may consist of a free rotor, the position of which is driven by active Terfenol-D rods; when correctly positioned, the rotor is clamped by another set of active rods. A succession of such cycles allows the rotor to turn step by step, with a high torque at low speed.

A third device which merits attention is the magnetostrictive linear motor designed by Kiesewetter [4]: it consists of a simple cylindrical rod moving in a non-magnetic stator tube which has exactly the same diameter. Due to the anisotropic character of magnetostriction, as the rod is elongated its diameter is reduced, since the process takes place at constant volume. The rod is introduced in its stator tube when a sufficient magnetic field is applied, then the field is removed: as its diameter increases, the rod is firmly fixed in its tube. When magnetizing one end of the rod, this end elongates while its diameter is reduced allowing for this elongation. The magnetising field being displaced upwards (see Fig. 2) the strained part of the rod follows this field, until it reaches the upper part of the rod. When switching off the magnetic field in the nth coil, the rod is again firmly fixed in its stator tube, but it has been moved downwards. Repeating this cycle generates a step by step motion of the Terfenol-D rod: this linear motor is able to support large loads with no backlash. In addition, in the absence of electrical power, its position is rigidly fixed-an important security issue.

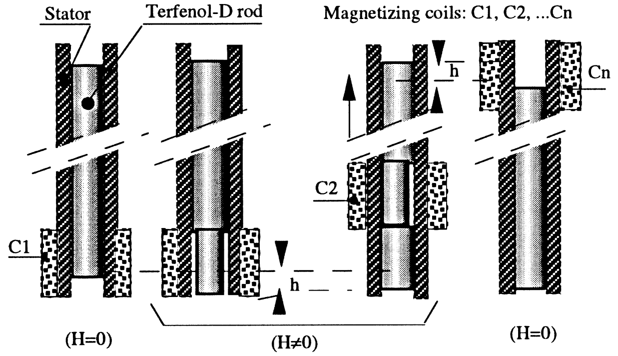

Fig. 2. A linear magnetostrictive motor, after Kiesewetter.

3. The principles of magnetostrictive actuation with thin films

The magnetostrictive strain of a bulk isotropic sample can be written: $\lambda = -b^{\gamma,2}\,\mathrm{f}(H)/c^{\gamma}$, where c^{γ} is the relevant elastic coefficient, $b^{\gamma,2}$ the magnetoelastic coupling coefficient and f(H) a function of the applied magnetic field which will depend on the initial state of the sample, on the magnetic anisotropy of the material, etc... Hence, λ is a function of H. The saturation strain observed when starting from an isotropic demagnetized state is called $\lambda_S = (2/3)\lambda^{\gamma,2} = -(2/3)b^{\gamma,2}/c^{\gamma}$. The magnetoelastic coupling coefficient $b^{\gamma,2}$ and the elastic coefficient c^{γ} are characteristic of the material, hence λ_S is the relevant parameter for describing magnetostriction in bulk samples.

On the other hand, a thin film is necessarily firmly attached to its substrate, and cannot freely expand when submitted to an applied magnetic field, but will try to deform its substrate. The resulting strain of the bimorph (film+substrate) is represented in Fig. 3 for a film with positive magnetostriction magnetized along its length. It may be expressed as a function of the ratio of $b^{\gamma,2}$ and c', where c' is the effective elastic constant of the bimorph. Usually, the substrate is much thicker than the film, and c' is essentially the elastic constant of the substrate and may differ markedly from c^{γ}: hence, λ_S is no longer the relevant parameter for describing the magnetoelastic effects in thin films, but must be replaced by $b^{\gamma,2}$.

The analytical expression of the deflection angle ϑ, has been derived independently by various authors [5–7] in terms of the length L of the sample, of the film and substrate thicknesses (t_f and t_s) and Poisson's ratios (ν_f and ν_s) and of the shear modulus of the substrate G_s:

$$\vartheta = 3\,\frac{t_f}{t_s^2}\,\frac{L}{G_s}\,b(H) \ \text{with}\ b(H) = \frac{1}{2}\,b^{\gamma,2}\left[(\alpha_1^2-\alpha_2^2) - \frac{(1-\nu_s)}{(1+\nu_s)}\,\frac{(1+\nu_f)}{(1-\nu_f)}\left(\alpha_3^2-\frac{1}{3}\right)\right] \tag{1}$$

The α_i's are the direction cosines of the magnetization. This equation defines the $b(H)$ function; the difference

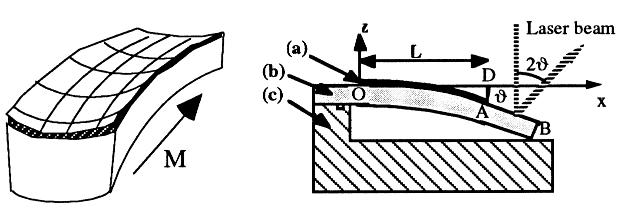

Fig. 3. Deflection under a magnetic field of a bimorph consisting of a non magnetic substrate and a magnetic thin film with a positive magnetostriction coefficient. (left): Anticlastic deformation of the bimorph; (right): experimental set-up for measuring the magnetoelastic effects in such a thin film.

$\vartheta_\parallel - \vartheta_\perp = 3\ [(t_f)/(t_s^2)]\ [(L)/(G_s)]\ b^{\gamma,2}$ (or $b_\parallel - b_\perp = b^{\gamma,2}$) which is independent from the demagnetized state, provides an experimental determination of the magnetoelastic coupling coefficient $b^{\gamma,2}$. The deflection can be easily measured using a laser beam reflected by the free end of the bimorph and detected by a position sensitive diode (PSD) with a resolution of typically 5×10^{-8} radian [8].

An experimental verification of the expression for $b(H)$ has been performed using nickel thin films prepared by sputtering and by high vacuum evaporation [9]. The sputtered Ni-films gave $b^{\gamma,2}$ coefficients that were too small together with low magnetisations, indicating that the density was lower than expected. The films prepared by high vacuum evaporation exhibited both magnetisation and magnetostriction values in excellent agreement with bulk values. However, we observed a surprising result, namely $b_\parallel$ and $b_\perp$ exhibited the same sign. This situation may be easily understood: the deflection ϑ given in Eq. (1), which depends on the magnetization direction, is null for an isotropic demagnetized state ($\langle a_1^2\rangle = \langle a_2^2\rangle = \langle a_3^2\rangle = 1/3$). Otherwise, it can be derived letting:

$$b(H) = b_\parallel = \frac{1}{2} b^{\gamma,2} \left[1 + \frac{(1-\nu_s)}{(1+\nu_s)} \frac{(1+\nu_f)}{(1-\nu_f)} \frac{1}{3} \right]$$

for $\alpha_1 = 1$ (2)

$$b(H) = b_\perp = \frac{1}{2} b^{\gamma,2} \left[-1 + \frac{(1-\nu_s)}{(1+\nu_s)} \frac{(1+\nu_f)}{(1-\nu_f)} \frac{1}{3} \right]$$

for $\alpha_2 = 1$ (3)

$$b(H) = b_0 = -\frac{1}{2} b^{\gamma,2} \left[1 + \frac{(1-\nu_s)}{(1+\nu_s)} \frac{(1+\nu_f)}{(1-\nu_f)} \frac{2}{3} \right]$$

for $\alpha_3 = 1$ (M out of plane) (4)

One may observe that $b_\parallel + b_\perp + b_0 = 0$, and $b_\perp = b_0$ when $\nu_f = \nu_s$.

When very strong in-plane stresses are present, the demagnetized state is defined either by $\alpha_3 = 1$, or by $\alpha_3 = 0$, depending on the sign of the stress.

In the former case, the spontaneous magnetostrictive deflection is the one observed when the film is magnetised out of its plane $\alpha_3 = 1$, i.e. it is characterised by $b(H) = b_0$. When magnetising the film in its plane, the observed deflections may then be expressed using:

$$b_\parallel - b_0 = \frac{1}{2} b^{\gamma,2} \left[1 + \frac{(1-\nu_s)}{(1+\nu_s)} \frac{(1+\nu_f)}{(1-\nu_f)} \right] = b^{\gamma,2} \frac{1-\nu_s\nu_f}{(1+\nu_s)(1-\nu_f)} \qquad (5)$$

$$b_\perp - b_0 = \frac{1}{2} b^{\gamma,2} \left[-1 + \frac{(1-\nu_s)}{(1+\nu_s)} \frac{(1+\nu_f)}{(1-\nu_f)} \right] = b^{\gamma,2} \frac{\nu_f - \nu_s}{(1+\nu_s)(1-\nu_f)} \qquad (6)$$

$b_\perp - b_0$ may exhibit the same sign as $b_\parallel - b_0$ provided the Poisson's ratio of the film is larger than that of the substrate: this differs markedly from the case of bulk samples. Evidently, $b_\parallel - b_\perp$ remains always independent from the initial state and provides the $b^{\gamma,2}$ coefficient.

For applications as microactuators, it is the difference $\vartheta_\parallel - \vartheta_0$ (i.e. $b_\parallel - b_0$) which will define the performances of the device: the demagnetized state should be reproducible.

4. Nanocrystalline R-Fe (R=Tb-Dy) magnetostrictive alloys

Provided the dimensions of crystallites are sufficient for allowing the exchange coupling to be effective but small enough for preventing any macroscopic anisotropy energy to be observed, one may consider the material as an isotropic ferromagnet, the magnetostriction of which is just the one expected for a polycrystalline material: magnetostriction is not reduced but magnetocrystalline anisotropy has disappeared. It is likely that this is just the situation in amorphous R-Fe$_2$ films prepared at Stuttgart which have been annealed (see Fig. 4) [10]. Only in a very small temperature range the magnetostriction increases where the crystallisation takes place. Up to now, this effect was unfortunately linked with an increase in coercive field which indicates that the structure is not really nanocrystalline. Research is in progress for preventing the crystallites from growing up, and better performances are expected in the near future.

5. Amorphous Tb-Co-based magnetostrictive alloys

5.1. Sperimagnetic structure in 3d–4f amorphous alloys

The random anisotropy model of Harris, Plischke and Zuckermann [11] describes the competition between exchange interactions and random anisotropy in amorphous alloys. This model accounts for the sperimagnetism, i.e. the non collinear arrangement of rare earth magnetic

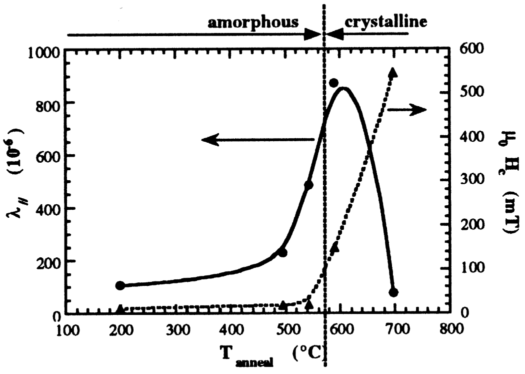

Fig. 4. Magnetostriction (circles) and coercive field (triangles) as functions of annealing temperature for amorphous Terfenol-D alloys.

moments in 3d–4f alloys due to the large random anisotropy of rare-earth atoms (R). The Fe–Fe exchange interactions in amorphous $R_{33}Fe_{67}$ alloys may be either positive or negative depending on the interatomic distances: due to such a competition between the ferro- and antiferro-magnetic interactions, one observes a spin-glass structure in the Fe sub-lattice. As a consequence of this magnetic frustration, the Curie temperature and the room temperature magnetoelastic coupling in Fe-based amorphous alloys are markedly reduced, as compared to those observed in several crystalline RFe_2 alloys (R=Tb, Dy, Sm). In addition, a large high field susceptibility is observed.

This is not the case with R-Co alloys where all cobalt moments are strongly ferromagnetically coupled and are colinear. Fig. 5 shows the magnetic structure of Tb-Co and Nd-Co amorphous alloys, as compared with the corresponding Fe alloys. One may expect better magnetoelastic performances with R-Co alloys. The Co-rich $Tb_{1-x}Co_x$ amorphous alloys exhibit a higher Curie temperature than the $Tb_{1-x}Fe_x$ amorphous alloys for $x>0.67$. In Tb-Co alloys, the compensation temperature is 300 K for $x=0.78$, and the terbium moment is dominant for $x<0.78$.

5.2. Magnetostriction in R-Co_2 and R-$(CoFe)_2$ amorphous alloys

The magnetoelastic coupling coefficient of $Tb_{1-x}Co_x$ amorphous alloys is negative which means that the magnetostriction is positive. It reaches its maximum absolute value near $x=0.7$ ($b^{\gamma,2}=-24.5$ MPa). Assuming that the shear modulus of this alloy is the same value as that of its polycrystalline counterpart, this gives a magnetostriction coefficient $\lambda^{\gamma,2}=3/2\ \lambda_s=4\times10^{-4}$.

Fig. 6 shows the field dependence of the functions $-b(H)$ and $\lambda(H)$ for a $Tb_{0.32}Co_{0.68}$ alloy. Again, as in the case of nickel thin films, one observes that $b_\perp-b_0$ exhibits the same sign as $b_\parallel-b_0$ due to the low value of the Poisson's ratio for the silicon substrate along $\langle110\rangle$, and to the existence of a perpendicular magnetic anisotropy in the film: for $H=0$, most of the moments point out of the film plane.

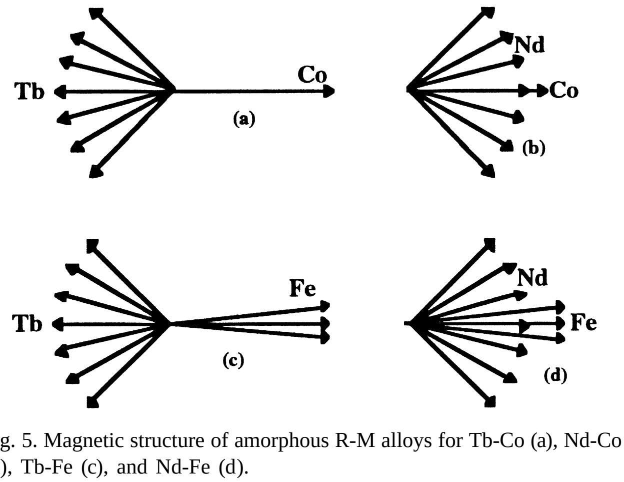

Fig. 5. Magnetic structure of amorphous R-M alloys for Tb-Co (a), Nd-Co (b), Tb-Fe (c), and Nd-Fe (d).

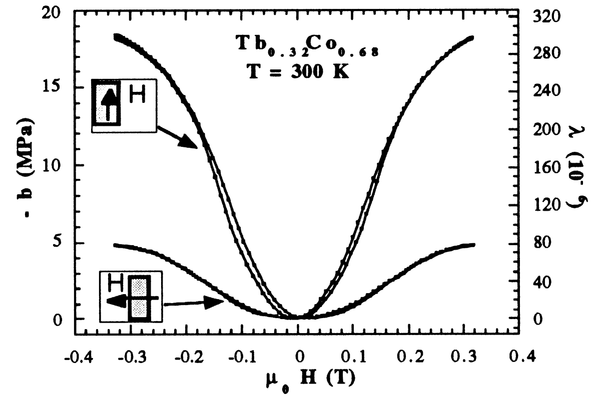

Fig. 6. Magnetostriction for $Tb_{0.32}Co_{0.68}$ with perpendicular anisotropy in two field directions in the plane.

A few Sm-Co amorphous alloys have been also studied [12]: the magnetoelastic coupling coefficient is positive (negative magnetostriction) and somewhat smaller (by 30% – 50%) than that of Tb-based alloys.

Substitutions of iron for cobalt in $Tb_{1-x}Co_x$ amorphous alloys may enhance the magnetoelastic coupling coefficient by a factor of 2: for an amorphous alloy $Tb_{0.32}(Fe_{0.45}Co_{0.55})_{0.68}$ after an annealing at 250°C, one observes $b^{\gamma,2}=-40$ MPa for 0.3 tesla, and -63.5 MPa for 2 T. However, the low field magnetoelastic susceptibility remains too small for technical applications.

6. Magnetostrictive multilayers

Most traditional attempts to reduce the driving fields required for giant magnetostrictive rare-earth transition metal (R-T) materials have been based around techniques for reducing the macroscopic anisotropy. However, the saturation field, H_s, can also be reduced by increasing M_s, the saturation magnetization (provided there is no associated increase in anisotropy). Giant magnetostrictive materials used up to now have rather low magnetizations due to their ferrimagnetic nature. For the compositions of interest, the *R* moments dominate and so an increase in the *T* component in the alloy will only further reduce the magnetization. On the other hand, an increase in the *R* component results in a lowering of the Curie temperature and is thus inappropriate for room temperature applications. It is thus difficult to see how M_s can be notably increased using homogeneous alloys of this type. However, using multilayers, it is possible to engineer new composite or 'artificial' materials which have properties that cannot be achieved with homogeneous alloys.

The new idea is to take two materials, one material having a large room temperature magnetostriction (but low magnetization) while the other is magnetically very soft and having a very high magnetization. The layers must be magnetically coupled and thinner than the magnetic exchange length. For these layers, domain walls cannot be formed at the interfaces and the magnetization processes will be an average of those of each individual layer. In particular the saturation field of such a system can be considerably reduced while keeping relatively large values of the magnetostriction. To illustrate this point, we can consider the simple case of a multilayer system of this type with layers of equal thickness in which we have a perfectly rigid coupling across the interfaces and therefore an unique magnetization axis throughout the sample thickness (only the magnetization direction along this axis can change from layer to layer depending on the sign of the exchange coupling). The soft material has a magnetization M_s^{himag}, zero magnetostriction and negligible anisotropy while the magnetostrictive material has a magnetization M_s^{MS}, a magnetostriction λ and an anisotropy of K^{MS}. Typically M_s^{himag} can be 5 M_s^{MS} giving a magnetization for the composite material of 3 M_s^{MS} if both layers are coupled ferromagnetically. The anisotropy of the composite multilayer material halved and so the saturation field is $H_s^{\mathrm{MS}}/6$. For the magnetostriction, even if the soft material has zero magnetostriction, the magnetostriction of the composite multilayer material will be $\lambda/2$. Thus λ/H_s for the composite material is 3 times that of the homogeneous magnetostrictive alloy.

The advantage of these composite multilayer materials over the normal homogeneous alloys is that each layer of soft material is Fe or FeCo which have very large magnetization values. Alloys with R additions always have much lower magnetization values at room temperature. In addition to this, increasing magnetization by closing the cone angle in a sperimagnetic structure is an efficient means of increasing the low field magnetostrictivity, since (i) this decreases the saturation field (K/M) and (ii) the magnetoelastic coupling is strongly correlated to the mean value of the magnetic moments correlation function. Increasing the molecular field should be a pretty way for achieving this process.

When alternating layers of amorphous Tb-Co and crystalline Fe-Co alloys are successively deposited, the low molecular field of Tb-Co may be strongly increased in the very neighborhood of the strongly magnetic Fe-Co layer. If layers are sufficiently thin, a noticeable volume of the Tb-Co layer may be submitted to the large molecular field and this increases both its magnetization and its magnetoelastic coupling.

A very large magnetostrictivity has been observed in multilayers defined as: $Fe_{0.75}Co_{0.25}/Tb_{0.18}Co_{0.82}$: (6.5nm/4.5nm)*3, which means three series of bilayers Fe-Co and Tb-Co, respectively 6.5 and 4.5 nm in thickness (Fig. 7).

The largest peak (4800 $\mathrm{MPa\,T^{-1}}$) is not reversible, while the smaller one is. When a minor loop is described around the zero field, the reversible magnetostrictivity is still 800 $\mathrm{MPa\,T^{-1}}$, which is larger than the value (570 $\mathrm{MPa\,T^{-1}}$) observed with bulk Terfenol-D [8].

A very large magnetostrictivity has been reached also by Quandt and his co-workers with TbFe/Fe multilayers. For instance, a reversible peak as large as 1000 $\mathrm{MPa\,T^{-1}}$ has been reported [13] for a $Tb_{0.4}Fe_{0.6}$(4.5 nm)/Fe(6.5 nm) multilayer.

7. Conclusions

In the recent past, significative improvements have been achieved in the field of giant magnetostrictive films. It has been possible to design and test some prototypes of microactuators.

One of the main drawbacks of magnetoelastic actuators is their thermal drift. For bulk Terfenol-D actuators, performances are thermally-dependent since magnetocrystalline anisotropy varies strongly with temperature, and the thermal expansion (10^{-5} $\mathrm{K^{-1}}$) develops strains comparable to magnetostrictive ones for $\Delta T = 150$ degrees.

For thin films, at present, one knows only a little about the temperature dependence of the magnetoelastic coupling. But the comparison between the strains due to the thermal expansion and to the magnetoelastic coupling will depend strongly on the actual geometry of the actuator:

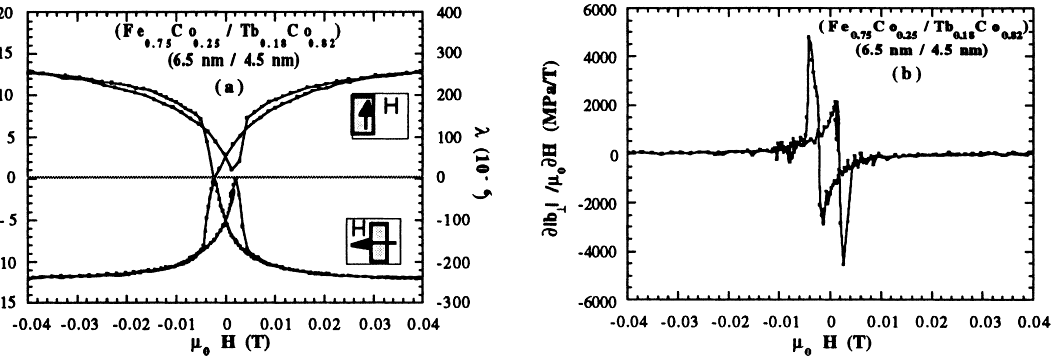

Fig. 7. Low-field magnetoelastic properties in the two configurations ↑H and H← for the multilayer system $Fe_{0.75}Co_{0.25}/Tb_{0.18}Co_{0.82}$: (6.5nm/4.5nm)*3 (left) and its magnetoelastic susceptibility (right).

these strains are comparable for only $\Delta T = 75$ degrees with the best multilayer material when a simple rectangular bimorph fixed at one end has been chosen as an actuator: films appear to be still more sensitive to thermal drifts than bulk materials.

But the influence of thermal drifts may be drastically reduced if one chooses for the actuator the geometry shown in Fig. 8, namely a square plate fixed at three corners A, B, C: the fourth one, D, is free to move. Any temperature change will deform this plate into a spherical shell, but by virtue of the figure symmetry, the D corner should not move. On the other hand, applying a magnetic field along the BD diagonal will induce a marked shift of this end. Simulations and analytical derivations have shown that the actual efficiency of such a device will depend on the quality of the three hinges. A decoupling factor of 50 should be the maximum possible value for hinges which have 2 degrees of freedom in torsion provided they are sufficiently soft in the plane (see Fig. 8).

Micropumps for medical drug delivery, microscalpels, and any micro-actuators for biological applications seem to be one possible area of applications for such devices, since the required energy is very small; a significant advantage of such microactuators, as compared to piezo-electric ones, is their remote command: the magnetic energy is transmitted without any wires or connections. Another kind of possible applications is for integrating in microsystems some micro-switches locally excited by microcoils.

One should note that presently materials are not yet optimised and one may still expect some good news in the field of magnetoelastic multilayers and composites.

Going away from technical considerations, we must stress that studies on magnetoelastic effects in thin films present also a fundamental interest. Magnetoelastic coupling is very dependent on the symmetry of the material: at the surface of any solid, symmetry is broken since there are no longer any neighbouring atoms. For instance, symmetry which was spherical in an isotropic amorphous substance becomes cylindrical at the surface: magnetoelastic coupling will exhibit two contributions, the one from the bulk, and the other one from surface, with a higher number of coupling coefficients. This latter contribution is usually ignored since the number of concerned atoms is negligible in a bulky sample, but this is no longer true for very thin films where surface effects are essential. Moreover, magnetization exhibits a faster decrease with temperature in very thin films, and this causes also a rapid thermal dependence of the magnetoelastic coupling coefficients. Finally new structures, which cannot be prepared in the bulk state, may be prepared in vacuum deposited ultra-thin films or multilayers, and open up new exciting research areas.

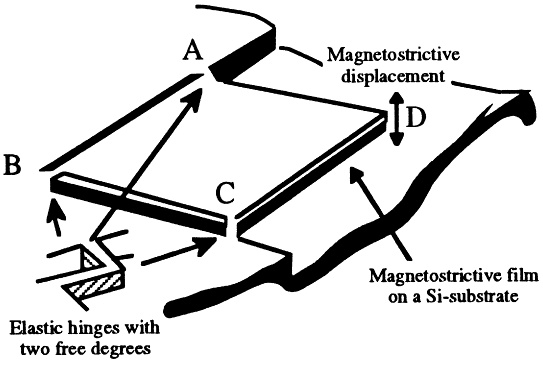

Fig. 8. Magnetostrictive microactuator compensated for thermal drifts; insert: hinges with two degrees of freedom.

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