Properties of polycrystalline NiMnGa – alloys for application as magneto-mechanical Actuator

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Abstract:
Magnetic shape memory materials can deform by some percent via twin boundary motion under the influence of a magnetic field. This effect is well documented for single crystals of Ni-Mn-Ga alloys with compositions around Ni2MnGa. It is demonstrated that this effect can be achieved in textured polycrystalline samples as well, which can be produced much more efficient than single crystals. Directional solidified samples were subjected to thermal and thermo-mechanical treatment in order to adjust structure and microstructure. A shape change larger than 1% was achieved in a Ni50Mn29Ga21 alloy by applying a magnetic field at room temperature.

Keywords: magneto-mechanical actuator, magnetic shape memory alloy, polycrystalline NiMnGa

Introduction
Magnetic shape memory materials have gained a large research interest owing to their capability to deform by some percent via twin boundary motion under the influence of a magnetic field. Concurrently, they are supposed to react faster than conventional shape memory materials because neither heating nor cooling are involved. The predominant material system is Ni-Mn-Ga with compositions around Ni2MnGa in the form of single crystals [1]. Recently we used Ni50Mn30Ga20 and Ni50Mn29Ga21 alloy samples to produce polycrystalline textured samples. Utilising the unique properties of textured samples our goal is to overcome the use of single crystals, which are expensive and the fabrication of which takes a lot of time. To obtain the texture in these alloys, samples are hot mold cast for directional solidification. Upon cooling these alloys transform martensitically at 100 °C and 55 °C, for Ni50Mn30Ga20 and Ni50Mn29Ga21 respectively. The Ni50Mn29Ga21 alloy samples had a 5 fold modulated (5M) martensitic structure. Mechanical training of this alloy in the martensitic state may lower the detwinning stress below the maximum magnetic field inducable stress. Thus, magnetic field induced strain (MFIS) can be observed.

Experimental
A master alloy ingot with a diameters of 15 mm was prepared by induction melting of the pure elements (Ni 99.98, Mn 99.8, Ga 99.999) in Ar atmosphere followed by casting into a cold copper mold. The mass loss was negligible, hence the composition is assumed to be the intended one. Pieces of those master alloys were remelted and cast into a preheated ceramic mold with a heat sink at its bottom. Thus, the solidification front moves upwards and a coarse-grained columnar structure arises with the [100] direction preferred along to the solidification direction. The obtained ingots are of 10 mm diameter and 50 mm length. Finally, electro-discharge machining was used to cut samples of the desired shape, e.g. cubes of 5 to 7 mm edge length from the ingots. Mechanical training was performed by cyclic compression along the axes of those cubes. Magneto-mechanical behaviour was characterized by a mechanical testing machine which was additionally equipped with a magnetic yoke with a maximum field of \( \mu_0 H = 0.75 \) T. Structure was determined by XRD and microstructure was characterised by optical microscope and SEM (scanning electron microscope) with EBSD (electron backscatter diffraction). All structural features depend on preparation conditions [2, 3].

Results

Microstructure
The principle of the casting set up is displayed in Fig. 1. A hot mold is mounted on a cooling copper plate. To avoid heating of the cooling plate by the hot mold, an insulation ring is placed between the mold and the plate. When the melt is cast into the mold, a temperature gradient and thereby a heat flow through the melt toward the bottom is generated, which causes a directional solidification in opposite direction. The texture occurs along the direction of
the heat flow, preferentially in [100] direction [4]. The influence of various casting and annealing parameters was demonstrated in [5]. The microstructure in the as cast state and after annealing at 1000 °C for 48 h is shown in Fig. 2 together with differential scanning calorimetry (DSC) – plots.

Four different regions of the microstructure of the samples can be distinguished. In the first very narrow zone at the bottom of the mold fine almost globular grains are observed. This is due high cooling rate at the beginning of the solidification. In the second zone columnar grains can be seen. The size of this zone depends on the casting parameters, especially the mold temperature. With increasing mold temperature the area of elongated grains becomes larger [2]. At the top of this zone a conical region can be seen, where in the centre still columnar grains can be observed whilst globular grains are on the rim of the sample. There, the solidification direction changes gradually from unidirectional (to the copper plate) to radial (to the mold). Above this cone in zone 3 all the grains are globular. In the fourth zone sink holes and other cavities due melt shrinkage are observed. Samples are taken from the zone with columnar grains which is marked in Fig. 3.

Annealing at 1000 °C for 48 h in Ar-5% H₂ atmosphere causes grain coarsening and homogenization cf. Fig. 3. In the DSC curves in Fig. 3 the transformation peaks become sharper after the annealing process. This is due to homogenization, stress relaxation and possible ordering during cooling. Is to be mentioned, that the martensite transition temperature does not vary for samples taken along the direction of solidification [2]. This demonstrates, that the chemical composition of the ingot does not vary significantly along its length because the martensite temperature depends very strongly on composition [6]. EBSD measurements on annealed samples showed, that the columnar grains are preferably aligned along the direction of solidification with an [100] axis (Fig. 3).

**Mechanical training**

Plastic deformation of martensitic NiMnGa alloys by dislocation motion is not possible. Dislocation motion is hampered strongly due to atomic ordering and the low crystal symmetry of theses alloys. This causes large Burgers’ vectors for complete lattice dislocations. And since the stress to move a dislocation is proportional to \( \exp(-1/b) \) (\( b \) = absolute value of the Burger’s vector) dislocation motion is hampered in ordered systems. Therefore, the main source of plastic deformation observed in such alloys is the movement of twin boundaries [7]. Magnetic field induced strain (MFIS) in ferromagnetic shape memory alloys is also a result of twin boundary movement in the special case as a result of an external magnetic field. Thus, plastic deformation provides information about the mobility of the twin boundaries. Furthermore, it is expected,
that cyclic plastic deformation increases the mobility of the twin boundaries.

Cubes of 5 to 7 mm edge length were cut from the columnar region of directionally solidified and annealed ingots. Such cubes were cooled through the martensitic transformation under 30 MPa load. A strain of typically 1–2% is achieved by this procedure which means that many twins remained in the sample, since a detwinned sample would cause a strain of \( \varepsilon_{\text{trans}} = 5\% \) during the martensitic transformation \( \varepsilon_{\text{trans}} = 1 - \Delta \alpha / \varepsilon_{\text{max}} \). The samples were successively compressed all three axes in order to cause maximum deformation. This means a sample that was compressed along axis 1 is pressed along directions 2 and 3, before the next compression along axis 1 is performed. (Fig. 4). The influence of successive training for a 7M martensite is described in [8]. A clear effect of the training can be observed. The residual strain increases to 8%, which is rather close to the crystallographic strain limit \( \varepsilon_{\text{max}} = 1 - c/a \approx 10\% \) for this structure. It was also found that most of the strain is produced at comparably small loads. However, the strain at 2 MPa, which is approximately the stress an external magnetic field can generate \([9]\), is very low. Thus, the polycrystalline 7M martensite seems not to be suited for magnetomechanical actors. Additionally these samples withstand only about 5 – 10 training cycles before they fail by fracture.

The effect of training a 5M martensite is given in Fig. 4. Training of the 5M martensite is much more efficient than training of the 7M martensite. It is possible to apply a large number of training cycles without any harm to the samples. The maximum stress which can be generated by a magnetic field is about \( \sigma_{\text{mag,max}} \approx 2.6\text{ MPa} \) for the 5M martensite \(^{10}\). This value is higher than the corresponding value of the 7M martensite due to the lower crystallographic strain of the 5M martensite which is \( \varepsilon_{\text{max}} = 1 - c/a = 6\% \). There is a considerable large amount of strain at stress values smaller than \( \sigma_{\text{mag,max}} \) for the 5M martensite. Therefore, we can expect an effect of a magnetic field on the stress – strain curve of such samples.

Magneto-mechanical response

Trained samples of the 5M martensite were subjected to mechanical testing with and without a magnetic field. The axis of the magnetic field was perpendicular to the axis of the mechanical load. The axis of the columnar grains was parallel to the load axis. I turned out, that there is a considerable influence of the magnetic field on the mechanical behaviour. Fig. 5 shows unloading curves for a sample made from trained 5M martensite taken at 50 °C. The initial load was 2 MPa and the strain upon stress release was recorded. There strain observed during stress release with a magnetic field of 0.7 T is about 1 % larger than the strain observed when no magnetic field is applied. Thus, a MFIS of 0.3 % was found for the trained 5M martensite.

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**Fig. 4**: Stress and strain for successive compression along all three axis for 5M martensite. Curves are given for one axis only.

**Fig. 5**: Stress release with and without magnetic field for trained 5M martensite at 50°C. The inset gives the direction of load and magnetic field.

**Fig. 6**: Influence of a magnetic field on the stress-strain curve during compression. A magneto-strain of 2.07 MPa is observed.
Fig. 6 shows a stress–strain curves of a well trained sample taken when a magnetic field is switched on for some time. The magnetic field induces a magneto-strain of $\sigma_{\text{mag}} = 2.07 \text{ MPa}$ which is smaller than the maximum stress, $\sigma_{\text{mag,max}} \approx 2.6 \text{ MPa}$ which can be induced by a magnetic field in the 5M martensite. The difference is due to the residual misalignment of the grains relative to the stress and field axes, and those misaligned grains which do not contribute to the magnetic stress but take cross section.

References


