Chapter 11

Designing High-Mn Steels

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11.1 Introduction

High-Mn austenitic steels have attractive mechanical and functional properties associated with characteristic plasticity mechanisms in the γ -austenite that has a face-centred cubic (FCC) structure. These include an extended dislocation glide, mechanical twinning, and mechanical martensitic transformation into ϵ -martensite with a hexagonal close-packed (HCP) structure and further to α' -martensite with a body-centred cubic or tetragonal (BCC/BCT) structure. Extensive research has been conducted on the deformation mechanism of high-Mn steels, particularly on the transformation- and twinning-induced plasticity (TRIP/TWIP) effects, which lead to the remarkable combination of high strength, ductility, and toughness.

Hereinafter, high-Mn austenitic steel is defined in a broad sense as the steel characterised by a high concentration of Mn, an initially fully austenitic phase state, and low-to-moderate stacking fault energies (SFEs). This category includes traditional Hadfield steels (Hadfield 1888), cryogenic high-Mn steels (Charles et al. 1981; Kim et al. 2015; Sohn et al. 2015), non-magnetic high-Mn steels, TRIP/TWIP steels (Grassel et al. 2000; Bouaziz et al. 2011; Cooman et al. 2018), high damping Fe–Mn alloys (Wang et al. 2019; Shin et al. 2017; Jee et al. 1997), and Fe–Mn–Si-based shape-memory alloys (SMAs) (Sato et al. 1982; Otsuka et al. 1990). In most cases, steels contain other alloying elements, such as Cr, Ni, Al, Si, C, and N, and the concentration of Mn depends on the alloy system (e.g. 12–29 wt% in the case of binary Fe–Mn systems).

This article focuses on a design strategy for improving the tensile and fatigue properties of high-Mn steels from microstructural, thermodynamic, and crystallographic perspectives. Referring to the concept of ‘plaston’, a plastically deformed infinitesimal volume element, various plasticity mechanisms in high-Mn steels are visualised

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by the deformation of deltahedral (either tetrahedral or octahedral) models. These are the minimum volumetric units representing the atomic coordination in each phase. The extended dislocation glide of the γ lattice, mechanical γ twinning, and $\gamma \rightarrow \epsilon$ martensitic transformation commonly involve the twinning shear of the deltahedrons as an elementary process, and the α' -martensite is associated with the Bain distortion of the deltahedrons.

Section 11.2 presents a brief review of the atomic rearrangement and the related dislocation-based models for the extended dislocation glide, mechanical γ twinning, and mechanical $\gamma \rightarrow \epsilon$ martensitic transformation. In Sect. 11.3, the polyhedron models and their mutual transformations are systematically drawn for the four plasticity mechanisms, including the $\gamma \rightarrow \alpha'$ martensitic transformation, along with the former three. In Sect. 11.4, the design of the tensile properties of high-Mn steels is described, which considers the selection rule of the plasticity mechanisms, microstructure development, and strain hardening mechanisms. Structures developed at the intersection of crossing habit-plane variants of the planar deformation products are highlighted. The atomic rearrangement caused by the double shears is visualised with distortion or kinking of the deltahedron models. In Sect. 11.5, the design of the fatigue properties of high-Mn steels is described. The reversible back-and-forth motion of partials with the aid of the unidirectionality of the shear deformation of the deltahedrons plays an important role in these properties.

This article addresses only microstructural, thermodynamic, and crystallographic aspects of the plasticity mechanisms. It does not address the improvement of mechanical properties by grain refinement, precipitation, and interstitial atoms, all of which have been extensively studied, especially to improve the yield stress, which is a critical weak point of TRIP/TWIP steels. For these issues, review articles (Bouaziz et al. 2011; Cooman et al. 2018, 2012; Chowdhury et al. 2017; Zambrano 2018; Lee et al. 2017; Chen et al. 2017) or other chapters of this book can be referred.

11.2 Plasticity Mechanisms in γ -austenite

The extended dislocation glide, mechanical γ twinning, and the mechanical $\gamma \rightarrow \epsilon$ martensitic transformation proceed via shear displacement on the $\{1\ 1\ 1\}_{\gamma}$ planes, and their products commonly have a planar morphology with crystal habit on the $\{1\ 1\ 1\}_{\gamma}$ planes. The plated deformation products and the remaining γ -austenite form a lamellar structure. As an example of the lamellar structure, Fig. 11.1 shows a differential interference micrograph of the surface relief on an Fe–30Mn–5Si–1Al alloy, which was formed by the shear displacement of the pre-polished surface on a specimen plastically deformed to 3%. The banded surface relief is caused by the homogeneous shear associated with the mechanical $\gamma \rightarrow \epsilon$ martensitic transformation, which is the predominant plasticity mechanism in this alloy. ϵ -Martensite bands are formed on a unique $\{1\ 1\ 1\}_{\gamma}$ plane in certain grains (e.g. A in Fig. 11.1), while ϵ -martensite plates on different $\{1\ 1\ 1\}_{\gamma}$ planes cross in the other grains (e.g. B in Fig. 11.1). The frequent appearance of the annealing twin boundaries, some of

Fig. 11.1 Optical micrograph of the surface relief on the pre-polished surface of tensile deformed Fe–30Mn–5Si–1Al alloy

which are indicated by arrows in Fig. 11.1, is also a microstructural characteristic of high-Mn steels with low-to-moderate SFEs.

The shear-type atomic displacements associated with the planar plasticity mechanisms on the $\{1\ 1\ 1\}_\gamma$ planes are shown in Fig. 11.2. With respect to the first close-packed layer A, there are two possible second layers, B and C, with different relative atomic positions, as shown in Fig. 11.2a, b, respectively. The atomic positions on the third $\{1\ 1\ 1\}_\gamma$ layer indicated in Fig. 11.2c, d) represent two types of

Fig. 11.2 The stacking sequence of the closest packing atomic planes and their change by the shear displacement of the planes. **a** AB stacking unit, **b** AC stacking unit, **c** ABC stacking, **d** ABA stacking, **e** γ -austenite, **f** lattice slip, **g** γ twin, **h** ϵ -martensite, **i** extended dislocation glide, **j** mechanical γ twinning, and **k** mechanical $\gamma \rightarrow \epsilon$ martensitic transformation

periodic stacking sequences of the close-packed layers, ABCABC or ABABAB, which correspond to the FCC and structures, respectively.

Figure 11.2e–h shows the atomic arrangements in the γ -austenite and the three plastic deformation products, which are viewed from the $[1\ 0\ 1]_\gamma$ direction. The perfect dislocation glide ($e \rightarrow f$) does not change the stacking sequence of ABCABC, as the atoms move to the next position of the same type, for example, C to C. Mechanical twinning ($e \rightarrow g$) is the homogeneous shear displacement of atoms, which changes the atomic positions regularly ($A \rightarrow B$, $B \rightarrow A$, and $A \rightarrow C$), and the stacking sequence changes from ABCABC to ACBACB, as shown in Fig. 11.2g. The resulting structure remains unchanged, whereas the crystallographic orientation is modified into a mirror-symmetric atomic arrangement with respect to the $\{1\ 1\ 1\}_\gamma$ plane. The $\gamma \rightarrow \epsilon$ martensitic transformation ($e \rightarrow h$) is a regularised shear displacement of atoms on the alternate $\{1\ 1\ 1\}_\gamma$ planes. The stacking sequence changes from ABCABC to ABABAB, as shown in Fig. 11.2h, accompanying the structural change from FCC to HCP.

Figure 11.2i–k shows the three plasticity mechanisms as the motion of partial dislocations of the type $1/6\{1\ 1\ 1\} \langle 1\ 1\ 2 \rangle_\gamma$ (Shockley partials). An extended dislocation in (e) consists of a few (leading and trailing) partials and stacking faults (SFs) between them. The passage of the extended dislocation causes a two-step atomic displacement from C to B and then to the next C, as shown in Fig. 11.2b, i. During this process, the stacking sequence of the SF changes locally to ABA. The mechanical γ twinning (j) and the mechanical $\gamma \rightarrow \epsilon$ martensitic transformation (k) proceed via the regularised group motion of partials on every plane or on alternate $\{1\ 1\ 1\}_\gamma$ planes, respectively. In other words, they are regarded as the ordering of fully extended SFs.

Owing to the common elementary process consisting of the partial–SF units, the slip band, γ twins, and ϵ -martensite have similar morphological and crystallographic characteristics (plated form and crystal habit on the $\{1\ 1\ 1\}_\gamma$ plane), resulting in a generally high strain hardening capability. The selection of the plasticity mechanisms essentially depends on thermodynamic conditions, which determine the extension width of SF and the degree of ordering of SFs. However, when the plasticity mechanisms are competing thermodynamically, crystallographic orientation, and dislocation interaction can also affect the deformation microstructure.

For this reason, the Thomson tetrahedron (Fig. 11.3) is useful. The faces of the tetrahedron correspond to the $\{1\ 1\ 1\}_\gamma$ planes, and the edges of the tetrahedron correspond to the six $\langle 1\ 1\ 0 \rangle_\gamma$ directions of the FCC structure. The corners of the tetrahedron are denoted by A, B, C, and D, and the midpoints of the opposite faces are denoted by α , β , γ , and δ . The planes opposite A, B, C, and D are denoted by a, b, c, and d, respectively, on the outer surface, and $-a$, $-b$, $-c$, and $-d$ on the inner surface. The $\langle 1\ 1\ 2 \rangle_\gamma$ Shockley partials are represented by Roman–Greek pairs (arrows from corner to mid-point). This enables us to discuss operational shears, their interactions, and the crystallographic orientation of the deformation products with respect to the crystallographic orientation of the γ -austenite, which is described in Sect. 11.4.

Fig. 11.3 Notations of the Thomson tetrahedron

11.3 Polyhedron Models for FCC Plasticity Mechanisms

Interstitial lattice sites in FCC, HCP, and BCC structures are generally visualised by either tetrahedrons or octahedrons, whose vertices are at the centre of neighbouring atoms. These polyhedron models are widely used to discuss the distribution of interstitial atoms and their transitions during phase transformations. They can be used as unit volume elements to entirely fill the lattice spaces instead of cubic or hexagonal unit lattices. In this section, we visualise the plasticity mechanisms as distortions of the polyhedron models. Here, the extended dislocation, γ twin, and ϵ -martensite can be composed of a regular tetrahedron, regular octahedron, and their twin-sheared products. The α' -martensite can also be visualised by the distorted tetrahedrons produced by the so-called Bain distortion of the γ -austenite.

Figure 11.4 presents the four-unit polyhedrons. Figure 11.4a shows the atomic arrangement in the γ -austenite, in which a regular tetrahedron and regular octahedrons are drawn as the unit volume elements. Eight such tetrahedrons and four such octahedrons are included in an FCC unit lattice. The ratios between the tetrahedron and octahedron in terms of volume, number density, and space occupancy are 1:4, 2:1, and 1:2, respectively. When an $a/6 (1\ 1\ 1) [-1\ 2\ -1]_{\gamma}$ partial passes on the upper atomic layer in Fig. 11.2a, the regular tetrahedron and the regular octahedron are subjected to twinning shear. Hereinafter, we refer to the distorted tetrahedron and octahedron in Fig. 11.2b as the “twin-sheared” tetrahedron and octahedron. By this twinning shear, one side of the regular tetrahedron is extended from $(\sqrt{2}/2)a_{\gamma}$ to a_{γ} , with the length of the other sides remaining unchanged, as shown in Fig. 11.2c, d. Note that the regular octahedron consists of four twin-sheared tetrahedrons, as shown

Fig. 11.4 Four-unit polyhedrons. **a** Regular tetrahedron and regular octahedron in the γ -austenite, **b** twin-sheared tetrahedron and twin-sheared octahedron after a partial passage, **c** a twin-sheared tetrahedron inside the original regular octahedron, **d** mutual transformation between the regular tetrahedron and twin-sheared tetrahedron from the **c** state

in Fig. 11.2c, while the twin-sheared octahedron consists of two regular tetrahedrons and two twin-sheared tetrahedrons, as shown in Fig. 11.2d.

Using these four-unit polyhedrons, we can fill the FCC lattice, including the SFs, γ twins, ϵ -martensite (= HCP lattice), and partials. An SF is a monolayer consisting of twin-sheared tetrahedrons and twin-sheared octahedrons, most of which can be replaced with regular tetrahedrons and regular octahedrons with different orientations according to the interrelation between Fig. 11.2c, d. A γ twin can be built by stacking SF monolayers, while ϵ -martensite can be built by alternating stacking of the SF and γ monolayers. The partial is the boundary between the twin-sheared monolayer and the γ monolayer. Along the partial, the polyhedrons are elastically distorted to adjust the atomic displacement of $a/6(1\ 1\ 1)[-1\ 2\ -1]_\gamma$.

Figure 11.5 demonstrates the deformation of the polyhedrons associated with the $\gamma \rightarrow \alpha'$ martensitic transformation. The two regular tetrahedrons (a) are compressed in the 001 axis and stretched in the 100 and 010 axes (d) by the Bain distortion. The regular octahedron (b) is also distorted in the same manner (e). Hereafter, we refer to the distorted tetrahedron and octahedron in α' -martensite as the α -tetrahedron and α -octahedron, respectively. The twin-sheared tetrahedron residing in the regular octahedron (c) is transferred into the α -tetrahedron (f). Comparing Fig. 11.5e, f shows that the α -octahedron consists of four α -tetrahedrons.

Fig. 11.5 Polyhedron models for the Bain distortion during $\gamma \rightarrow \alpha'$ martensitic transformation. **a** Regular tetrahedrons in a γ unit lattice, **b** a regular octahedron in the γ unit lattice, **c** a twin-sheared tetrahedron inside the regular octahedron, **d** distorted tetrahedrons in a α' unit lattice, **e** a distorted octahedron in the α' unit lattice, **f** and a distorted tetrahedron inside the distorted octahedron

11.4 Plasticity Mechanisms Under Tensile Loading

11.4.1 Selection Rule and Generation Processes

There are several transmission electron microscopy (TEM) and high-resolution transmission electron microscopy (HRTEM) research studies on the SF as the embryo of ϵ -martensite and γ twins (Bray and Howe 1996; Brooks et al. 1979a, b). In designing high-Mn austenitic TRIP/TWIP steels, the stacking fault energy of the γ -austenite (SFE_γ) is used as an important indicator to predict the appearance of mechanical γ twinning and mechanical ϵ -martensitic transformation (Choi et al. 2020; Lee et al. 2019, 2018, 2010; Lu et al. 2017; Zambrano 2016; Das 2015; Xiong et al. 2014; Curtze et al. 2011; Nakano and Jacques 2010; Curtze and Kuokkala 2010; Tian and Zhang 2009a, b; Saeed-Akbari et al. 2009; Tian et al. 2008; Dumay et al. 2008; Cotes et al. 2004, 1999; Allain et al. 2004; Lee and Choi 2000; Grassel et al. 1997). The SFE_γ affects the extension width of the dislocation in the γ -austenite, which is determined when the repulsive force between the leading and trailing partials is equivalent to the attractive force to reduce the surface energy. The extension width increases with a decrease in the SFE_γ . The generation of the mechanical γ twins and ϵ -martensite requires a sufficiently wide extension width that is referred to as the ‘infinite separation’ (Byun 2003; Copley and Kear 1968) with respect to the grain size; thus, these plasticity mechanisms appear in low SFE_γ ranges. Remy schematically indicated the

Fig. 11.6 Plasticity mechanisms observed at room temperature as a function of stacking fault energy of the austenite. (Reproduced with permission from Mater. Sci. Eng. 26 (1976) 123, Copyright 1976, Elsevier Sequoia S.A., Lausanne)

SFE_{γ} ranges for the plasticity mechanisms, as shown in Fig. 11.6 (Remy and Pineau 1976). Although the threshold values differ depending on the study (Nakano and Jacques 2010; Saeed-Akbari et al. 2009; Allain et al. 2004; Pierce et al. 2014), most studies have reported that the order of appearance is dislocation glide, mechanical γ twinning, and mechanical $\gamma \rightarrow \epsilon$ martensitic transformation with decreasing SFE_{γ} .

The appearance of the mechanical γ twinning and the mechanical $\gamma \rightarrow \epsilon$ martensitic transformation in certain SFE_{γ} ranges is associated with their generation mechanisms. Various nucleation mechanisms have been proposed to describe how and why the SFs are regularised on every or on alternate $\{111\}$ plane to form γ twins or ϵ -martensite, respectively. For the nucleation of the mechanical γ twin, these are listed as follows: (1) the pole mechanism (Venables 1974), (2) successive deviation of Shockley partials leaving Frank pairs at the Lomer–Cottrell lock (Cohen and Weertman 1963) (3) stair-rod cross-slip mechanism (Fujita and Mori 1975; Mori and Fujita 1980), (4) interaction of co-planar perfect dislocations into three-atomic-layer Shockley-partial embryos (Mahajan and Chin 1973), among others. Among these, (1) the pole mechanism (Hoshino et al. 1992), (3) stair-rod cross-slip mechanism (Fujita and Ueda 1972), and (4) co-planar dislocation interaction model (Mahajan et al. 1977) have also been applied to ϵ -martensite nucleation. Many microstructural studies have supported these models (Idrissi et al. 2010, 2013; Mahato et al. 2017; Steinmetz et al. 2013).

Microstructural observation has revealed that the overlapping of SFs into ϵ -martensite proceeds approximately but irregularly (Brooks et al. 1979b). For the gradual growth of the ϵ -martensite, the thermodynamic stability of the ϵ -phase is an additional requirement. The Gibbs free energy difference associated with the $\gamma \rightarrow \epsilon$ martensitic transformation and the SFE_{γ} are interrelated via the following equation

(Olson and Cohen 1976):

$$\Gamma = 2\rho\left(\Delta G_{\text{chem}}^{\gamma \rightarrow \varepsilon} + \Delta G_{\text{mag}}^{\gamma \rightarrow \varepsilon}\right) + 2\sigma^{\gamma/\varepsilon}, \quad (11.1)$$

where Γ is the stacking fault energy of the austenite, ρ is the molar surface density of atoms in the (111) planes, $\Delta G_{\text{chem}}^{\gamma \rightarrow \varepsilon}$ is the chemical Gibbs free energy difference between the γ - and ε -phases, $\Delta G_{\text{mag}}^{\gamma \rightarrow \varepsilon}$ is the magnetic contribution to the Gibbs free energy differences, and $\sigma^{\gamma/\varepsilon}$ is the energy per surface unit of the (111) interface between the γ - and ε -phases. Equation (11.1) was developed based on the concept that the stacking fault is an ε -martensite embryo consisting of a double atomic layer, as shown in Fig. 11.5. There are three types of martensitic transformation paths, namely, $\gamma \rightarrow \varepsilon$, $\gamma \rightarrow \alpha'$, and $\gamma \rightarrow \varepsilon \rightarrow \alpha'$, depending on the Gibbs free energies of the phases, G^γ , G^ε , and G^α .

Studies have attempted to obtain reliable SFE_γ values. SFE_γ can be determined by measuring the radius of the nodes of partial dislocations using TEM (Kim and Cooman 2011; Pierce et al. 2012), measuring the extension width with the TEM weak-beam method, measuring peak broadening in XRD (Tian et al. 2008; Balogh et al. 2006; Barman et al. 2014; Schramm and Reed 1975; Tian and Zhang 2009c), and by thermodynamic and first principles calculations (Zambrano 2016; Xiong et al. 2014; Curtze et al. 2011; Nakano and Jacques 2010; Saeed-Akbari et al. 2009; Dumay et al. 2008; Allain et al. 2004; Pierce et al. 2014; Tian et al. 2017; Medvedeva et al. 2014). The critical SFE_γ values vary depending on the methods used by the investigators, alloy systems, and the method used for calculating/measuring SFE_γ . Some investigators have proposed Γ (mJ/m²) ranges for the appearance of the plasticity mechanisms, for example, $12 < \Gamma < 35$ for mechanical γ twinning and $\Gamma < 18$ for the mechanical $\gamma \rightarrow \varepsilon$ martensitic transformation (Allain et al. 2004).

In the case of the $\gamma \rightarrow \varepsilon$ martensitic transformation, the corresponding temperature is a direct indicator for predicting its occurrence. The γ - and ε -phases have equal Gibbs free energies at the thermodynamic equilibrium temperature of T_0 . The forward $\gamma \rightarrow \varepsilon$ martensitic transformation starts at $M_s < T_0$ and ends at M_f upon cooling; the reverse $\varepsilon \rightarrow \gamma$ martensitic transformation starts at $A_s > T_0$ and ends at A_f upon heating. Undercooling and overheating are required to obtain a sufficient driving force, $\Delta G^{\gamma \rightarrow \varepsilon}(M_s)$ or $\Delta G^{\varepsilon \rightarrow \gamma}(A_s)$, respectively, for the transformations. The relationship among the transformation temperatures and the driving forces is shown in Fig. 11.7a. Mechanical martensitic transformations are divided into two groups depending on the nucleation mechanism, as schematically shown in Fig. 11.7b. At temperatures immediately above M_s , spontaneous martensitic transformation occurs by a ‘stress-assisted’ nucleation mechanism, as the external stress reaches the critical value required to compensate the lack of a chemical driving force for the transformation. Plastic yielding occurs at the critical stress for the martensitic transformation, σ_M , that linearly increases with the deformation temperature. Above M_s^σ , σ_M becomes greater than the yielding stress by slip, σ_y . No more spontaneous transformation occurs; however, the martensitic transformation is still available below M_d

Fig. 11.7 Martensitic transformation temperatures for **a** spontaneous transformation on cooling and heating and **b** mechanical transformation with stress-assisted and strain-induced nucleation mechanisms. (Reproduced with permission from Journal of the Less Common Metals 28 (1972) 107, Copyright 1972, Elsevier Sequoia S.A., Lausanne)

via the ‘strain-induced’ nucleation mechanism where dislocations act as nucleation sites for the martensite.

This concept of strain-induced nucleation was first proposed by Olson and Cohen for $\gamma \rightarrow \alpha'$ martensitic transformation (Olson and Cohen 1972). ϵ -Martensite was treated as a precursor or an intermediate phase during the $\gamma \rightarrow \alpha'$ martensitic transformation. The formation of α' -martensite was interpreted by a double shear mechanism, which is described in Sect. 11.4.2. Subsequently, Andersson applied the concept to the $\gamma \rightarrow \epsilon$ martensitic transformation (Andersson and Stalmans 1998).

11.4.2 Transformation- and Twinning-Induced Plasticities

In recent decades, TWIP and TRIP have been extensively studied in high-Mn steels. The TWIP effect is interpreted to be a result of the so-called dynamic Hall–Petch effect, which is a continuous grain subdivision by planar mechanical twins during loading. This effect causes considerable strain hardening and suppresses local necking, and consequently, a good combination of high ultimate tensile stress and high elongation is available. A similar effect is expected for the mechanical $\gamma \rightarrow \epsilon$ martensitic transformation with a similar planar morphology. Further martensitic transformation into α' -martensite can also contribute to strain hardening. In this case, however, the strain hardening is not due to the dynamic Hall–Petch effect but by the intrusion of a high density of dislocations through lattice expansion associated with the martensitic transformation. Originally, the term TRIP was given to the superior plasticity of an Fe–Ni alloy undergoing a direct $\gamma \rightarrow \alpha'$ martensitic transformation.

Fig. 11.8 Transformation- and twinning-induced plasticities and three types of martensitic transformation paths: **a** α -TRIP, **b** ε -TRIP and TWIP, and **c** double TRIP

The three types of TRIP effects together with the TRIP effect are summarised in Fig. 11.8.

As an example of the tensile behaviour dependent on the plasticity mechanisms, Fig. 11.9 shows the stress–strain curves of Fe–30Mn–(6– x)Si– x Al alloys ($x = 0$ to 6 in wt%) (Nikulin et al. 2013). In these alloys, the formation of α' -martensite is strongly inhibited, and thus, the tensile behaviour can be interpreted by ε -TRIP, TWIP, and dislocation glide. The alloy with $x = 0$ is a typical SMA, which plastically deforms via mechanical $\gamma \rightarrow \varepsilon$ martensitic transformation. As the concentration of Al that replaces Si increases, the value of SFE_γ increases. The mechanical $\gamma \rightarrow \varepsilon$ martensitic transformation appears for alloys up to $x = 2$. The elongation increases with the change in the nucleation mechanism from stress-assisted to strain induced. The alloy with $x = 2$ is a typical TRIP steel, and the alloy with $x = 3$ is a typical TWIP steel and has the highest elongation; a further increase in the concentration of Al results in a decrease in both strength and elongation.

Fig. 11.9 Stress–strain curves of Fe–30Mn–(6– x)Si– x Al alloys ($x = 0$ to 6). (Reprinted with permission from Mater. Sci. Eng. A 587 (2013) 192, Copyright 2013, Elsevier B.V.)

The extended dislocation glide can also contribute to the strain hardening due to the suppression of the recovery, which is the stress relaxation associated with dislocation rearrangement into a cell structure. There is a variation in the dislocation structures: Yoo and Park have reported a TWIP-like hardening mechanism by a banded dislocation structure in an Fe–28Mn–9Al–0.8C alloy ($\Gamma \approx 85 \text{ mJ/m}^2$), which is called microband-induced plasticity (MBIP) (Yoo and Park 2008); Gutierrez-Urrutia and Raabe have reported that multiple strain hardening is caused by dislocation substructures, such as Taylor lattices, cell blocks, and dislocation cells (Gutierrez-Urrutia and Raabe 2012).

11.4.3 Martensite/twin Variants

Martensite/twin variants are important microstructural components in high-Mn steels; these variants and γ -austenite all have crystallographic orientation relationships (ORs).

The OR of the mechanical ε -martensite with respect to the γ -austenite, which is called the Shoji–Nishiyama (S–N) OR, is shown as follows:

$$\{111\}_{\gamma} // \{001\}_{\varepsilon}, \langle 11-2 \rangle_{\gamma} // \langle 1-100 \rangle_{\varepsilon} \text{ (S–N OR)}$$

There are four ε -martensite variants depending on the habit plane denoted by a, b, c, or d, using the notation of the Thompson tetrahedron in Fig. 11.3.

A remarkable microstructural event is the formation of a new crystal phase or a new crystallographic variant by crossing shears at the intersection of different habit-plane variants of ε -martensite. In Fe–Cr–Ni stainless steels (Olson and Cohen 1972; Venables 1962; Bogers and Burgers 1964; Lee et al. 2013) and Fe–high-Mn steels, α' -martensite is reported to form at the intersection of the ε -martensite habit-plane variants. In Fe–Mn–Si SMAs, a wide variety of other intersection products have been reported, such as ε -twins (Matsumoto et al. 1994; Zhang and Sawaguchi 2018), γ -twins (Matsumoto et al. 1994), and the ε -phase rotated 90° from the original ε -phase (ε_{90}) (Yang and Wayman 1992a; b). In Fe–30Mn–4Si–2Al TRIP steel, the γ -phase rotated 90° from the parent γ -austenite (γ_{90}) (Zhang et al. 2011a; b) has also been reported.

These intersection reactions act not only as a self-accommodation mechanism but also as a source, obstacle, or sink of dislocations by providing new interfaces with surrounding phases and volumetric changes. To control the mechanical and functional properties, it is important to clarify the selection rule of the intersection reactions and their contributions to macroscopic properties. The thermodynamic stability of γ -, ε -, and α' -phases is evidently an important factor affecting the selection of the three thermodynamically distinct intersection reactions, that is, the secondary martensitic transformation into α' , the reverse martensitic transformation into γ , and the plastic deformation (e.g. twinning) of ε .

Zhang et al. (2011a) pointed out that the directions of the crossing shears with respect to the intersection axis, either 90° (Type I) or 30° (Type II), also determines the intersection products. Figure 11.10 shows the EBSD images of the structures at

Fig. 11.10 EBSD images for structures at the intersection of ϵ -martensite variants in an Fe–30Mn–4Si–2Al alloy. (Reprinted with permission from *Philos. Mag. Letters* 91 (2011) 563, Copyright 2011, Taylor & Francis)

the intersections of crossing ϵ -martensite variants in Fe–30Mn–4Si–2Al steel tensile deformed to 10%. Figure 11.10a–c shows a single variant structure in the grain close to $\langle 4\ 1\ 4 \rangle_\gamma$. Two types of double variant structures are shown at an orientation on the $\langle 0\ 0\ 1 \rangle - \langle 1\ 0\ 1 \rangle_\gamma$ boundary in Fig. 11.10d–f and the $\langle 0\ 0\ 1 \rangle - \langle 1\ 1\ 1 \rangle_\gamma$ boundary in Fig. 11.10g–i. Using Schmid's law, plausible crossing shears are identified as Type I for the former and Type II for the latter. The $\gamma\ 90^\circ$ phase was observed at the Type I intersection, while a non-basal slip was observed at the Type II intersection. At an orientation close to $\langle 0\ 0\ 1 \rangle_\gamma$ in Fig. 11.10j–l, the martensite fraction is significantly lower than that of the other grains because of the low Schmid factors of Shockley partials compared with those of perfect dislocations.

The orientation dependence and double shear mechanisms for various intersection products are summarised in Fig. 11.11. Starting from the γ -austenite (a) represented by the regular tetrahedron, (b) one-third, $\mathbf{T}/3$, (c) half, $\mathbf{T}/2$, and (d) full twinning, $\mathbf{T}$, shears along $\mathbf{C}\delta/\mathbf{d}$ are considered as intermediate states. Bogers and Burgers proposed that the second shear on the non-conjugate plane with a magnitude of $\mathbf{T}/2$ can produce α' -martensite (Bogers and Burgers 1964). Figure 11.10e is one such example produced by the second shear along $\mathbf{B}\alpha/\mathbf{a}$. This double shear mechanism is expected at the Type II (30°) intersection. From the half-twin state (c), ϵ -martensite (f) is produced by shuffling on the d-plane, γ_{90} is produced by $\mathbf{T}/2$ shear on the c-plane, and $\{1\ 0\ -1\ 2\}_\epsilon$ twinning is produced by shuffling on the c-plane. The latter two are both Type I intersection products. The double shear mechanism for the γ_{90} phase was first schematically drawn by Sleeswyk (Sleeswyk 1962) and

Fig. 11.11 Double shear mechanisms for various intersection products

experimental observation of the γ_{90} and the $\{1\ 0\ -1\ 2\}_\epsilon$ twin at Type I intersections was confirmed by EBSD and TEM in Zhang et al. 2011b.

11.5 Plasticity Mechanisms Under Cyclic Loading

Remy and Chalant reported a positive effect of the planar plasticity mechanisms on fatigue life (Chalant and Remy 1980). Suppression of the cross slip of dislocations owing to planarity is considered to decelerate the accumulation of fatigue damage. Figure 11.12 schematically shows the reversible back-and-forth movement of the partials associated with martensitic transformations, twinning, and dislocation glide (Sawaguchi et al. 2015). These mechanisms commonly involve the movement of a partial dislocation connected with an SF as an elementary step. All of these are expected to improve the fatigue life; however, the degree of improvement is different, probably due to different degrees of reversibility.

Recently, studies have extensively investigated the reversal of mechanical twinning and its positive effect on fatigue properties. Direct observations of the twinning and detwinning in TWIP steels under cyclic loading have been made using an in situ neutron diffraction technique (Xie et al. 2018) and by EBSD analysis on interrupted reverse loading (McCormack et al. 2018).

Fig. 11.12 Reversible movement of Shockley-partial dislocations associated with **a, b** mechanical $\gamma \rightarrow \epsilon$ martensitic transformation, **c, d** mechanical γ twinning, and **e, f** extended dislocation glide. (Reprinted with permission from Scripta Mater. 99 (2015) 49, Copyright 2014, Acta Materialia Inc. Published by Elsevier Ltd.)

Sawaguchi et al. reported a reversible change in the surface relief formed on the pre-polished surface of an Fe–28Mn–6Si–5Cr–0.5NbC SMA during tensile and subsequent compressive loadings (Sawaguchi et al. 2006). They suggested possible improvement of the fatigue resistance by the reversible deformation characteristics of the SMA. However, subsequent investigations revealed that the optimum chemical composition was different from that of the SMA.

Figure 11.13 shows the fatigue lives of Fe–30Mn–(6–x)Si–xAl alloys with different compositions ($x = 0–6$) at room temperature after strain-controlled tension–compression loading at a total strain amplitude of 0.01 (Nikulin et al. 2013). Note that the alloys are those discussed in Sect. 11.4.2. regarding their tensile behaviour. The Fe–30Mn–4Si–2Al TRIP steel exhibits the longest value of the fatigue life (N_f), whereas the Fe–30Mn–6Si SMA and the Fe–30Mn–3Si–3Al TWIP steels possess lower values of N_f than those of the former. SMAs can also exhibit improved N_f when the deformation temperature is increased. Figure 11.14 represents the fatigue lives of an Fe–28Mn–6Si–5Cr SMA after strain-controlled tension–compression loading at a total strain amplitude of 0.01 at temperatures ranging from 223 to 474 K.

The composition and temperature dependence of the fatigue life of high-Mn steels reached a maximum when they were deformed in the vicinity of the M_d temperature. The temperature range is characterised by the strain-induced nucleation of martensite,

Fig. 11.13 Fatigue lives of Fe-30Mn-(6-x)Si-xAl alloys ($x = 0$ to 6) for cyclic tension-compression loading at a strain amplitude of 0.01. (Reproduced with permission from Mater. Sci. Eng. A 587 (2013) 192, Copyright 2013, Elsevier B.V.)

as discussed in Sect. 11.4.1. The accumulated volume fraction of the ϵ -martensite increases slowly under cyclic loading, and the post-fatigue fractured microstructure shows a dual γ/ϵ phase. An example of such dual phase microstructure observed on the Fe-30Mn-4Si-2Al TRIP steel by SEM-EBSD is shown in Fig. 11.15b, while the single ϵ -phase microstructure on the Fe-30Mn-6Si SMA and the single γ -phase microstructure on the Fe-30Mn-3Si-3Al TWIP steels are shown in Fig. 11.15a, c, respectively.

A possible reason for the highest N_f value occurring at this temperature is the small difference in the Gibbs free energy between the γ - and ϵ -phases. The Shockley-partial shear occurs only in such a way that the atom goes through the saddle point between the two atoms on the lower layer, as shown in Fig. 11.4, while the movement in the opposite direction is strongly forbidden. After partial shear, however, reversal shear can easily occur under counter-directional loading. Owing to the unidirectionality, the Shockley-partial shear yields reversibility under cyclic loading. However, if the Gibbs energy of ϵ -martensite is sufficiently low, once ϵ -martensite is formed, it cannot reversely transform into austenite. That is, the thermodynamic condition of $\Delta G \lesssim 0$ is required for the reversible martensitic transformation. Such reversible twinning shear is also expected in the twinning-detwinning process. So far, however, the N_f values at temperatures above M_d are lower than those at temperatures below M_d . This is probably due to complicated dislocation interactions at relatively higher SFE_γ values with narrower dislocation extension widths.

A new application of high-Mn steels with improved fatigue life has been proposed for the seismic response control of architectural constructions. Steel seismic dampers have recently been used to protect buildings by elasto-plastic deformation hysteresis

Fig. 11.14 Fatigue life **a** and 0.2% proof stress **b** as a function of deformation temperature with respect to martensitic transformation temperatures (Tasaki et al. 2016). (Reprinted with permission from Mater. Trans. 57 (2016) 639, Copyright 2016, Japan Institute of Metals and Materials)

that transforms seismic energy into heat. A new seismic damping alloy, Fe–15Mn–10Cr–8Ni–4Si, was developed based on the design concept described in this article and is currently used in some large-scale buildings (Sawaguchi et al. 2016).

11.6 Concluding Remarks

An attempt is made to describe the plasticity mechanisms in high-Mn austenitic steels in a relatively comprehensive manner, considering the distortions of polyhedron models. The selection rule of the plasticity mechanisms, microstructural development, and their effects on the tensile and fatigue properties are discussed in terms of microstructural, thermodynamic, and crystallographic aspects. The tensile and fatigue properties of coarse-grained and interstitial-free Fe–Mn–Si–Al alloys

Fig. 11.15 EBSD-IPF maps **a**, **c**, and **e** and phase/boundary maps **b**, **d**, and **f** of the fatigue fractured specimens of **a**, **b** Fe-30Mn-6Si, **c**, **d** Fe-30Mn-4Si-2Al, and **e**, **f** Fe-30Mn-3Si-3Al. (Reprinted with permission from Mater. Sci. Eng. A 587 (2013) 192, Copyright 2013, Elsevier B.V.)

with different SFE_γ values are shown as representatives. The effects of grain refinement, precipitation, and interstitial elements are excluded here. Nevertheless, some microstructural events, such as crossing shear reactions at the intersection of the different ϵ -martensite variants and reversible back-and-forth movement of Shockley, which are highlighted in this article, are of importance in designing mechanical and fatigue properties of high-Mn steels.

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