

Deformation and martensitic transformation

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Abstract. The influence of applied stresses and imposed plastic deformation on the martensitic transformation of a parent phase is described. Changes in mechanical properties such as flow stress, work hardening rate, fracture toughness, etc brought about by strain-induced martensitic transformation are briefly examined. In the absence of appreciable dislocation glide, atomic displacements associated with glissile boundaries are highly ordered and reversible modes of (plastic or nonlinear pseudoelastic) deformation. Such processes lead to large strains and are encountered in deformation twinning, martensitic transformations and in the reorientation of martensite units. The reversibility leads to phenomena such as elastic twinning, thermoelastic martensites, superelasticity, shape memory and two-way shape memory effects, and rubber-like behaviour. These are discussed using a unified approach based on thermoelastic equilibrium. The shape memory effect suggests several potential applications of the martensitic transformations in non-ferrous alloys in which the effect is most commonly observed. Recent developments in this area are reviewed with special reference to the prerequisites for the effect and the influence of metallurgical processing on the extent of shape recovery.

Keywords. Plastic deformation; pseudoelasticity; shape memory effect; martensitic transformation.

1. Introduction

A martensitic transformation is a first-order solid-state structural transformation that is diffusionless, involves relative atomic motion by amounts smaller than the interatomic spacing, and exhibits a lattice correspondence between the parent and product structures. The transformation is generally adiabatic, athermal and exhibits hysteresis. It proceeds by a shear mechanism, and the net macroscopic distortion of the crystal can be regarded as an invariant plane strain. This implies that the interface between the parent phase and the growing martensite crystal is a plane of zero average macroscopic distortion. This feature has been verified in several experiments on the surface relief effect produced by the transformation. According to the phenomenological theories proposed by Wechsler *et al* (1953) and by Bowles and Mackenzie (1954) it is possible to consider the invariant plane strain as a superposition of the following components: a lattice strain (Bain strain), a lattice-invariant shear, a rotation and a dilation. The lattice-invariant shear required can manifest itself as slip, twinning or faulting of the martensite units. Besides, the different variants of martensite units formed from a parent phase crystal are generally arranged as three dimensional self-accommodating groups such that the total macroscopic strains are minimised.

Martensitic transformations can be induced by the application of external stress or by lowering the temperature. The former possibility arises because the free enthalpies of the parent phase and the martensite, and their equilibria, depend not only on the temperature and the composition but also on the stress; moreover, the nucleation and

growth processes are associated with shear strains and these couple to the internally or externally applied stresses. This paper deals with the relationship between martensitic transformations and deformation (both recoverable and nonrecoverable). We first consider the thermodynamic and kinetic aspects of stress-assisted and strain-induced martensitic transformations. This is followed by an account of an experimental study of the influence of plastic deformation at cryogenic temperatures on the martensitic transformation in an austenitic stainless steel. Next, we deal with the phenomenon of transformation-induced plasticity and its applications. Finally, a series of related phenomena including thermoelasticity, pseudoelasticity and the shape memory effect are discussed from a unified point of view based on the reversibility of the deformation caused by the migration of glissile interfaces.

2. Stress-assisted and strain-induced martensitic transformations

The driving force for the spontaneous martensitic transformation is the difference ΔG^c in the chemical free energy between the parent and the martensite phases. It is known that the transformation does not set in at the equilibrium temperature T_0 ($\Delta G^c(T = T_0) = 0$). Instead, it requires substantial undercooling, and commences at a temperature $M_s < T_0$. If the parent phase is stressed at a temperature T ($T_0 > T > M_s$), a mechanical driving force, U , is added to the chemical driving force; stress-assisted martensitic transformation occurs at T if

$$|\Delta G^c(T)| + U \geq |\Delta G^c(M_c)|. \quad (1)$$

U is a function of the stress and the orientation of the martensite crystal with respect to the parent phase. Patel and Cohen (1953) have shown that

$$U = \tau \gamma_0 + \sigma \varepsilon_0, \quad (2)$$

where τ is the shear stress resolved along the transformation shear direction on the habit plane, γ_0 is the transformation shear strain, σ is the dilatational stress resolved normal to the habit plane, and ε_0 is the normal component of the transformation strain. If a single crystal of the parent phase is stressed according to the geometry shown in

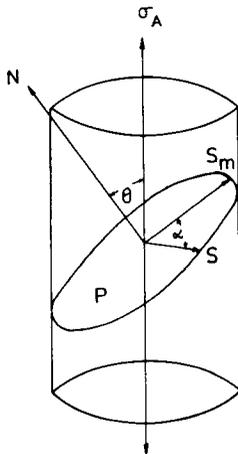


Figure 1. Geometry of stressing in a single crystal. P is the habit plane having N as its normal. S is the shear direction and S_m is the maximum shear direction of the applied stress on the habit plane. (The other symbols are explained in the text).

figure 1, τ and σ can be expressed for any given orientation of the martensite plate as

$$\left. \begin{aligned} \tau &= 0.5 \sigma_A \sin 2\theta \cos \alpha \\ \sigma &= \pm 0.5 \sigma_A (1 + \cos 2\theta) \end{aligned} \right\} \quad (3)$$

where σ_A is the absolute value of the applied stress, θ is the angle between the axis of the applied stress and the normal to the habit plane, and α is the angle between the shear direction of the transformation and the maximum shear direction of the applied stress on the habit plane (the sign of σ specifies whether the stress is tensile or compressive). If the chemical driving force (ΔG^c) decreases linearly with an increase in temperature above M_s , it is logical to expect that the critical applied stress for martensite formation should increase linearly with temperature. In practice, it is observed that the critical stress shows a linear dependence only between M_s and M_s^σ (figure 2). Beyond M_s^σ the stress required would exceed the flow stress of the parent phase and thus cause plastic deformation of this phase. Olson and Cohen (1972) have postulated that in the temperature range $M_s^\sigma < T < M_d$ the nucleation of martensite is aided by the large number of defects caused by the plastic deformation, and hence the transformation is considered to be strain-induced (M_d is the highest temperature at which the

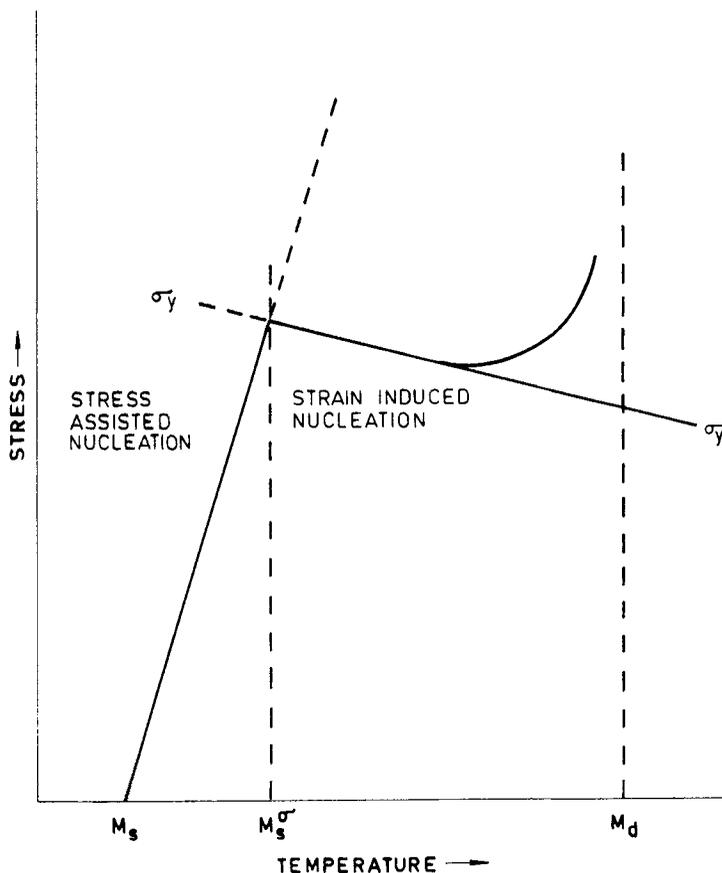


Figure 2. Variation of the critical stress required for the onset of the transformation at different temperatures. (The regimes of stress-assisted and strain induced transformation are indicated.)

transformation can be induced; $M_d < T_0$). There is thus a fundamental difference between stress-assisted and strain induced transformations.

The volume fraction f of the martensite formed at any given level of plastic strain ϵ_p is given by (Olson and Cohen 1975; Murr *et al* 1982)

$$f = 1 - \exp \{ -\beta [1 - \exp(-\alpha \epsilon_p)]^n \} \quad (4)$$

where α , β and n are temperature dependent parameters. Recently, Tamura (1982) has suggested that the transformation kinetics should be based on the stress rather than the strain, and has proposed a different expression for f that depends on σ_A , γ_0 and ϵ_0 . The role of plastic deformation in the strain-induced nucleation of martensite has been a subject of controversy. Olson and Cohen (1972, 1976) firmly believe that the intersection of slip bands produced as a result of plastic deformation acts as a nucleation site for martensite. In contrast, Suzuki *et al* (1977) and Onodera and Tamura (1979) have expressed the view that the local stress concentration near the grain boundaries due to the pile-up of dislocations would raise the applied stress to values obtained by the extrapolation of the straight line in figure 2 well beyond M_s^* .

3. Strain-induced martensitic transformation in a type 316 austenitic stainless steel

Type 316 austenitic stainless steel does not transform to martensite spontaneously even when it is cooled down to 77 K. In contrast, plastic deformation at low temperatures leads to the formation of two different types of martensites: ϵ martensite with an hcp structure and α martensite with a bcc structure. Figure 3 illustrates the variations in the volume fractions of these phases with plastic strain at 77 K. The main conclusions that can be drawn from this plot are: (i) With increasing plastic strain, the amount of ϵ phase present increases to a maximum and then decreases gradually, (ii) Detectable amounts of the α phase are formed only beyond a certain minimum value of the plastic strain, beyond which its volume fraction increases continuously with the plastic strain. These results imply that the α martensite forms at the expense of the ϵ martensite.

Plastic deformation of the austenite (γ) at temperatures below 220 K leads to the progressive formation of wide stacking faults, clusters of overlapping stacking faults, and finally, thin sheets the ϵ martensite. Transmission electron microscopy of the deformed samples reveals a large number of shear bands on $\{111\}_\gamma$ planes. These bands

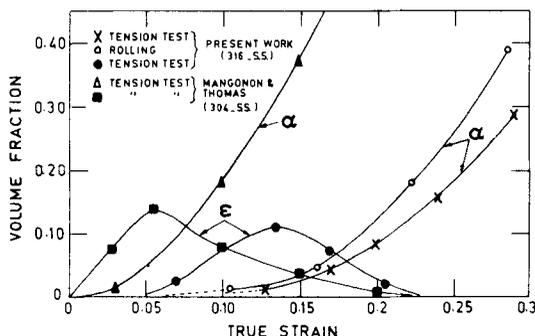


Figure 3. Variation of the volume fraction of the ϵ and α martensites in a type 316 stainless steel as a function of the true strain at 77 K. The results of Mangonon and Thomas (1970) in a type 304 stainless steel are shown for comparison.

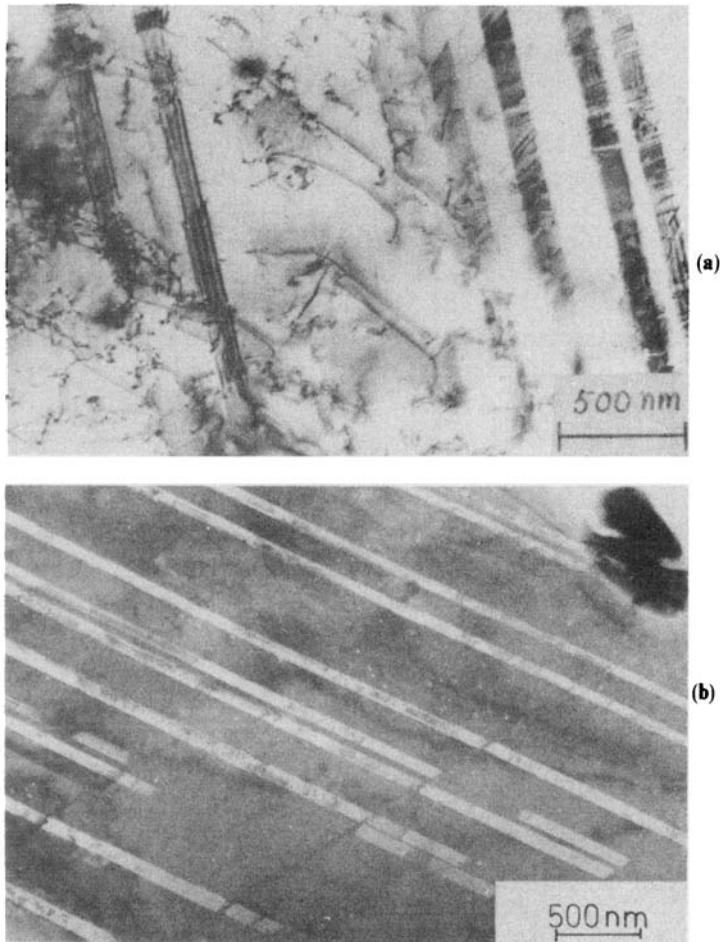


Figure 4. Transmission electron micrographs of type 316 stainless steel deformed at 77 K up to 5% strain. (a) wide stacking faults forming on the $\{111\}_\gamma$ planes, (b) dark field micrograph showing one of the variants of the deformation bands.

consist of overlapping faults, deformation twins and the ϵ martensite, the relative proportions of which vary with the temperature, strain rate and the extent of deformation. Typical examples of the microstructure obtained after tensile deformation at 77 K up to 5% strain are shown in figure 4. When the austenite is subjected to strains larger than 15% at 77 K, tiny crystals of α are observed at the intersections of two different variants of the shear bands. On continuing the deformation, the α nuclei grow along the cylindrical axes of the rod-shaped volumes comprising the intersections of these bands. The above experimental evidence obtained by x-ray diffraction and electron microscopy techniques proves conclusively that the ϵ martensite forms as an intermediate phase in the $\gamma \rightarrow \alpha$ transformation, the sequence being representable as $\gamma \rightarrow \epsilon \rightarrow \alpha$ (Seetharaman 1976; Seetharaman and Krishnan 1981).

4. Transformation-induced plasticity (TRIP)

As already mentioned, martensitic transformation can be considered as a mode of plastic deformation of the parent phase that competes with other modes such as slip and twinning. Therefore, if martensite forms dynamically during the mechanical testing of any alloy, the mechanical properties of the alloy would inevitably be altered considerably. Figure 5 shows the stress-strain curves of the Fe-29 Ni-0.26 C austenitic alloy deformed in tension at different temperatures (Tamura *et al* 1970a). The values of M_s and M_d for this alloy are 213 K and 298 K, respectively. The large serrations observed in the curves at 223 K and 203 K are due to the formation of stress-assisted martensite. On the other hand, the curves corresponding to 243 K or 263 K exhibit very fine serrations and also substantial tensile elongation: the samples corresponding to these curves show evidence of the presence of strain-induced martensite. Such an enhancement in the ductility of an alloy owing to the onset of strain-induced martensitic transformation is known as 'transformation-induced plasticity (TRIP)'.

The tensile properties of three different austenitic steels in the temperature range $M_s^a < T < M_d$ are shown in figure 6 (Tamura *et al* 1970b). The inverse temperature dependence of the 0.2% proof stress observed between M_s^a and M_s (figure 6a) is due to the transformation strain caused by the stress-assisted martensite formed before the yielding of the parent austenite phase. The total elongation exhibits a maximum at temperatures just above M_s^a . Such an enhancement in elongation has been attributed to the suppression of the necking phenomenon. This, in turn, is due to the increase in the work-hardening rate caused by the strain-induced martensitic transformation. Similar increases in the work-hardening rates have also been encountered in austenitic stainless steels (Angel 1954; Ludwigson and Berger 1969; Seetharaman 1976). The hardness of the martensite increases with increase in the carbon content of the steel. It follows that, for a given volume fraction of martensite, the increase in the work-hardening rates will be high for high carbon steels. Thus the TRIP phenomenon will be dominant in austenitic steels containing high levels of carbon.

Although ductility can be enhanced markedly by TRIP, strengthening must be achieved by other methods. Zackay *et al* (1967) developed an ultra-high-strength steel called TRIP steel in which the ductility is enhanced by the TRIP of the retained austenite

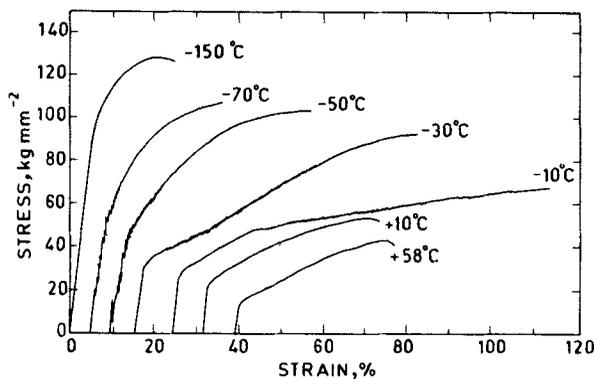


Figure 5. Stress strain curves of Fe-29Ni-0.26C austenitic alloy deformed in tension at different temperatures.

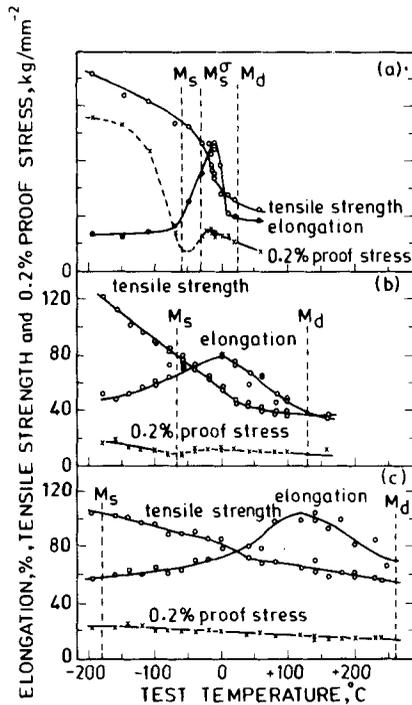


Figure 6. Effect of test temperature on tensile properties in three different types of metastable austenitic steels (a) Fe-29Ni-0.26C, (b) Fe-19-Cr-11Ni, (c) Fe-24Mn-0.26C.

while the strength is increased by the ausformed martensite. TRIP steels exhibit good properties at low temperatures because of the retained austenite, and are therefore suitable structural materials for chemical plants, containers for liquified gases, armour plates and earthquake dampers. Furthermore, the TRIP phenomenon is very effective in improving the formability, particularly the deep drawability, of austenitic stainless steels (Divers 1964).

5. Thermoelasticity, pseudoelasticity and shape memory effect

The thermodynamics of the growth of the martensite plate can be represented by the following energy balance:

$$G_T^c + U \geq \tau_0 \gamma_0 + \sigma_0 \epsilon_0 + S_M A + \zeta \gamma_0, \quad (5)$$

where τ_0 is the intrinsic shear resistance of the parent phase against the shear strain γ_0 , σ_0 is the intrinsic tensile resistance of the parent phase against ϵ_0 , S_M is the specific surface energy of the parent-martensite interface, A is the shape factor of the martensite, and ζ is a dissipation factor. Let us exclude the trivial case of the stoppage of a growing plate by a grain boundary or some equivalent obstacle. Then an equality in (5) implies that any small change in the thermal or mechanical components of the driving force will cause growth or shrinkage of the martensite plate. This equilibrium condition forms the basis of the thermoelastic martensitic transformation. A thermoelastic martensite forms and grows continuously as the temperature is lowered, and shrinks and vanishes as the temperature is raised. Figure 7(a) gives a schematic representation of this transformation indicating the increase in the internal stress σ_i and the volume fraction f of the martensite, these being measures of the progress of the transformation.

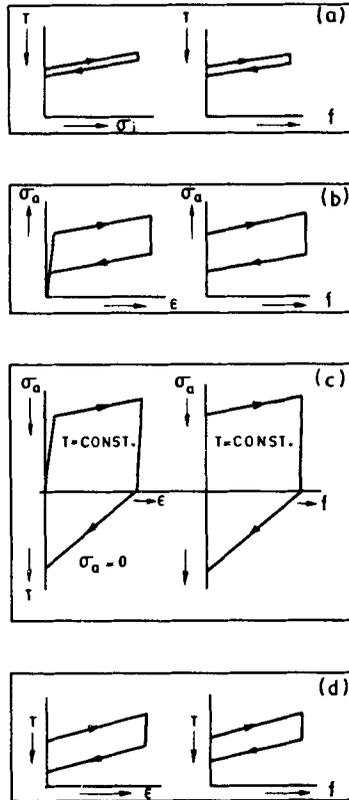


Figure 7. Schematic representation of (a) thermoelasticity (b) pseudoelasticity (c) shape memory effect and (d) two-way shape memory effect.

There is no sudden appearance or disappearance of large groups of martensite plates (the burst phenomenon).

Pseudoelastic behaviour (figure 7b) is a complete mechanical analogue of the thermoelastic transformation. In this case, the transformation proceeds continuously with increased applied stress. The 'plastic' strain is caused by the shape strain accompanying the formation of martensite. Recovery of the shape strain therefore occurs when the transformation is reversed. This type of pseudoelastic behaviour is obtained when the material is stressed at temperatures higher than the finish temperature A_f for the reverse transformation. Pseudoelasticity can also occur by the reorientation of the existing martensite variants (Krishnan *et al* 1974). Other terms such as superelasticity, ferroelasticity, rubber-like behaviour etc. have sometimes been used to describe pseudoelastic behaviour.

The shape memory effect arises if a macroscopic deformation is accompanied, as before, by a martensitic transformation which is not reversed on removal of the applied stress; in a second step the reverse transformation and a concomitant reversal of the macroscopic deformation are induced by heating. Figure 7(c) shows this behaviour schematically. The upper half represents the response of the specimen to the isothermal variation of the applied stress, while the lower half pertains to the effect of subsequent heating. The idealised curves shown in figure 7(a-c) suggest close interrelations between these processes. If the stress hysteresis in a pseudoelastic transformation is so large that the reverse transformation is incomplete at $\sigma_a = 0$, then the residual martensite can be

reverted by heating, *i.e.*, by employing the shape memory effect. This clearly demonstrates the interchangeability of stress and temperature as state variables.

A two-way shape memory effect occurs when the specimen deforms spontaneously during the cooling half-cycle and recovers during the heating half-cycle (figure 7d). An alloy which exhibits the one-way or normal shape memory effect can often be converted into a reversible memory alloy by "training" with a few cycles of the one-way type, or (more effectively), a few pseudoelastic cycles above M_s (Christian 1982). The maximum strains of the reversible effect are normally much smaller than those of the simple shape memory effect for a given alloy.

6. Common characteristics of materials exhibiting the memory effect

On the basis of extensive investigations conducted on Au-Cd, In-Tl, Ni-Ti, Cu-Zn, Fe-Pt and Cu-Al-Ni alloys, Wayman and Shimizu (1972) have concluded that the following conditions are necessary for an alloy to exhibit the shape memory effect: (i) The martensitic transformation should be a thermoelastic one; (ii) the parent phase and the martensite should be ordered; (iii) the lattice-invariant shear must occur by twinning rather than by slip.

While the first two conditions hold good for all the alloys investigated so far, the last condition, namely, the internal twinning of martensite does not appear to be very general. In fact, there is ample evidence in the literature to show that martensites in silver- and copper-based ordered alloys contain extensive and periodic faulting in the stacking sequence of the close packed planes, and these alloys do exhibit the shape memory effect to a significant extent.

6.1 Thermoelastic martensites

A thermoelastic martensite is generally characterised by (i) a small driving force, (ii) a small shear component of the shape strain, (iii) a small volume change, and (iv) a matrix with a high elastic limit (Dunne and Wayman 1973; Tong and Wayman 1975). It is clear that a thermoelastic martensite crystal must have a perfectly coherent interface with the surrounding matrix and that the coherence must be maintained during the whole process of growth or shrinkage. This is possible only if the volume change associated with the transformation is small. Furthermore, the small value of the thermal hysteresis is a direct consequence of the smallness of the driving force.

6.2 Effect of ordering

It is found that some degree of ordering (usually, long range ordering) is present in all parent phases which are known to transform thermoelastically. As a case in point, the ordered Fe₃Pt alloy undergoes a thermoelastic martensitic transformation and also exhibits the shape memory effect, while disordered specimens of the same composition display neither.

According to the phenomenological theory of martensitic transformation (Wayman 1964), the habit plane is a plane of zero average distortion. However, localised distortions still persist at the interface though they are macroscopically arranged out by the lattice-invariant shear. This fine-scale distortion is not expected to exceed the elastic limit of the matrix in ordered alloys (because of the order-hardening effects), but in

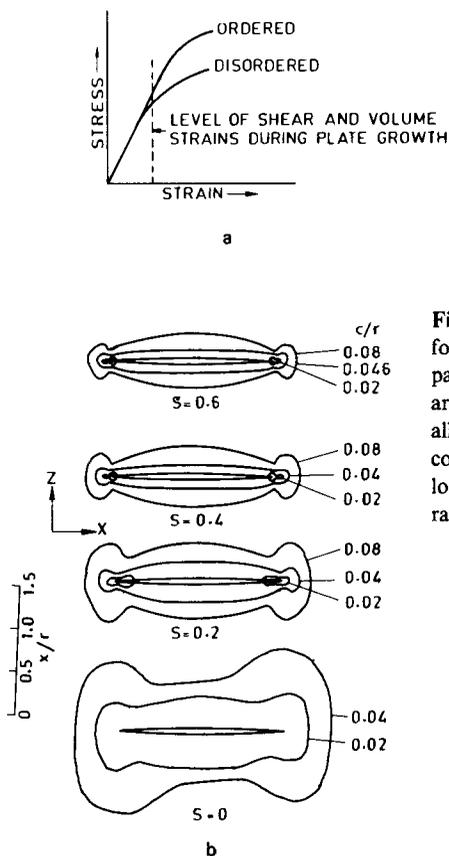


Figure 8. (a) Schematic stress-strain curves for the disordered and ordered states of the parent phase. (b) Matrix yield stress contours around an oblate spheroidal plate of an Fe-Pt alloy. The plate is in the x-y plane and the contours are drawn for varying values of the long range order parameter, S and the aspect ratio, c/r .

disordered alloys the elastic limit is evidently exceeded and the interface eventually becomes incoherent. This is explained with the help of the hypothetical stress-strain curves shown in figure 8(a). Ling and Owen (1979) have re-examined the plastic deformation around a growing thermoelastic plate, and have predicted the size of the plastic zone around plates of various shapes and in materials with varying long-range order (figure 8b). In the disordered alloy (longrange order parameter $S = 0$), general yielding has occurred all around the isolated plate with an aspect ratio $c/r = 0.02$. However, when $S > 0.4$, such yielding is confined to a small volume near the radial edge. It is therefore clear that the primary role of ordering is to strengthen the parent and the martensite phases without affecting significantly the value of the shear modulus, μ , thereby increasing the ratio σ_y/μ of the yield strength to μ .

The ordered arrangement of atoms is also important when one considers the reverse transformation upon heating. If a material is to exhibit the shape memory effect, it is essential that all the martensite plates formed in a single crystal of the parent phase should revert as a unit to the original orientation of the parent phase. When an ordered parent phase transforms to an ordered martensite, the arrangement of atoms, though ordered, exhibits a lower symmetry in the latter phase (following the deformation of the lattice). Considering that another lattice distortion must be operative during the inverse transformation, it follows that the number of reverse lattice correspondences is more limited than in cooling transformation, one imposes the restriction that the original ordered arrangement in the parent phase be recovered. In fact, in Cu-Al-Ni alloys it has been shown that there is only one lattice correspondence between the parent and the

martensite phases which maintains the original order (Wayman and Shimizu 1972). Accordingly, the ordered martensite is obliged to revert to the original orientation of the parent phase in this case.

6.3 Lattice softening

Shape memory alloys exhibit interesting and anomalous behaviour with respect to the temperature dependence (near M_s) of the electrical resistivity, elastic constants, etc. Moreover, 'extra' reflections appear in the x-ray and electron diffraction patterns taken at temperatures just above M_s . The damping capacity of these alloys increases anomalously to reach a maximum near M_s . Among all these pre-martensitic phenomena, the most important and striking feature is that the temperature derivative of the elastic shear constant dC'/dT (where $C' = (C_{11} - C_{12})/2$) is large and positive. Zener (1947) has explained the low values of $C_{11} - C_{12}$ observed in β -brass for $T \approx M_s$ on the basis of the instability of the bcc lattice with respect to a $(110)[\bar{1}\bar{1}0]$ shear. Thus it is suggested that the very low value of the thermodynamic driving force required at M_s is due to the assistance by long wavelength phonon instabilities of the nucleation of martensite in a soft lattice. As with ordering, it is probable that the major, and certainly the most direct consequence, of lattice softening in the nature of martensite growth would be with regard to the ratio σ_s/μ . A decrease in μ increases the magnitude of the shear strain which can be accommodated elastically.

6.4 Mechanisms for the shape memory and related effects

The most remarkable consequence of deformation by interface migration is that the substantial strains produced this way are often completely recoverable when the chemical or mechanical driving force is removed. This 'shape recovery' is a consequence of the highly ordered nature of the atomic displacements during thermoelastic interface deformation. It contrasts with the irreversibility of the plastic deformation produced by dislocation glide or by the formation of martensite plates or deformation twins with an accompanying dislocation deformation. The most important mechanisms based on interface migration which have been put-forward for the effects under consideration include (i) the stress-assisted reversible growth of martensite; (ii) the reorientation of the martensite plates under an applied stress; (iii) inter-twin growth; and (iv) martensite-to-martensite transformation under stress. We now discuss these mechanisms in brief.

6.4a Stress-induced reversible growth of martensite: The essential feature of this mechanism is that those plates of martensite grow which have orientation and shear directions favoured by the dominant shear due to the imposed macroscopic change in shape; while other less favourably oriented variants shrink. The growth must be mechanically reversible (and hence thermoelastic) if the shape is to be completely recoverable. The martensite plates which participate in this mechanism may be nucleated under the action of the applied stress, in which case only the favourably-oriented variants will be produced; or they may be plates which may have been formed athermally before the external stress is applied.

6.4b Reorientation of martensite plates: Macroscopic shape deformation can be achieved by an externally applied stress through the growth of martensite plates

favourably oriented in the stress field at the expense of less suitably oriented plates. It is to be noted that when the martensite interface moves, the habit plane itself remains unchanged, suggesting that ratio of the thickness of alternate segments of transformation twins within the plates remains unchanged. A special and specific mechanism whereby martensite plates can be reoriented under stress has been proposed by Wasilewski (1975). It is described as a 'stress-assisted double transformation'. When a stress is applied at a temperature between A_d and M_f , the martensite transforms to the β' -phase by the stress-assisted reverse transformation; (M_f is the martensite finish temperature during cooling and A_d is the lowest temperature at which martensite reverts to austenite under deformation) and then, because it is both thermally unstable and subject to an external stress, it transforms instantaneously to a martensite variant more favourably oriented with respect to the applied stress field than the original plate. However, there is no experimental evidence as yet for the transitory existence of an intermediate β' -phase.

If the specimen is fully martensitic and consists of self-accommodating groups of martensite, then on the application of external stress, movements of the existing plate boundaries or creation of new ones together with changes in internal structure will occur. Because no untransformed parent phase is present, no invariant plane strain conditions need to be satisfied; only a three-dimensional strain minimisation is necessary.

6.4c Inter-twin growth: Macroscopic deformation by change in the thickness of two twin variants under an applied stress was first reported by Olander (1932) and studied in detail in Au-Cd and In-Tl alloys by Chang and Read (1951). This mechanism does not involve any movement of the martensite-martensite interfaces. On applying a stress the twin boundaries move so as to thicken those twins which are favourably oriented with respect to the applied stress. On removal of the stress, the twin boundaries return (relatively slowly) to their original positions, and the specimen recovers its original shape. The strains thus recoverable may be as large as 0.08 in Au-Cd.

6.4d Martensite-to-martensite transformation under stress: In Cu-Ni-Al alloys two different martensites β'_1 and γ'_1 form from the parent β_1 . It is reported that the β' martensite can be transformed partially to γ'_1 by applying a stress below M_s . Similarly, athermally formed β'_1 martensite in Cu-Al alloys can transform either to an fcc or hcp structure by deformation. Delaey and Warlimont (1975) have demonstrated a similar martensite-to-martensite-transformation in a Cu-Zn-Al alloy.

The different possible routes in a pseudoelastic cycle and a shape memory cycle are illustrated in figure 9 (Cook 1981). This diagram compares the changes involved in a shape memory cycle (VABCDEZ) with those in a pseudoelastic cycle (VWXYZ). If the same martensite phase is produced in the two cycles, the structure at X is identical with that at C and D, and the state X may be converted into D by cooling at a constant strain.

7. Applications of the shape memory effect

One of the earliest applications of the shape effect was the introduction of tubing or pipe couplings that shrink during heating. Ni-Ti alloy couplings are used for connecting aircraft hydraulic lines. These are expanded by 4% in the martensitic

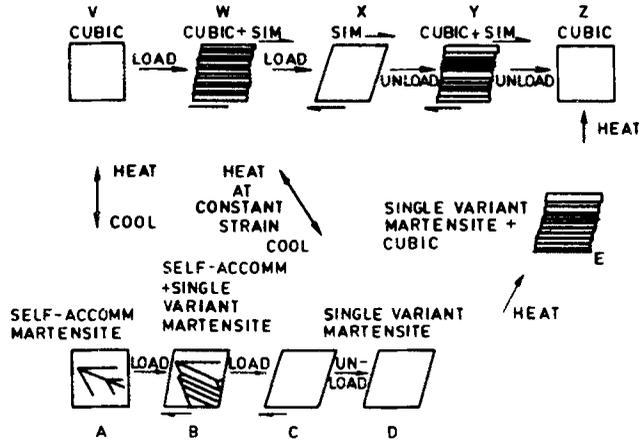


Figure 9. Schematic structural changes during a pseudoelastic cycle VWXYZ and a shape memory cycle VABCDEZ.

condition at liquid nitrogen temperatures, and placed around the tubes to be joined. Warming at room temperature produces a tight seal. Substantial developments have taken place in the field of copper-based shape memory alloys, particularly the Cu-Zn-Al alloys, with emphasis on their applications as thermostats, automotive control devices and actuators. Applications in the field of medicine include the use of such materials for orthodontic dental arch wires, blood clot filters and orthopaedic devices.

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Discussion

V K Wadhawan: What is the role of the superlattice structure in the shape memory effect?

V Seetharaman: Ordering of the parent and martensite phases has two consequences. (i) It increases the yield strength of the parent and product phases by the 'Fisher's order hardening mechanism' (ii) It reduces the number of variants available for the reverse transformation so that strain recovery can occur efficiently.

S Ray: Does the shape memory effect occur only in isothermal martensite systems (through mode softening) or is it possible in athermal martensites also?

Seetharaman: It is seen in athermal martensites though the mode may not soften completely.

R Krishnan: Are there examples of transformation induced plasticity in ceramic materials?

Seetharaman: I know only a few examples of these. The material must have a residual plasticity before we can introduce TRIP.