Mechanical Properties of Fe–Ni–Co–Ti–Shape Memory Alloys

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Introduction

It has been shown that Fe–Ni–Co–Ti alloys exhibit the shape memory effect (1–3). The effect can only be observed after an ausageing procedure, which leads to a two–phase microstructure consisting of austenite and coherent and ordered γ'–$(Ni,Co)_3Ti$–precipitates. Since the size of the particles is below a critical value d_c, the precipitation of γ' leads to a considerable decrease of the temperature hysteresis. The small particles shear into a metastable, tetragonal distorted structure α' during the transformation (4,5). α' is less stable than γ', so that more energy is required for the martensitic transformation.
According to that the reverse transformation is supported by regaining the invested energy. The small hysteresis ($\Delta T \approx 40°C$) can not be found in the homogeneous state.

In addition, the γ'–precipitates influence the martensite–start–temperature M_s in two ways:

1. A fine dispersion of γ' leads to a very efficient hardening of the austenite and thereby to a decrease of the M_s temperature;
2. The precipitation of γ' change the composition of the austenitic matrix. Consequently, the constituent of nickel decreases and the M_s–temperature is raised (4).

The aim of this work is to define the basic parameters of the mechanical properties of this new shape memory alloy and to discuss them in respect of possible applications.

Experimental procedure

Four different Fe–Ni–Co–Ti–alloys were melted, hot rolled and vacuum homogenized for t = 25 h at T = 1250 °C (table 1).

Alloy	Ni	Co	Ti	Fe
A	29.2	9.9	3.8	bal
B	29.9	14.85	3.9	bal
C	31.9	12.1	3.94	bal
D	26.1	20.4	3.85	bal

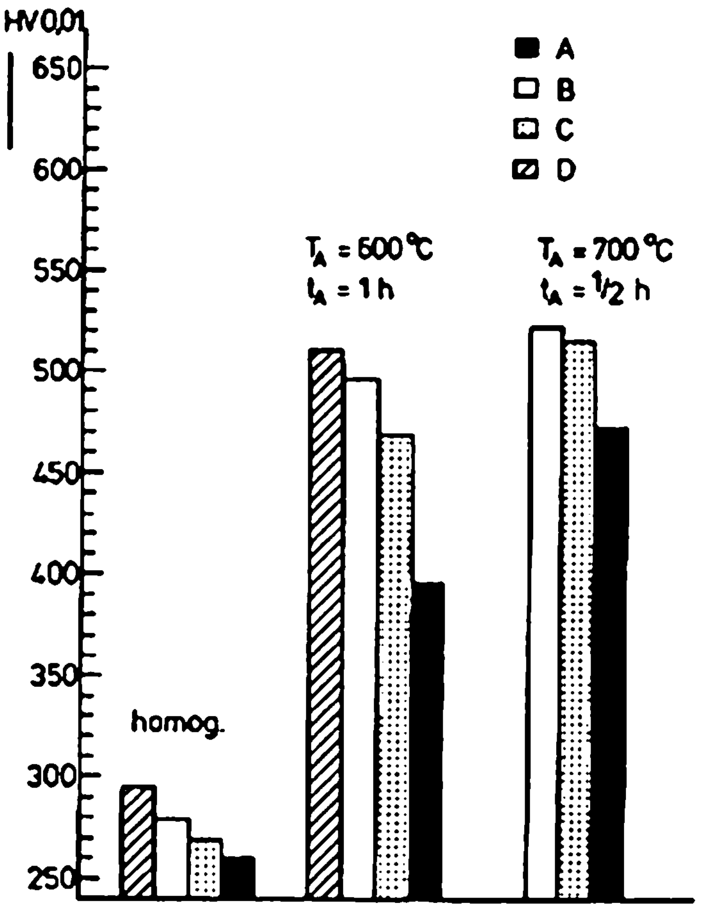
Table 1: Composition of the four investigated alloys A, B, C and D in wt.% Determined by EDX–analysis

The ausageing treatment was carried out for different times at T_A = 600 °C and T_A = 700 °C with subsequently water quenching. Some specimens were additionally deformed by hot rolling between homogenization and ausageing treatment. In order to characterize the transformation temperatures, dilatometric measurements were conducted. The mechanical properties were studied by means of tensile tests at different temperatures and Vickers–microhardness measurements. In addition, the microstructures were investigated by light– and scanning electron microscopy, whereby the morphologies of the martensite were of special interest. A special light microscope was used during the tensile tests in order to investigate the nucleation and growth of the stress induced martensite.

Results and discussion

Ageing causes a homogeneous precipitation of γ' resulting in an increase of the austenite–microhardness (Fig. 1).

The formation of martensite leads to an additional elastic and plastic deformation of the residual austenite which results in a further increase of the hardness (Fig. 2). After a complete reverse transformation the austenite hardness is at a higher value as it was before the mar-



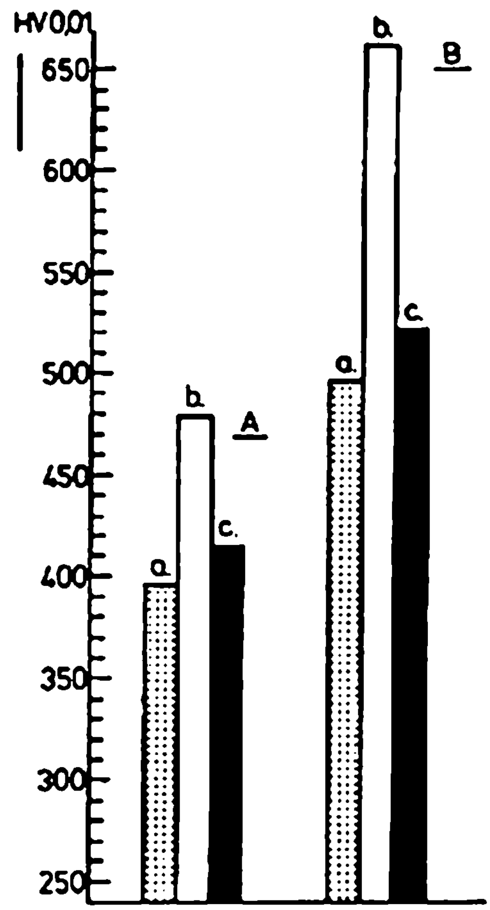


Fig. 1

Fig. 2

Fig. 1: Austenite microhardness HV 0.01 of the four different alloys after heat treatments.

Fig. 2: Austenite microhardness HV 0.01 of A and B; a) Hardness of ausaged specimens (1 h/600 °C), b) Hardness of residual austenite after cooling the specimens down to − 196 °C, c) Austenite hardness after the complete reverse transformation ($T \geq A_f$)

For a grain size larger than 150 µm the tensile strength of all alloys in the homogeneous state is about 500 – 700 N/mm². The yield strength was measured to 350 – 450 N/mm², the elongation at fracture amounts to about 30 %. Martensitic specimens do not show an elastic deformation corresponding with Hooke's law. This can be explained by martensitic reorientation depending on the applied stress tensor and stress induced transformation of the residual austenite.

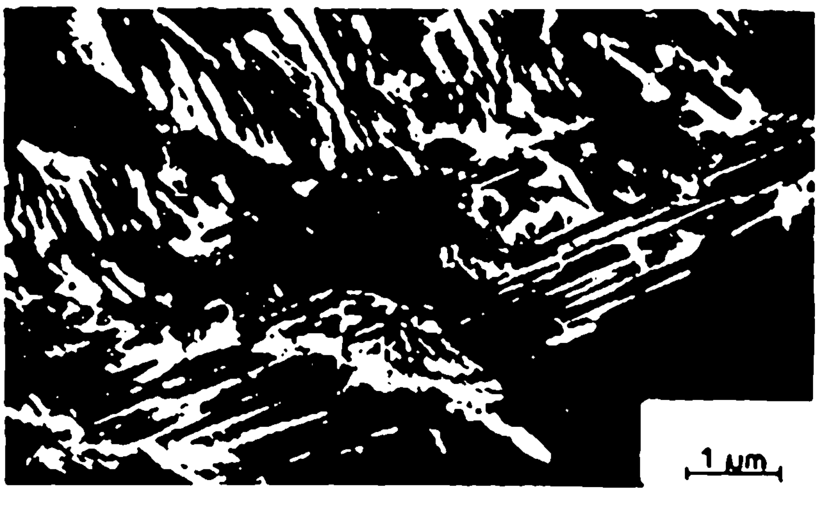


Fig. 3: η-phase at the grain boundaries, alloy C (1 h/ 700 °C)

Ausaged conditions show a more extensive yield point but a much lower elongation at fracture ($A \approx 0.2$ %). The reason for this very brittle behaviour is an inhomogeneous precipitation of the stable, hexagonal η–phase (Ni_3Ti) especially at the grain boundaries (Fig. 3). That leads to the observed intergranular fracture (Fig. 4a).

Considering the importance of the ausageing treatment this brittleness must be absolutely prevented for every kind of application.
Hot rolling between homogenization and ausageing treatment with subsequently water quenching impedes the intergranular failure of the ausaged conditions. The dislocation density increases providing a large number of nucleation sites for the precipitates, whereby the precipitation of η–phase at the grain boundaries is reduced. The elongation at fracture rises up to $A = 14$ % for a previous reduction in thickness of 70 % at $T = 1200$ °C. The fracture behaviour is characterized by dimples on the surface (Fig. 4b). The alloys treated in this way exhibit a tensile strength above 1000 MPa.

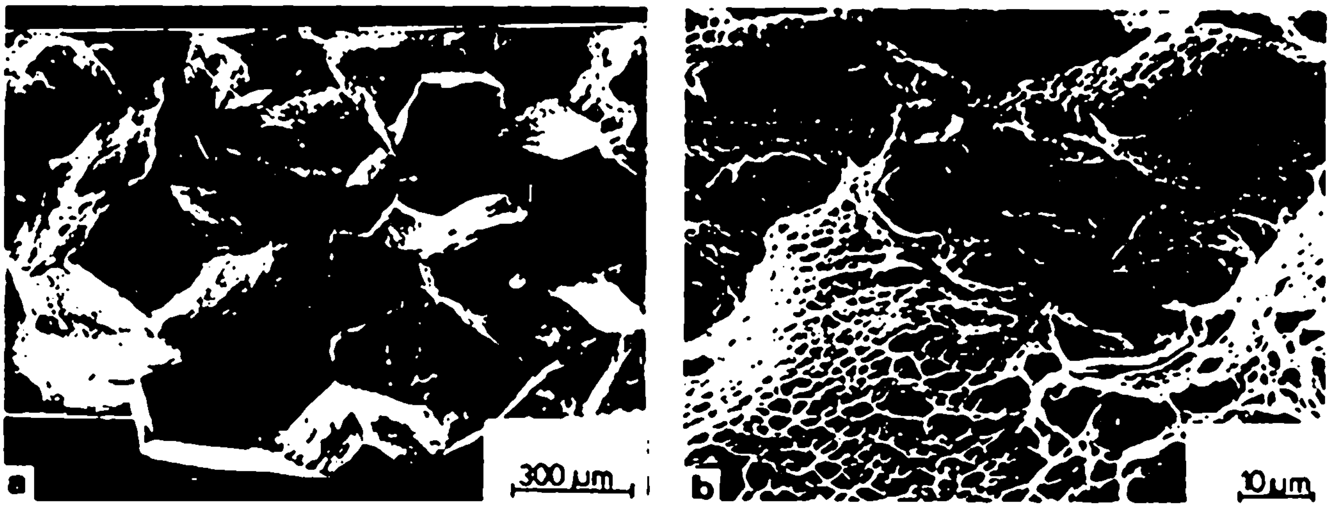


Fig. 4: a) Intergranular brittle fracture of ausaged specimens, alloy A (1 h/600 °C), b) ductile fracture characterized by dimples, alloy B, 50 % prior deformation at 1200 °C (1 h/600 °C)

The transformation characteristic of the ausaged Fe–Ni–Co–Ti–alloys leads to the assumption that these alloys show a one way shape memory effect due to stress induced transformation of martensite with a very low thermal hysteresis. The formation of the stress induced martensite depends on the testing temperature as it can be observed for other alloy–systems, too (6). Approaching the M_s–temperature, the requisite stress for the transformation decreases, while the tensile strength increases (Fig. 5).

The orientation of the stress induced martensite depends on the stress tensor and the crystal orientation of the grain (Fig. 6). The morphology of this martensite is thin plate like and strictly parallelly orientated opposite to the thermal induced martensite. The martensite orientation of hot rolled specimens is in accordance with the deformation structure (shear bands) (Fig. 7).

As it was assumed, the one way effect can be observed for austenitic as well as for martensitic specimens (Fig. 8,9). Reversibility of about 100 % after heating the specimens (i.e. nearly complete one way effect) can be found maximal in total strains up to $\epsilon = 0.5$ %. Raising the strain value leads to a decrease of reversibility caused by plastic deformation of the austenite. This agrees to the results of the microhardness testings and could be observed in other Fe–base shape memory alloys, too (7).

The absence of any martensite plateau that occurs in stress–strain–diagrams of other shape memory alloys like NiTi or CuZnAl (8), is caused by the described plastic deformation of the austenite. The curve shows quite a normal course after leaving the elastic straight line.

Fig. 5: Course of the tensile strength and yield strength of alloy A depending on the testing temperature

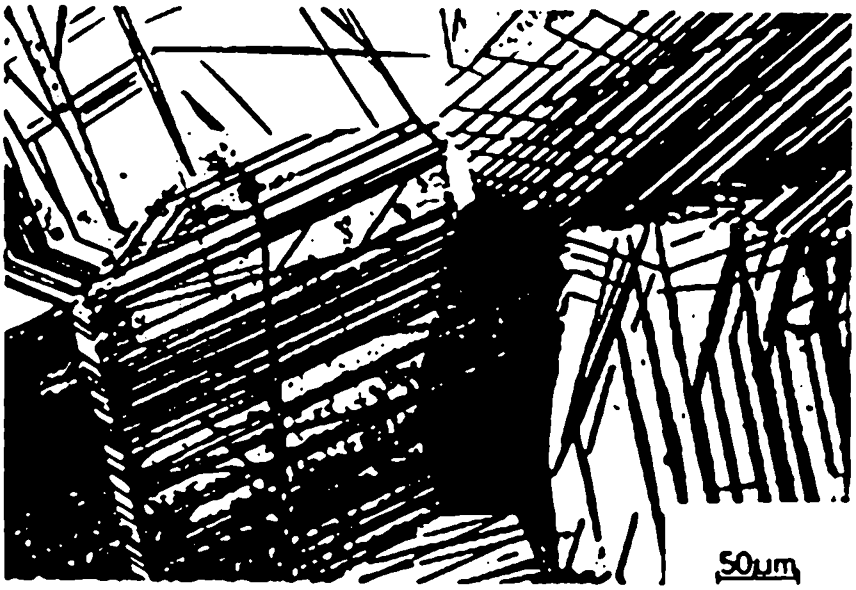
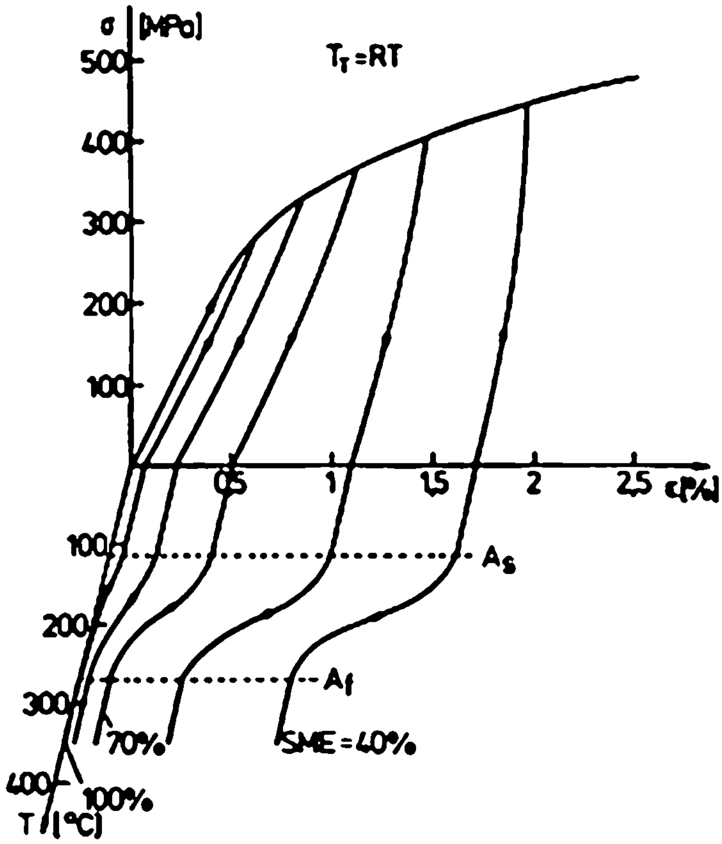
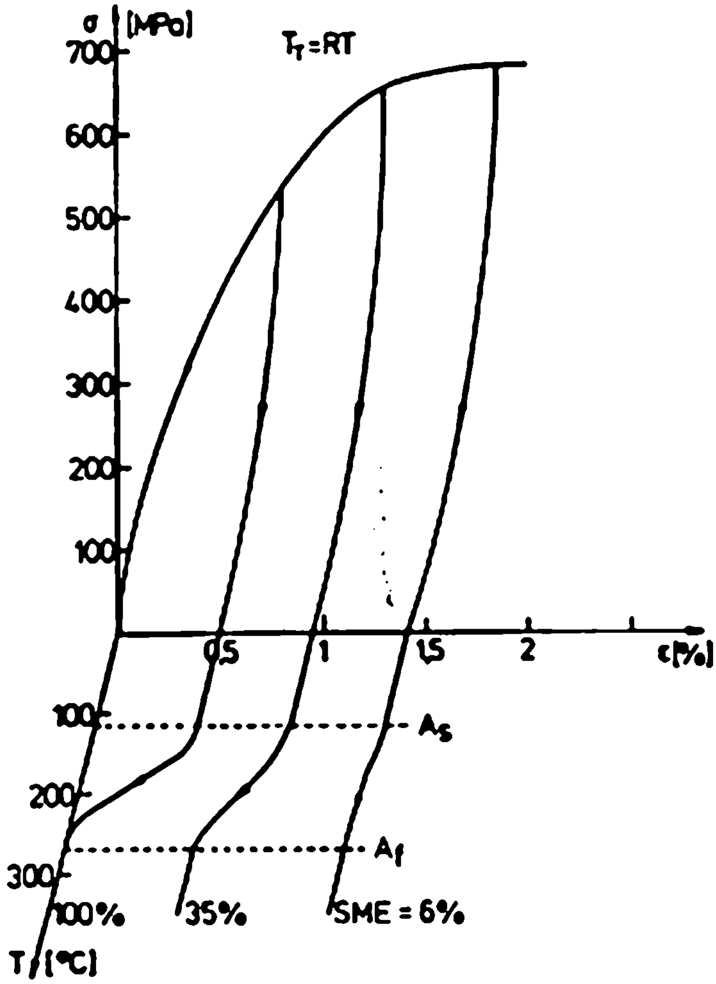


Fig. 6: Stress induced martensite, alloy D (30 min./600 °C)

Fig. 7: Stress induced martensite, alloy B with 30 %



<u>Fig. 8:</u> One way effect of the ausaged alloy D in austenitic state at different strain rates (30 min./600 °C, T_T = RT)



<u>Fig. 9:</u> One way effect of the ausaged alloy D in martensitic state at different strain rates (30 min./600 °C, – 196 °C, T_T = RT)

Conclusions

The following conclusions can be drawn from the present investigation:

1. Formation of martensite in Fe–Ni–Co–Ti–alloys leads to a small amount of plastic deformation of the residual austenite.
2. Ausaged specimens show brittle intercrystalline failure because of inhomogeneous η–phase precipitation at the grain boundaries. This brittleness can be prevented by hot rolling between homogenization and ausageing treatment.
3. The morphology of stress induced martensite is thin plate like and depends on the applied stress and the crystal orientation.
4. Fe–Ni–Co–Ti–alloys exhibit the one way effect in the austenitic as well as in the martensitic state.

Acknowledgement

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