Review

Iron-based shape memory alloys for civil engineering structures: An overview

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HIGHLIGHTS

• The development of iron-based SMAs is presented, focusing on features for civil engineering.
• Differences between the martensitic transformation in Ni–Ti and Fe–Mn–Si SMAs are highlighted.
• High recovery stresses, which are necessary for prestressing, can be obtained for FeMnSi alloys.
• Pilot experiences on the application of FeMnSi alloys are presented.
• This paper collects unsolved aspects for future research.

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ABSTRACT

Iron-based shape memory alloys (SMAs), especially Fe–Mn–Si alloys, are materials that have great potential in civil engineering structures, but their application is still in a pioneer stage. Recent developments in alloy composition and manufacturing envisage new perspectives, especially in the field of repairing structures as well for new structures, when using these SMAs as prestressing tendons. This paper presents the fundamentals of the martensitic transformation from an engineering perspective as well as some key properties, such as recovery stress, corrosion resistance, weldability and workability. Finally, some unsolved aspects are collected, and new perspectives for the use of these SMAs are presented.

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Contents

1. Introduction ...................................................................................................... 282
2. The martensitic transformation ........................................................................ 283
3. Material properties. .......................................................................................... 287
   3.1. Recovery stresses. ...................................................................................... 288
   3.2. Corrosion resistance ................................................................................ 288
   3.3. Weldability ............................................................................................... 288
   3.4. Production ............................................................................................... 289
   3.5. Workability at room temperature .............................................................. 289
4. Pilot experiences on the application of Fe–Mn–Si alloys. .............................. 289
   4.1. Shape memory effect .............................................................................. 289
   4.2. Damping ................................................................................................. 290

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1. Introduction

Shape memory alloys (SMAs) are unique materials that have the ability to achieve great deformations and to return to a predefined shape after unloading or upon heating [1]. This shape memory effect is the result of the reversible phase transformation that SMAs undergo, the so-called martensitic transformation. The martensitic transformation can be produced by changes in temperature or by the action of stresses. In the former case, the martensitic transformation takes place at defined temperatures (Fig. 1). The martensitic transformation, or forward transformation, is induced upon cooling the austenite phase (stable at high temperatures), and consists of the appearance of the martensite phase (stable at low temperatures). In the absence of applied stresses, the temperature at which the process begins is known as \( M_s \) (martensite start), whereas \( M_f \) (martensite finish) is the temperature at which the transformation finishes (Fig. 1). If the material is in martensite \((T < M_s)\), then the reverse transformation can be induced by heating the material. The formation of austenite will start at temperature \( A_s \) (austenite start) and will finish at temperature \( A_f \) (austenite finish). The transformation shows thermal hysteresis, in other words, the forward and reverse transformations do not take place at the same temperature [2].

The first discovery of a material with shape memory was documented by Chang and Read, who observed a reversible phase transformation in an Au–Cd alloy [1]. Buehler et al. discovered in 1962 the shape memory effect in a Ni–Ti alloy [3], a fact that led to the boom in international research in this field and the appearance of the first real applications of these alloys. Since then, different types of alloys with the shape memory effect have been discovered. Ni–Ti alloys, in some cases having a third component, are the ones that, to date, hold the first position in the industrial market. The main drawback of these alloys for their application in civil engineering structures is their cost. In 1982, the shape memory effect (SME) in an Fe–Mn–Si alloy was discovered [4], and since then, new iron-based SMAs with improved SME properties have been developed. It is assumed that this progress will contribute to lowering the price of these materials and to making them much more competitive for civil engineering applications.

Janke et al. [5] presented possible applications of SMAs in civil engineering structures: passive vibration damping and energy dissipation, active vibration control, actuator applications and the utilization of the SME for tensioning applications or sensors. However, these authors stated that due to the size of the civil engineering structures and the actions of relatively high forces, low cost SMAs were needed.

The repairing of structures using the SME for prestressing tendons is a promising field for the use of iron-based SMAs, as well as their use in new structures. If iron-based SMAs can be used for prestressing applications, they have several advantages compared to the traditional prestressing/posttensioning technologies, for example, there are no friction losses, no anchor heads or ducts are required, and no space is necessary for applying the force with a hydraulic device. The reason is that the prestressing of the SMA tendons is not performed mechanically, as in conventional prestressing steel, but with heating, as will be explained later in this paper.

There are two different groups of iron-based SMAs [6]. The first group contains alloys such as Fe–Pt, Fe–Pd and Fe–Ni–Co, which exhibit the typical characteristics of thermoelastic martensitic transformations similar to Ni–Ti, with a narrow thermal hysteresis. However, in spite of extensive studies, no pseudoelasticity at room temperature has been reported with Fe–Pt or Fe–Pd alloys. In 2010, Tanaka et al. [7] presented an Fe–29Ni–18Co–5Al–8Ta–0.018B (mass %) SMA that shows a recovery strain of over 13% at room temperature and a very high tensile strength of 1200 MPa, placing it at the cutting edge of knowledge as far as new materials are concerned [8,9]. This iron-based SMA could be very useful for applications that are related to pseudoelasticity and damping capacity. Additionally, good superelasticity properties at room temperature have been found in an Fe–36Mn–8Al–8Ni (mass %) alloy [10], with a recovery strain of over 5% and a fracture tensile strain of over 8%. However, these two new alloys still need further development to be able to produce them in large amounts for real-scale elements in the construction industry, and the cost of the material would most likely be too high for construction standards because they should be cast in special conditions due to their composition.

The second group is a group of alloys such as Fe–Ni–C and Fe–Mn–Si, which have a larger thermal hysteresis in transformation but still exhibit the SME. The Fe–Mn–Si SMAs have received considerable attention over the past two decades due to their low cost, good workability, good machinability and good weldability [11], although the real applications are still limited except for some remarkable exceptions, i.e., large size joining pipes for tunnel construction and crane rail joint bars [6].

The SME of the Fe–Mn–Si alloys is attributed to the stress-induced martensite transformation from a parent \( \gamma \)-austenite (fcc – face-centered cubic) phase to an \( \varepsilon \)-martensite phase (hcp – hexagonal closed-packed) (Fig. 2) at low and intermediate temperature and the reverse transformation (\( \varepsilon \)- to \( \gamma \)-phase) at high temperature. In fact, it had long been known that Fe–Mn alloys could undergo this transformation, but the desired SME had not been obtained [12]. The problem was that for high amounts of Mn, \( \gamma \)-austenite was stabilized, making it difficult for the stress-induced martensitic transformation to occur. On the other hand, for lower amounts of Mn, when the alloy was subjected to stress, not only \( \varepsilon \)-martensite but also \( \alpha' \)-martensite (bct – body-centered tetragonal) was generated (Fig. 2c). This phase is irreversible. The occurrence of \( \alpha' \)-martensite induces dislocations markedly, preventing the SME from developing [12]. Sato and his co-workers discovered that the addition of Si allowed having the SME [4]. It was also observed that Cr, among other elements, was effective to a minor extent. This finding suggested that Cr was suitable as a
fourth element to be added to an Fe–Mn–Si alloy to improve its corrosion resistance [12]. If the amount of Cr exceeds 7%, the brittle σ phase intrudes in the alloy, impeding the SME. It was found that Ni was the most effective component in restraining the formation of σ phase and, therefore, Ni had to be added for increasing amounts of Cr [12]. The research at the materials level has been principally focused on anti-corrosion, training effects, cyclic deformation and strengthening [13].

The development of these Fe–Mn–Si alloys can be summarized by having two distinct stages. The most significant alloys of the first stage are summarized in Table 1. The main components that differentiate them are the amount of Mn and Cr. With these alloys, an excellent recovery strain has been achieved by means of cyclic thermo-mechanical treatments, the so-called “training” [14]. In general, the recovery strain due to the SME in these alloys without training is limited to only 1–2.5%, but the training improves the recovery strain, enabling very good shape recovery and enhancing the recovery strain by up to 4% [14]. Other more simple heat treatments combined with different rolling conditions have been recently reported [15–17]. The second stage of the development of Fe–Mn–Si alloys started when Kajiwara and his co-workers observed, in 2001, that the recovery strain and stress of iron-based SMAs, new perspectives for the use of these SMAs in the construction industry. The paper will highlight the differences between these alloys and other SMAs, i.e., Ni–Ti alloys, and it will collect unsolved aspects that can be a starting point for future research in this field. Moreover, based on the unique properties of iron-based SMAs, new perspectives for the use of these SMAs will be presented.

2. The martensitic transformation

The forward phase transformation in SMAs is a martensitic transformation. The martensitic transformation is, in general, a diffusionless solid state transformation in which the atoms move in an organized manner relative to their neighbors. This homogenous shearing of the parent phase creates a new crystal structure that does not have any compositional change (no diffusion). Although the variation in the relative position of the atoms is very small, the coordinated movement of all of the atoms leads to changes in the volume and can bring about significant macroscopic deformations. The martensitic transformation can be induced by a temperature change, e.g., during quenching and/or by mechanical deformation.

Fig. 3a shows schematically the atomic arrangement in a hypothetic lattice after mechanically induced martensitic transformation. The lattice is distorted without reorganization of the atoms (neighbors stay neighbors). This martensitic transformation is sometimes called thermoelastic and is typical for Ni–Ti-alloys, and it usually shows a small hysteresis. In a plastic deformation, on the other hand, the atoms are rearranged by slip, as shown in Fig. 3b. In this case, the neighbors are changed, but the lattice structure remains intact. This transformation can occur in a general alloy or in an SMA and cannot be reversed by a temperature change because the crystal structure is identical to the original state.

The martensitic transformation in Fe–Mn–Si-based alloys is also achieved by slip, but in contrast to plastic transformation, the
atoms are rearranged into a different lattice structure, namely from face-centered cubic (fcc) to hexagonal close packed (hcp) (see Fig. 2). This transformation cannot be visualized in a simple 2D picture and instead must be explained in a 3D setup. To this end, the fcc and hcp lattice structures are represented by close-packed spheres, whose centers correspond to the position of the atoms. The correspondence can easily be verified by looking at the cells shown in Fig. 4 on the right. Alternatively, the two lattice structures can be represented by three layers of close-packed spheres arranged in two different ways as shown in Fig. 4 on the left. In the first case, the third layer is arranged exactly above the first layer. This arrangement can be denoted as ABA and corresponds to the hexagonal close-packed lattice structure. In the second case, all of the three layers are in a different position, which is denoted by ABC. This arrangement corresponds to the face-centered cubic lattice when rotated in such a way that the center layer (B) coincides with a plane that intersects three diagonally opposite vertices in the fcc cell. The fcc structure can thus be transformed into the hcp structure by moving the top layer from the C-position to the A-position (Fig. 5).

The martensitic transformation from the γ-austenite (fcc structure) to the α′-martensite (hcp structure) by translating one atomic plane as described above and shown in Fig. 5 is called partial Shockley dislocation. Partial dislocations are movements of an atomic layer with less than one atomic distance, which therefore change the lattice structure. For an Fe–Mn–Si SMA, this martensitic transformation can be reversed by a temperature change. The slip that is involved in the transformation is responsible for the large hysteresis.

In an fcc structure, there are four possible slip planes. These can easily be visualized by rotating the fcc cube in Fig. 4 around the vertical axis in steps of 90°. Each of them contains three shear directions, which leads to twelve possible martensite variants in the Fe–Mn–Si alloys. If only one variant is generated in a single crystal, then the corresponding mode is called a ‘monopartial stacking’ [11]. An SMA alloy is normally a polycrystalline material consisting of several crystals or grains. In such a material, the difference in grain orientation as well as in the internal stress distribution will cause the conditions for monopartial stacking to not be fulfilled, and several variants will be activated. At these variant intersections, other phases can be formed [30].

Experimentally, the generation and propagation of the partial dislocations in the fcc matrix has been enhanced by lowering the stacking fault energy (SFE) [31,32]. It has been demonstrated that adding alloying elements such as Ti and Cu increase the SFE [33], whereas elements such as N, Ta, Ce or Sm lead to a decrease in the SFE [33,34].

The mechanically induced martensitic transformation occurs not only in shape memory alloys; it is also a well-known transformation in carbon steel, stainless steel and many other alloys. For example, high Mn steels (15–30% mass) are used in the automotive industry. Due to their high ductility with strains at failure up to 80–90% and quite high ultimate strengths, they offer very good conditions for the crash resistance of structural car body parts. The amount of Mn plays a significant role in the behavior of these steels [35]. Ultra high manganese steels exhibit the formation of mechanical twins, specifically, stress-induced α′- and α″-martensite in the γ-austenite matrix (Fig. 2) under internal or external stresses. These steels are commonly called TWIP (twin induced...
plasticity) steels. The TWIP effect is dominant when the manganese content is equal to or higher than 25%. TWIP steels (i.e., Fe–25Mn–3Si–3Al) exhibit a total elongation of approximately 92% and an ultimate tensile strength of 650 MPa at room temperature [35]. For steel alloys that have lower amounts of manganese content, approximately 15–20%, the alloy has a so-called TRIP (transformation induced plasticity) effect. TRIP Fe–Mn steels offer a slightly lower elongation than TWIP steels but a higher ultimate strength. For example, an Fe–20Mn–3Si–3Al steel exhibited a total elongation of approximately 82% and an ultimate tensile strength of approximately 830 MPa at room temperature [35]. The TRIP effect is due to the formation of $\alpha'$-martensite (Fig. 2c), which enhances the tensile elongation due to a retardation of the slipping. Ding et al. [36] showed that alloys that have an intermediate amount of Mn (23.8%) show simultaneously both effects. Therefore, the very high ultimate strains in the TRIP/TWIP steels are due to the martensite transformation, but without SME, since the $\alpha'$-martensite formation prevents the SME in the TRIP/TWIP steels.

Figs. 6 and 7 show some schematic phase diagrams (stress – temperature) and stress–strain curves for different thermomechanical paths, for a Ni–Ti alloy in Fig. 6a–c and for an Fe–Mn–Si alloy in Fig. 6d–f. As discussed above, SMAs have two different crystalline phases: austenite and martensite. Austenite is composed of a highly symmetric cubic phase, with the existence of only one single possible structure in this case, whereas martensite has a structure that has a lower symmetry, which allows, in the case of Ni–Ti for example, up to 24 different configurations or variants. The assembly of martensitic variants can exist in two forms in alloys that have a thermoelastic martensite transformation, i.e., Ni–Ti: twinned martensite, which is formed by a combination of self-accommodated martensitic variants, and detwinned martensite, in which a specific variant is dominant. When an SMA with a thermoelastic martensitic transformation is cooled below $M_f$ in the absence of an applied load, the crystal structure changes from austenite to martensite (forward transformation, see Fig. 1). The arrangement of variants occurs such that the average macroscopic shape change is negligible, which results in twinned martensite (path 1 in Figs. 6a and 7). If the twinned martensite is deformed due to external forces, its crystal structure changes to the variant, or variants, which enable(s) it to accommodate the maximum elongation and, as such, allow permanent deformations (detwinned martensite), as can be schematically seen in path 2 of Figs. 6a and 7. In this step, the applied stress must be sufficiently large ($\sigma_r$) to start the detwinning process. The total detwinnning of martensite is reached with the detwinning finish stress ($\sigma_f$). After unloading (path 3 in Figs. 6b and 7), the martensite remains detwinned. The change in the phase from detwinned martensite to austenite can be brought about by raising the temperature (path 4 in Figs. 6b and 7), and the SMA regains its cubic crystal structure, returning to its original shape if the deformations are unconstrained or generating recovery stresses otherwise, as will be seen in detail later in this paper.

For an Fe–Mn–Si alloy, it is also possible to create thermal martensite without an external apparent shape change (path 1 in Fig. 6d). During path 1, the full austenite–martensite transformation is not reached in an Fe–Mn–Si alloy, in contrast to alloys that show the typical thermoelastic martensitic transformation (Fig. 1). Moreover, the temperature-induced martensite variants cannot be easily changed by applying a stress (monopartial stacking), and it is not generally interesting for engineering applications to conduct path 2 from a temperature that is lower than $M_f$ or $M_s$ as shown in Fig. 6d. The reorientation of different variants can hardly be activated in Fe–Mn–Si alloys, due to the high level of the barrier energy between the martensite variants, and plasticity, or the TRIP effect, is activated before this type of energy level can be reached.

If the initial temperature of an Fe–Mn–Si alloy is higher than $M_s$, but not too close to $A_s$, then stress-induced martensite (path 2 in Fig. 6e) is recoverable. However, the full austenite–martensite transformation is still not reached for this stress-induced path. For an Fe–Mn–Si–Cr alloy, the percentage of stress-induced e-martensite has been reported to be 30% or less [12] and approximately 48% for an Fe–Mn–Si–Cr–Ni [38]. Plasticity (or irreversible slip) as

![Fig. 6. Schematic phase diagrams for Ni–Ti alloy (a–c) and Fe–Mn–Si alloy (d–f), showing: (a) the detwinnning of Ni–Ti with an applied stress; (b) the unloading and subsequent heating to austenite under no deformation constraint for Ni–Ti; (c) pseudoelastic loading path for Ni–Ti; (d) creation of thermal martensite in an Fe–Mn–Si alloy; (e) stress-induced martensite and subsequent heating to austenite under no deformation constraint for Fe–Mn–Si alloy; and (f) plastic deformation with an irreversible slip in the Fe–Mn–Si alloy.](image-url)
well as the formation of $\alpha'$-martensite can give rise to an incomplete shape memory. At high temperatures, irreversible slip is easier to activate and does not require a high mechanical loading level. Consequently, only plasticity occurs at high temperatures, while at low temperatures, phase transformations occur first, followed by plasticity [39]. At moderate temperatures, both mechanisms are coupled. It must be highlighted that the limits between the plasticity and the phase transformations related to the TRIP/TWIP effects are still not clearly reported in the scientific literature for these Fe–Mn–Si alloys. In a recently published paper [40], the authors reported a study on scanning electron microscopy and optical microscopy images of an Fe–Mn–Si alloy in undeformed samples and after a 4% tensile deformation at $-45 \degree C$ and 100 $\degree C$. When the strain was applied at $-45 \degree C$, the microstructure mostly consisted of the $\gamma$- and $\varepsilon$-phase. However, in the case of the sample that was deformed at 100 $\degree C$, a significantly lower amount of the $\varepsilon$-phase appeared, and on the other hand, a larger amount of the $\alpha'$-phase was found. Because the formation of the $\alpha'$-phase does not contribute to the SME in Fe–Mn–Si alloys, it is clear that the strain-induced transformation must be kept at temperatures that are not too high.

During unloading (path 3 in Fig. 6e), no reverse transformation ($\varepsilon$-martensite to $\gamma$-austenite) takes place in the region in which both phases are stable. This region extends over a broad temperature range due to the very large hysteresis of these Fe–Mn–Si alloys, although some pseudoelastic effect has been reported depending on the temperature, which indicates that the reverse transformation takes place partially during unloading [40]. Heating above $A_c$ (path 4 in Fig. 6e) will definitely activate the reverse transformation, which is the key point for the recovery strain and the generation of recovery stresses. When the temperature reaches the austenite finish temperature $A_f$, a part of the strain is recovered (not all due to trapped martensite plates retained by various defects).

In Fig. 6, the slopes of the lines, or boundaries, that define the transformation temperatures are known as “stress influence coefficients”. It is typically assumed that each pair of lines for the two transformations shares a characteristic slope [41], although this assumption is not necessarily the case for Fe–Mn–Si alloys. Moreover, in these alloys, the $M_t$ boundary for stress-induced martensite crosses the $x$-axis (zero-stress level) of the phase diagram for a temperature below the measured $M_t$ for thermally induced martensite. This property is an indication of the non-thermoelastic character of its stress-induced martensitic transformation [42]. It must be highlighted that the phase diagrams shown in Fig. 6 for an Fe–Mn–Si alloy will change considerably depending on the composition of the alloy and the thermomechanical treatments that are conducted. Specific phase diagrams for the Fe–Mn–Si alloys can be found elsewhere [39,40,42,43].

Fig. 8 schematically shows the generation of recovery stresses for three different alloys: Fig. 8a for a narrow hysteresis alloy (i.e. Ni–Ti), Fig. 8b a wide hysteresis alloy (i.e. Ni–Ti–Nb) and Fig. 8c for an Fe–Mn–Si alloy. In the two first cases, the prestraining follows path 2 from Fig. 6a, starting with the material being twinned martensite and evolving with an almost horizontal plateau, in which the detwinning process takes place (Fig. 8a and b).

For Ni–Ti alloys, the recovery stress increases during heating but decreases during subsequent cooling (Fig. 8a). The recovery stress during heating may exceed the detwinning stress value ($\sigma_t$) because the energy (or stress) needed to detwin the martensite is significantly lower than the energy (temperature) needed to carry out the reverse transformation. However, when cooling, the recovery stress may drop down almost to zero because of forward transformation. To avoid any loss of recovery stress, the $A_t$ temperature is often chosen to be below the ambient temperature (situation not represented in Figs. 8 or 9). In this case, the SMA is typically cooled down below the $M_f$ temperature for prestraining and stored at a temperature below $A_t$, by liquid nitrogen if needed. Throughout the service life, the SMA then remains in the high-temperature austenitic phase where it keeps the high recovery stress [44]. For Ni–Ti–Nb alloys (Fig. 8b), which have a larger hysteresis, the prestraining is still done at cooled conditions but storage can be at ambient temperature, if it is below $A_t$ [45].

For an Fe–Mn–Si alloy, prestraining (Fig. 8c) is typically done at ambient temperature following path 2 in Fig. 6e. The transformation takes place directly from austenite to martensite without twinning and detwinning. The unloading process for the Fe–Mn–Si alloys, a straight line in Fig. 8c, is sometimes curved, which demonstrates some pseudoelasticity [40,42].

Fig. 9 shows the evolution of the recovery stress in the phase diagrams during heating and cooling. During the heating step, the elastic stress in the SMA first decreases due to the suppressed thermal expansion. Once the stress-temperature path crosses the $A_t$ boundary, a reverse transformation to the austenite phase occurs. This transformation activates the SME, which generates tensile stresses in the SMA, because the contraction is inhibited. However, some of the stress is lost due to thermal expansion. The thermal expansion effect is recovered by thermal contraction.

![Fig. 7. Typical stress–strain–temperature diagrams for Ni–Ti. For Fe–Mn–Si alloys, the pseudoelastic curve disappears, passing from the shape memory effect to an ordinary plastic deformation. Adapted from [37].](image-url)
during subsequent cooling [46]. The slope of the stress variation due to pure thermal expansion/contraction varies depending on the alloy. For example, for an Fe–Mn–Si alloy, considering an elastic modulus that is equal to 200 GPa and a coefficient of thermal expansion (CTE) of $13 \times 10^{-6}/\text{°C}$, this slope would be 2.6 MPa/°C. For a Ni–Ti alloy in the austenite phase, considering an elastic modulus that is equal to 60 GPa and a CTE of $11 \times 10^{-6}/\text{°C}$, the slope would be 0.66 MPa/°C, which is much lower than the slope for the Fe–Mn–Si SMA, as seen in Fig. 9.

The behavior during cooling back to ambient temperature (AT) depends mainly on the width of the hysteresis. For a Ni–Ti SMA with a narrow hysteresis, the recovery stress drops down due to forward transformation (Figs. 8a and 9a). For a wide hysteresis Ni–Ti–Nb alloy, the austenite remains stable as long as the AT is above the $M_s$ boundary, maintaining a high recovery stress (Figs. 8b and 9b). For an Fe–Mn–Si alloy, the recovery stress drops down slightly (Figs. 8c and 9c) or even increases due to thermal contraction. An alloy with AT between $M_s$ and $A_y$ and a narrow temperature hysteresis (Fig. 9a, i.e., Ni–Ti) is not appropriate for acting as permanent prestressing because the recovery stresses decreases considerably after heating and cooling back to the AT. The recovery stresses would even become negligible if the martensite finish temperature, $M_f$, was higher than the AT, a situation that is not shown in Fig. 9.

In the absence of heating or cooling, the SMA is at AT. Therefore, this temperature defines the phase in which the alloy should be stable in civil engineering. Janke et al. [5] proposed that, for external applications in structures, the ambient temperature can be assumed to be situated between $-20$ °C in winter and $60$ °C under intense solar radiation in summer. For this reason, high hysteresis SMAs are required for structural applications.

Going back to SMAs with thermoelastic martensitic transformations, i.e., NiTi, it is also possible to induce martensitic transformation through the application of an external mechanical stress on an SMA that is in the austenite phase ($T > A_y$), as seen in path 5 of Figs. 6c and 7. In this case, only variants with an intrinsic change of shape in the direction of the applied stress will appear (detwinned martensite). The stress–strain diagram would show pseudoplasticity (or superelasticity when referring to engineering applications). This is made up of an initial elastic branch (path 5 in Fig. 7), with the initial elastic modulus of the austenite, a horizontal branch, in which the phase transformation from austenite to martensite is produced by mechanical induction, and another elastic branch, which has the initial modulus of the martensite. After unloading (path 6 in Figs. 6c and 7), the material will return to the origin of the diagram without permanent deformations and by performing a hysteresis loop that dissipates energy, producing a damping effect. For higher initial temperatures ($T > M_d$, where $M_d$ is the threshold temperature above which martensite induced by external forces is not produced), the SMA would undergo an initial elastic deformation followed by a plastic slip deformation at high stresses, similar to the deformation that occurs in ordinary steel (path 7 in Fig. 7, effect not shown in Fig. 6 for Ni–Ti).

For Fe–Mn–Si alloys, it is not possible to have the superelasticity effect because when applying stresses at a temperature that is higher than $A_y$, the alloy will suffer plasticity with an irreversible slip (path 7 in Fig. 6f). However, new findings reveal that some iron-based alloys, i.e., Fe–29Ni–18Co–5Al–8Ta–0.01B (mass %) or Fe–36Mn–8Al–8.6Ni (mass %) alloys, show superelasticity at room temperature, although as was mentioned in the introduction section, these alloys still require further development to be produced at the scale that is needed for civil engineering applications, and their cost would probably be very high.

3. Material properties

The mechanical properties of each Fe–Mn–Si alloy will strongly depend on its composition, heat treatments and the hot and cold work that is performed. For this reason, only some trends can be commented on in this section. Table 3 shows, as an example, the fundamental properties of a solution treated Fe–28Mn–6Si–5Cr SMA after hot working [6].

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![Fig. 8. Schematic stress–strain diagrams of prestraining and generation of recovery stresses during the activation of the reverse transformation in a constrained SMA: (a) Narrow hysteresis alloy, i.e., Ni–Ti; (b) Wide hysteresis alloy, i.e., Ni–Ti–Nb; (c) Fe–Mn–Si alloy.](image)

![Fig. 9. Schematic thermomechanical paths during the activation and cooling of a constrained SMA: (a) Narrow hysteresis alloy, i.e., Ni–Ti; (b) wide hysteresis alloy, i.e., Ni–Ti–Nb; (c) Fe–Mn–Si alloy.](image)
3.1. Recovery stresses

The key factor for the use of iron-based SMAs as prestressing tendons in civil engineering structures is the recovery stress. As seen in Table 3, shape recovery properties for this Fe–28Mn–6Si–5Cr SMA increase significantly after thermomechanical training. However, as different authors note, this treatment requires additional processing steps, which increases the production cost and is applicable only to components that have simple geometries [28,29].

The modulus of elasticity of the Fe–28Mn–6Si–5Cr alloy considered in Table 3 is 170 GPa, which is much higher than that reported for Ni–Ti alloys, which varies from 30 to 98 GPa in austenite to 21–52 GPa in martensite [47]. Other authors have reported a modulus of elasticity in the range between 160 and 200 GPa for an Fe–17Mn–5Si–10Cr–4Ni–1(V, C) alloy [28,40].

Most of the research that has been conducted to date with Fe–Mn–Si alloys has been focused on studying the SME and on having the maximum recovery stress, paying secondary attention to the recovery stresses. However, for prestressing in civil engineering applications, the recovery stress is a fundamental property. The alloy with the maximum recovery stress does not enjoy the maximum recovery stress, as was observed for an Fe–Mn–Si–Cr–Ni alloy with varying amounts of C [48]. The carbon in this alloy increases the plastic yield stress and thus favors the production of stress-induced martensite during prestraining. A moderate addition of carbon (0.12%) can thus improve the recovery stress. However, an excessive addition of carbon (0.18%) will shift the Ms temperature much below room temperature and thus diminish the amount of martensite that is produced during prestraining, which leads to a lower recovery stress. For the recovery stress, on the other hand, a low M, temperature will prevent the austenite that was produced during heating from transforming back to martensite during cooling. The recovery stress in the Fe–Mn–Si–Cr–Ni alloys therefore increases with increasing carbon content [48].

The recovery stresses are not only dependent on the alloy composition but also on the microstructural features such as the grain size or the presence, size and distribution of fine second-phase particles. These features have an influence on the mechanical properties of the alloys on the one hand, but also on the stacking fault energy or the transformation temperatures on the other hand. They are influenced by the thermo-mechanical pre-treatment of the alloy. Table 4 presents the recovery stresses that are reported for different Fe–Mn–Si alloy compositions and different thermomechanical procedures. For an Fe–28Mn–6Si–5Cr SMA, the recovery stresses vary from approximately 130 MPa to 180 MPa, without and with a cycle of thermomechanical training. However, the recovery stresses are substantially higher for alloys that contain small VN or VC precipitates, up to values of 500 and 580 MPa, respectively. Values of 460 MPa and 565 MPa were also reported for alloys that are deformed by equal channel angular pressing (a thermomechanical procedure that has been successfully applied to produce various ultra-fine grained materials) and subsequent annealing as well as cold rolling and annealing [49,50].

No data on the long-time relaxation of the recovery stresses is reported in the literature, although good creep and stress relaxation characteristics for high stress levels have been reported for Fe–Mn–Si SMAs [51], the relaxation being lower for the alloys that contain VC precipitates compared to that of alloys without such precipitates [52].

For structural concrete applications, it is also very important to consider the temperature at which the transformation occurs because high temperatures inside the concrete could cause micro-cracking, deteriorating the mechanical properties of the surrounding concrete. In this sense, the reported Fe–17Mn–5Si–10Cr–4Ni–1(V, C) [28] reached the recovery stresses of 580 MPa after heating to only 130 °C (Table 4). The heating temperature is limited in internal prestressing applications by the bond strength reduction due to the increased temperature in the surrounding concrete. Pull-out tests with ribbed reinforcing bars reported in the literature suggest a reduction in the bond strength of 10–20% for a concrete temperature of up to 200 °C [53–55], although this reduction varies depending on the concrete compressive strength and the ratio between the concrete cover depth and the rebar diameter [56].

3.2. Corrosion resistance

The corrosion resistance of Fe–Mn–Si SMAs has been studied for different alloy compositions in aggressive environments such as H2SO4 and NaCl solutions [57–66]. To the authors’ best knowledge, the corrosion characteristics under the alkaline environment of concrete (pH of 12–13) that surrounds embedded bars have not yet been reported.

In highly oxidizing environments, i.e., H2SO4, Fe–Mn–Si–Cr–Ni–Co) alloys that contain from 8.8 wt% to 12.80 wt% of Cr exhibit a similar or better corrosion resistance than stainless steel 304 in spite of the lower amount of Cr. This behavior is attributed to the high amount of Si in the iron-based SMAs [63,65,66].

The corrosion behavior of the Fe–Mn–Si alloys in NaCl solution is a key point that is still unclear, with some contradictory results in the available literature [68]. On the one hand, Fe–15Mn–7Si–9Cr–Ni alloys did not show passivation or localized pitting behavior in a 3.5% NaCl solution and showed the best behavior when tested as single-phase austenite [69]. However, the training process or the cold rolling reduced the corrosion resistance of the alloys [57]. On the other hand, a recent study reports that Fe–Mn–Si–Cr–Ni–Co alloys that contain from 8.8 wt% to 12.80 wt% of Cr show poor corrosion resistance in a 3.5% NaCl solution compared to that of stainless steel 304 and that these SMAs are prone to pitting corrosion in chloride environments due to their high manganese content [66]. In any case, the higher the Cr content is, the higher the corrosion resistance [59]. The addition of Cu or Rare Earths, mainly La and a small quantity of Ce, could improve the corrosion resistance in NaCl solutions [59,61].

3.3. Weldability

The welding characteristics of Fe–Mn–Si alloys have been studied while considering different welding technologies, such as tungsten-inert gas (TIG), laser beam (LB) and electron beam welding (EB) [67–71]. Although some researchers report only the welding optimization parameters, only a few of them focus on shape recovery stress or strain after welding. The methodology of the tests was, in general terms, as follows: welding two small pieces of a specified Fe–Mn–Si alloy in the
γ-austenite phase, and after that, predefining the sample and obtaining the shape recovery strain or stress. After the welding, the phase structure of the weld zone is still austenite, which is the same as that of the base alloy but includes dendrite crystals. There is no obvious change in the microstructure of the heat-affected zone except for some degree of growth of austenite grains. In general terms, the variation in the shape recovery strain due to welding is within ±10% for the small tested pieces [67,68]; as a result, insignificant changes would be produced for a complete element in which the heat-affected zone is significantly small compared to the overall dimension of the element. However, the corrosion resistance of the welded zones is worse than that of the base materials, due to the micro-segregation and welding stress that occurs within the welded zones. The recovery stresses that are generated in small 10 mm long samples, including the weld zone of an Fe–28Mn–6Si–5Cr–0.53Nb–0.06C alloy, were 250 MPa, independent of the welding technique used (TIG, LB or EB), compared to the 300–330 MPa reported for the non-welded material [68]. The dendritic structures that formed during the welding procedure disappeared after an annealing treatment of the welded pieces, improving the corrosion resistance. Finally, a manufacturing process for shaft and pipe coupling by forming and welding of an Fe–15Mn–5Si–9Cr–3Ni shape memory alloy was recently proposed [72]. The results showed that welding affects mechanical and shape memory properties: the material fractured in the welded zone and the degree of shape recovery decreased 15%. However, the couplings manufactured with this new procedure could develop sufficient coupling force and this manufacturing procedure should be more studied [72].

3.4. Production

In general, Fe–Mn–Si based SMAs are produced by melting and subsequent casting under high vacuum (or in air) followed by thermo-mechanical processing [73]. The composition of Fe–Mn–Si alloys is very similar to a high-Mn stainless steel. For that reason, some authors think that these alloys could probably be conveniently melted and processed by utilizing the production facilities of steels or stainless steels [6]. However, a mass production electric furnace cannot be adopted for the melting of any SMA that contains a large amount of Mn elements and that has a high heat capacity [6]. As Murayama and Kubo remark in the case of an SMA that contains over 20% Mn, the impurities that are contained in the raw Mn materials give rise to fatal faults in the mechanical properties of the SMA. For this reason, decreasing the amount of Mn would be better for a mass production furnace and would decrease the total cost reasonably [6]. In fact, the alloy Fe–17Mn–5Si–10Cr–4Ni–1V, C, which has a relatively low Mn content, has already been produced under normal atmospheric conditions [28,60]. A recent research has highlighted that Fe–Mn–Si alloys with the Mn content above 13 mass% exhibit a good shape recovery strain of more than 2% with Ni and Cr substitutes, while the alloys with the Mn content below 13 mass% indicate a sudden decrease in the shape recovery strain with all the substitutes [74].

An alternative production methodology for the Fe–Mn–Si SMAs by mechanical alloying and subsequent sintering has been explored [73]. This mechanical alloying is a synthetic technique involving solid state reaction among powders due to high energy collisions. Powder metallurgy may offer several advantages for manufacturing industrial products, as it enables the production of alloys in near net shape, minimizing the additional machining required to form the final product. The results show that the mechanical properties and the shape memory effect can be comparable to that alloys manufactured by conventional casting [73].

3.5. Workability at room temperature

Although many authors refer to the good workability of Fe–Mn–Si SMAs, not much specific research has been conducted on this topic. As Sató et al. [13] note, it is fortunate that these Fe–Mn–Si SMAs are quite similar to the TRIP/TWIP steels, which have good workability at room temperature. However, it is sometimes necessary to prevent the SMA from breaking when heavy deformation is needed. Superposition of either a hydrostatic pressure or a proper constraint is effective [13]. Finally, it must be taken into account that the addition of C or N could produce carbide and nitride, respectively, which reduce workability [22].

4. Pilot experiences on the application of Fe–Mn–Si alloys

The application of iron-based SMAs in civil engineering structures is still in a pioneering stage, and only a few pilot applications can be found in the literature. Nevertheless, two applications have succeeded in other related industries: the production of crane rail fishplates [75] to connect finite lengths of rail for heavy duty cranes (i.e., steel factories) and pipe joints for steel pipes for a construction method for tunneling work [6], with both products manufactured by Awaji Materia Co. Application of other SMAs in civil engineering structures, i.e., Ni–Ti or Cu-based alloys, can be found elsewhere [5,47,76–86].

4.1. Shape memory effect

A bridge in Michigan was strengthened through external post-tensioning that was performed with Fe–Mn–Si–Cr SMA rebars in 2001 [87], after a laboratory research study on the possibilities of external reinforcement of shear cracks. In this case, an Fe–28Mn–6Si–5Cr alloy was used that could develop recovery stresses of 255 MPa after heating it to 300 °C. The alloy was manufactured by Nippon Steel Corporation. The effect of extreme environmental temperatures on the restrained recovery stresses of the selected iron-based alloy was studied, and the conclusion was that the variations in the stress were relatively small. The study showed that post-tensioning with SMAs enabled the shear crack to be closed to a large extent, recovering the load capacity of the rehabilitated beam.

Watanabe et al. [88] worked on the reinforcement of an 80 mm long plaster prism specimen with a 1 mm diameter prestressed wire of Fe–27Mn–6Si–5Cr–0.05C alloy. The wires were subjected to pretension strain at room temperature (1%, 2% and 3%), and they were embedded into a plaster matrix. The specimens were then heated up to 250 °C to generate a compressive stress in the matrix. Three-point bending tests were performed for mechanical property characterization as well as pull-out tests. These tests showed that the fracture toughness of plaster could be improved by the wires and that the bending strength of the composite specimen increased with increasing levels of prestrain in the SMA wires.

Small mortar prisms were reinforced with square Fe–28Mn–6Si–5Cr–1(NbC) bars by Sawaguchi et al. [89] (Fig. 10). As was previously mentioned in this paper, Kajiwara et al. had reported that the shape memory properties of Fe–Mn–Si based SMAs could be improved by the fine dispersion of NbC carbides [18]. The mortar specimens, with embedded SMA square bars (2 x 2 mm square section), or with steel bars with the same dimensions, or without any reinforcement, were first cured for two days. The hardened mortar prisms were extracted from the mold, and a high-temperature curing was performed in an autoclave. The first stage of the autoclave curing was performed at 87 °C (360 K) for 24 h, to increase the compressive strength of the mortar matrix. The second stage of the curing was intended to generate the prestress in the SMA, for which two conditions, i.e., 177 °C for 6 h and 247 °C for 30 min, were applied. The mechanical properties of the mortar increased with the 177 °C curing, but they worsened for the 247 °C treatment, except for the surprising results they got from prisms strengthened with steel subjected to the high temperature curing (Fig. 10). The authors concluded that the iron-based SMAs were usable for producing prestress in the mortar because the bending strength increased significantly in the specimens with the SMA respect to the specimens reinforced with steel or the non-reinforced specimens. Sawaguchi et al. also highlighted that further strengthening could be achieved by lowering the reverse transformation temperatures of the SMAs, thus avoiding significant thermal damage to the mortar matrix.

Watanabe et al. [90] investigated the fabrication method and mechanical properties of smart composites that were reinforced with Fe–28Mn–6Si–5Cr SMA machining chips. These chips are a by-product of the fabrication of one of the main industrial applications of iron-based shape memory alloys: the pipe joining of steel pipes for tunneling constructions. The study was also conducted at a feasibility level, using 80 mm long plaster prisms and obtaining good preliminary results.

Lee et al. [46] studied the recovery stress behavior of an Fe–17Mn–5Si–10Cr–4Ni–1(V, C) shape memory alloy for prestressing concrete structures. The prestressing effect due to the shape memory effect was simulated by a series of tests that involved pre-straining the material followed by heating and cooling at a constant strain. The tested SMA specimens came from
a 15 kg alloy ingot that had been induction melted under normal atmospheric conditions. The iron shape memory alloy used in this research is very promising because it shows a very wide hysteresis from −60 °C to 103 °C that is determined from a differential scanning calorimetry (DSC) analyses. The authors conducted tests for different heating temperatures and initial pre-strains, obtaining the highest recovery stress at 400 MPa, for a pre-strain of 4% and a heating temperature of 160 °C. In this study, the behavior under additional loading after prestressing was also studied, both under cyclic mechanical loading and under cyclic thermal loading. These tests showed that the prestressed SMA undergoes an inelastic deformation during the first cycle, which leads to a reduction in the recovery stresses. However, after the first cycle, the SMA behaved elastically, and the level of stress remained stable on subsequent cycles. Finally, the authors experimentally showed that this loss could be fully recovered by reheating the SMA element.

The same alloy but from a much larger cast has recently been used to manufacture 20 mm wide and 750–950 mm long ribbed Fe-SMA strips for near-surface mounted reinforcements [91]. The recovery stress after heating to 160 °C and cooling down to room temperature measured in a tensile machine with climate chamber was in the range of 250–300 MPa. The bond behavior tested in lap-shear experiments was in the usual range of reinforcing steel. The strips were embedded in 700 mm long mortar prisms of 35 × 50 mm² cross-section and heated by a current supply with approximately 14 A/mm² and a voltage of approximately 15 V/m during less than 7 s. Strain measurements on one of the prism indicated a compression stress in the concrete of more than 3 MPa.

4.2. Damping

As far as the authors know, no real application has been reported to date that is related to damping in civil engineering structures when using iron-based shape memory alloys. However, the large damping effect in the TRIP/TWIP Fe–Mn alloys due to the martensitic transformation has been studied in depth [92–95]. More recently, based on very similar principles, studies on the vibration mitigation by the martensitic transformation in Fe-based alloys for their application as a seismic damping material has also been reported [96,97]. The Fe–28Mn–6Si–5Cr–0.5NbC SMAs that was studied by Sawaguchi et al. [96,97] showed a significant damping capacity in the large-strain amplitude region above 0.1%. The specific damping capacity increased with increasing strain amplitude and reached a saturation value of approximately 80% above a strain amplitude of 0.4% (Fig. 11). Their analysis revealed that the tension-induced martensite reversely transforms into austenite by compression. The authors thus concluded that the vibration mitigation during the cyclic tension–compression is accomplished by the reversible motion of the γ/ε interfaces.

The hypothetical use of the superelastic Fe–Ni–Co–Al–Ta–B alloy developed by Tanaka et al. [7] in a wire-based smart natural rubber bearing was recently investigated [98]. The authors conducted a study that was based on finite element modeling of different configurations of bearings. They concluded that the ferrous superelastic SMA is a good candidate due to its high superelastic strain range and very low austenite finish temperature (Af).

5. Research needs

A large number of researches that study the behavior of Fe–Mn–Si alloys have been performed, but some aspects are still not clearly understood, and more research is needed on specific topics. For example, most of the research on thermomechanical treatments during the production process has focused on improving the recovery strain, whereas the recovery stresses are the true key point for the application of the SMAs as prestressing reinforcements. Therefore, a systematic study on the optimization of the recovery stresses for different alloy compositions is needed. Additionally, with respect to the material properties, relaxation and fatigue properties have not been studied in-depth. More information can be found on the corrosion behavior, although for prestressing applications, it would be necessary to know the corrosion characteristics under the alkaline environment of concrete. Large-scale production also needs research to allow having the large amounts of materials that are needed in civil engineering applications. For the development of some products or devices, it is also necessary to improve the knowledge on weldability. Different papers address the weldability properties of the iron-based SMAs in the austenite phase, but it would be very useful to know the temperature-affected zone in a welded alloy in martensite for different welding technologies.

At the application level, most of the research that was conducted during recent years for Ni–Ti alloys exploiting the SME and the damping capacity can be adapted for iron-based shape memory alloys. In this paper, some pilot experiences on the application of Fe–Mn–Si alloys have been highlighted, but many others could be conducted in the near future. For example, it is possible to consider concrete that is prestressed by Fe–Mn–Si alloy short fibers, which has been already proposed for Ni–Ti alloys [99,100], or on the confinement of concrete columns, which has been broadly tested with Ni–Ti or Ni–Ti–Nb alloys [101–103]. For many concrete applications, it would be necessary to have more information about the bond strength reduction due to the temperature increase of the embedded SMA. In fact, the tests that were reported...
in the literature and commented on in this paper address the heating of the concrete and not the direct heating of the rebars.

6. Conclusions

Research on new iron-based SMAs has experienced a considerable boost during the past decade, developing Fe–Mn–Si alloys that can generate high recovery stresses at lower temperatures compared to the first alloys developed in the 1980s. The current state of the art related to these SMAs for their application in civil engineering structures has been presented in this paper. The following conclusions can be drawn:

- The use of iron-based SMA tendons for the repairing of existing structures or the reinforcing of new structures is a promising field for the near future. Such iron-based SMA tendons have several advantages, such as no friction losses, no anchor heads and ducts required, and no space necessary for applying the force with a hydraulic device.
- The lower cost of these new iron-based SMAs results not only from the use of cheap iron (Fe) but also from the possibility of melting and producing these SMAs under normal atmospheric conditions.
- These new Fe–Mn–Si alloys have a wider temperature transformation hysteresis and a higher elastic stiffness than other SMAs, i.e., Ni–Ti alloys. They also present a good workability, corrosion resistance and weldability.
- New Fe–Mn–Si alloys with fine precipitates have been developed during recent years and allow high recovery stresses without the need for thermomechanical training.
- More research is still needed to allow a better understanding of the material behavior. The key topics have been summarized in this paper, and they constitute a starting point for future research in this field.

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A. Cladera et al. / Construction and Building Materials 63 (2014) 281–293


