# 3. HIGH MANGANESE MATERIALS

## Classification of chapter:

### 3. High manganese materials

#### 3.1. Basic types of high manganese materials

#### 3.2. Hadfield steel

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#### 3.3. TWIP steel

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#### 3.6. Applications of TWIP and TRIPLEX steels

## Summarization of chapters and questions

## Literature

## Time necessary for study: 430 minutes

## Cíl: After studying of this capture:

- you gain general view about basic manganese materials;
- you get information about core of high manganese materials formation and will understand the differences among individual steel types;
- you will be able to distinguish properties of individual high manganese types;
- you will acquaint with basic applications;
3. High manganese materials

3.1. Basic types of high manganese materials

High manganese materials have Mn content higher than 12 wt. %, which is substantially the level of Mn in the Hadfield steel, from which other types of high-manganese materials have been developed. The aim of the present process is to show physical and engineering properties of n TWIP and TRIPLEX steel in the context of Hadfield steel. High manganese materials can be categorized as follows:

1. Hadfield steel
2. TWIP alloy (twinning induced plasticity) – i.e. the type of the realized deformation process
3. TRIPLEX alloy (besides the fundamental Fe bases there are at least three elements)

3.2. Hadfield steel

3.2.1. Hadfield steel properties

Materials such as Hadfield steels contain about 12 wt. % of Mn and about 1 wt. % of C. Both of these elements, especially of Mn, significantly stabilize the austenitic matrix, i.e. the FCC structure. Under normal conditions or at low deformations, stabilizing of austenite matrix is ensured and its transformation to ε-martensite and α'-martensite is suppressed. Austenitic matrix has good plasticity, namely strength, ductility, and it is characterized by favourable corrosion response. It is not hard enough, thus robustly resistant, even when Mn and C have very good solubility in austenite, and through solid solution they are able to partially solidify it.

Hadfield steel is produced mostly by casting, so that in the cast state at a certain cooling rate of castings, precipitation of carbides may occur, therefore heat treatment, i.e. homogenizing the structure by dissolving annealing is necessary. This is realized by heating to a temperature of about 1,150 °C followed by cooling. The aim is to achieve uniformity in
the solid solution. The material of Hadfield steel type does not exhibit high strength, or hardness after casting and heat treatment, and it is therefore necessary to solidify the austenite matrix in deformation or transformation way in the areas of the most intense stress. This is done only in the surface and subsurface area, either by using the “explosion” (detonation) or by conventional heat treatment (quenching). In certain types of stresses, the material is heated and subsequently cooled. As a result of this process, martensitic structure of the material is formed on the surface, which comprises ultra-fine primary grains of up to a nano-size – Fig. 3.1. When the material is loaded, the non-transformed matrix is capable of absorbing large amounts of energy. However, the condition is not permanent and surface and subsurface zones become brittle due to gradual stress. This results in the formation of cracks in these areas, in which the austenite has been transformed to martensite and gradually, this hard but brittle martensite begins to flake off – Fig. 3.2.

![Fig. 3.1 Ultra fine martensitic layer with grains of nano-size](image)

![Fig. 3.2 Schematic demonstration of material abrasion with potential crack formation](image)

![Fig. 3.3 Perpendicular bands representing ε-martensite](image)
Fig. 3.4 $\alpha'$-martensite formation in place of bands intersection, i.e. intersection of $\varepsilon$-martensite plates (laths)

The finer is the microstructure of the initial austenite ($\gamma$), the slower and shallower is the progress of the actual flaking of the material. Defects that slightly extend into the depth of the matrix may be “rolled” by a certain manipulation (e.g. travels). If not, the material will further flake off in the transformed surface and subsurface area, and during the ongoing exploitation, the gradual transformation of austenite matrix into ever greater depths will also occur. The austenitic microstructure will gradually transform to $\varepsilon$-martensite, which is a transition stage of transformation of austenite to $\alpha'$-martensite. The process may proceed until the plasticity of the austenite matrix lasts, i.e. until the transformation of austenite to $\varepsilon$-martensite, or $\alpha'$-martensite is completed. This process is shown in Fig. 3. Continuation of the process is shown in Fig. 3.4. - austenitic grain during the transformation from $\varepsilon$-martensite to $\alpha'$-martensite arising from intersecting strips of $\varepsilon$-martensite plates (see Fig. 3.5 a, b). The overall process of transformation ($\gamma$ through $\varepsilon$-martensite to $\alpha'$-martensite) is constantly repeated to exhaustion of plasticity and it is a common phenomenon for the material. To extend the life, it is necessary to ensure thorough maintenance (abrasion of surface layers with defects in the already transformed matrix), thus avoiding possible further spread of fissures and flaking of the matrix, which functions as a source of undercuts (Fig. 3.2).

As mentioned above, this steel has good plasticity, toughness, ductility, and good anticorrosive properties. The yield stress is at least 390 MPa, the strength around 900 MPa, ductility ca. 40 %, and the hardness is at the level of 400 HB. It is able to resist abrasive wear,
it is more difficult to machine, and it is also resistant to impact. Due to the high content of manganese, austenitic matrix is paramagnetic. These properties are dependent on the chemical composition and heat treatment of surface / subsurface layers.

Fig. 3.5 Image of Hadfield´s steel. Deformation bands in austenitic grains a) appearance of transformed austenite into ε-martensite, resp. α´-martensite b)

### 3.2.2. Applications of Hadfield steel

It is applied in mechanical engineering, mining, extractive industry, steel industry, it is used in rail transport (railway and tramway points of switch) and the processing of products such as cement and clay. Steel can also be applied, for example, for presses for car wrecks or the jaw of crushers and excavator bucket teeth.

Fig. 3.6 Summary of material generations applied in automotive industry
3.3. TWIP steel

3.3.1. Conditions for TWIP steel existence

Its specific position is presented in Fig. 3.6. The fundamental chemical base corresponds to Fe-Mn-C, where Mn is around 22-25 wt. % and the content of C must be balanced in relation to the present Mn, which must maintain C in the solid solution, and austenite must not transform to ε-martensite as shown in Fig. 3.7. The given chemical constitution corresponds to a certain level of stacking fault energy - SFE (with TWIP steel, it is higher than 18 mJ.m⁻²).

![Diagram of phases stability in ternary Fr-Mn-C system (300K)](image)

**Fig. 3.7** Diagram of phases stability in ternary Fr-Mn-C system (300K)

At high strength level ranging from 1100 to 1150 MPa, the microstructure of Fe-Mn-C steel (TWIP) ensures high plastic feedback through twinned deformation in the stable austenitic (FCC) matrix. The basis of this behaviour is to meet the following physical and metallurgical conditions:

a) the final constitution of microstructure must be composed of 100% of FCC structure in the temperature range from -100 °C to 300 °C, which corresponds to the working temperatures of the effective technical use of this type of steel;

b) suppression of ε, or α' - martensite formation during cold working, or the application of deep tensile stress;

c) reaching of the optimum level (the full balance) of $R_p$, $R_m$ and uniform elongation in a tensile test at the normal temperature;

d) suppressing (preventing) of carbide particles formation during the individual technological procedures realized during the process of drawing;

e) maintaining of technological compatibility during the conventional process of continuous casting depending on hot deformation processes;
f) Stacking fault energy (SFE) must be higher than \(18 \text{ mJ.m}^{-2}\) (permanently FCC matrix – Fig. 3.8).

**Fig. 3.8**  Map of constant level of stacking faults energies (SFE) with marking of predominant mechanism in given area

TWIP steel is characterized by low SFE, which makes it possible to carry out multiple or sequential deformation processes. In addition, in response to the SFE level, martensitic transformation, i.e. the stress induced phase transformation (formation of hexagonal \(\varepsilon\)-martensite) may occur as well. This acts as a competitive mechanism for the twinning deformation mechanism. It is desirable to achieve deformation conditions in which the twinning process will be applied preferentially. Optimum conditions are achieved if so called twinned deformation intensity is sufficiently high. This corresponds to the state when the process of formation (development) of the volume fraction of twinning for a given level of applied deformation reaches the maximum. It is necessary to monitor and suppress the possible formation of carbides and thus ensure the effective use of the strength of the solid solution (matrix – with the content of C above 0.4 wt. %) thus preventing the carbide precipitation, which in turn, would lead to the impoverishment of the matrix with potential loss of chemical constitutional balance in terms of levels of stacking fault energy. This would lead to shifting of the realized processes towards the formation \(\varepsilon\)-martensite. Reduction of the C content is also reflected in a reduced level of SFE and thus the disintegration of the FCC matrix to \(\varepsilon\)-martensite is prioritized. The above is summarized in Fig. 3.8.

Processes of deformation twinning and induced formation of martensite are \(\varepsilon\)-deformation mechanisms, where twinning gives greater prerequisites for achieving optimum condition, namely in terms of the potential development of plasticity (achieving higher ductility). Mechanical twins are realized through partial dislocation slip \(a/6\{112\}\) on the
plane \{111\}. Crystallographic structure formed during twinning is a type of orientation comparable to the matrix. In case of \(\varepsilon\)-martensite formation, the same dislocation slip occurs as in the above case whereas the deformation is effected to every other plane. Under such conditions, the deformed area has a shape of a fine plate with HCP structure. Both mentioned deformation mechanisms are firmly connected to SFE that control the energy needed to create these defects. The experimental results show that the formation of \(\varepsilon\)-martensite may be substituted by a mechanical twinning if SFE is sufficiently low. SFE increases with the increasing content of C and Mn in the steel.

The analysis of the SFE level in the ternary system Fe-Mn-C at 300K shows that there is a limit SFE level, deciding which of the above deformation mechanisms can be implemented as a priority. This threshold for TWIP corresponds to ca. 18 mJ.m\(^{-2}\). If the SFE is below this level, the emergence of \(\varepsilon\)-martensite and other products can be observed. Therefore, the chemical composition determines the SFE level, which leads to achieving an optimized TWIP effect during deformation process realized at normal temperature. For optimized material TWIP SFE is most often at the level from 25 to 30 mJ.m\(^{-2}\).

![Tensile strength – strain plotting with marked contraction (neck)](image)

**Fig. 3.9a**  Tensile strength – strain plotting with marked contraction (neck)

In the case of TWIP steel, an important characteristic is the development of uniform ductility at normal temperature. It is a condition corresponding to the high deformation hardening -\(n\)- that exceeds the level of 0.4. Regarding the strength, the maximum value is achieved at low temperatures, but at lower values of uniform ductility. The opposite case of the mechanical response of the alloy is demonstrated in tensile tests at a temperature higher than 500K. Here, the elongation at the tensile test fracture is controlled by the intensity hardening (\(n\)) corresponding to the fulfilment of the criterion of equality between at \(\varepsilon\)rov. (\(n=\varepsilon\)). Strength values of the concerned alloy depend on both strain hardening intensity, and
on thermally activated dislocation motion as follows logically from the detected strength increasing with temperature decrease. Summary results are shown in Fig. 3.9 and 3.9b.

![Graph showing tensile strength vs. temperature](image)

**Fig. 3.9** Results of tensile tests under different temperatures (Fe-22Mn-0.6C), including deformation mechanisms

In the Fe-Mn-C (TWIP1100) alloy, e.g. the following values are obtained on average at the normal temperature: \( YS = 490 \text{ MPa} \), \( TS = 1100 \text{ MPa} \), \( \text{El}_{\text{tot.}} = 54-55 \% \), while in the TRIP (800) steel type, the following values are usually achieved: \( YS = 470 \text{ MPa} \), \( ST = 820 \text{ MPa} \), \( \text{El}_{\text{tot.}} = 7 \% \) and \( n (4\%) \approx 0.22 \). The characteristic microstructure of this alloy (twinning) is shown in Fig. 3.10.

![Micrograph showing twins in steel](image)

**Fig. 3.10** Twins detected in deformed ternary steel

The above mechanical and metallurgical facts may be combined with material engineering behaviour of the evaluated ternary alloy. At temperatures above 500K, SFE is
increased so much that a mechanism based on the matrix twinning during deformation is suppressed. In this case, you can only record the development of planar dislocation slip, including the development of hardening and the associated restrictions on the implementation of uniform ductility, including a degradation of values and ultimate strength limit. The aforementioned characteristics, namely decrease in the extent of uniform ductility corresponds to the “occurrence” of formation of localized deformation - formation of the neck – in the tensile test, which is accompanied by a reduction in total ductility. The most favourable relation between strength and ductility can be recorded in the event if twinning occurs during the intensive deformation, as it is at normal temperature. Under the given conditions, however, the dynamic aging mechanism may apply. This notion is supported by the findings of ongoing instabilities of the plastic deformation parallel to deformations, which are detected in Fig. 3.9a.

At lower test temperatures, the level SFE decreases, so stress-induced phase transformation associated with the emergence ε-martensite, i.e. a mechanism that replaces mechanical twinning as a “substitute” deformation process is potentially possible. The resulting ε-martensite boards also act as barriers to the movement of dislocations, and thus even higher intensity of hardening can be achieved. On the other hand, slip dislocation in this interval is more difficult, and therefore the efficiency of hardening is lower compared to the level of that characteristic at a normal temperature, leading to a reduction in the overall level of ductility.

Twinned microstructures were studied at different magnifications. TEM leads to the realization that twinning occurs in connection with the formation of micro twins connected in the form of layers, which act as strong barriers to dislocation motion. The thickness of the individual micro twins is only a few tens of nanometres, while ordering (layering) reaches a thickness of several micrometres. Tensile tests lead to the activation of two intersecting twinning systems which are activated sequentially. In this context, it should be added that the secondary twinned system, which is detected after deformation of about 15 % (in the considered set Fe-22Mn-0.6C) effectively contributes to deformation hardening by reducing the mean free path of mobile dislocations.

3.3.2. **Selected parameters of TWIP steels**

**High C content** in that type high manganese material could lead to the formation of carbides, M₃C type, M₂₃C₆, or to formation of pearlite constituted on the basis of cementite.
and ferrite. The above precipitation processes, however, can have adversely impact for the following reasons.

a) C bound in carbide precipitates lowers its concentration in the solid solution – it leads to SFE reduction;

b) cementite precipitated at grain boundaries adversely affects the achieved toughness of the material.

As it is evident from past experience, $M_2\text{C}_6$ precipitation was not detected in any of the examined cases. The resulting carbides are of cementitic (orthorombic) type $(\text{FeMn})_3\text{C}$ by up to about 30 wt. % Mn depending on the applied treatment regime. Parameters of carbide precipitation are shown in Fig. 3.11., where two types of carbide precipitation are plotted. Places where carbides were not formed were marked with empty squares. Dark squares indicate the occurrence of cementite, or perlite.

![Graph](image)

**Fig. 3.11**  Dependence time on temperature for case of carbide formation in Fe-Mn-C system

An integral part of the evaluation of TWIP material is its **recrystallization** response after graded hot deformation (30 and 70%). This material exhibits very favourable behaviour during recrystallization, which is manifested by emergence of homogeneous fine-grained microstructure (evenly spaced), with no sign of texture. Following the degree of deformation and annealing conditions, grains in the size of 1-5 $\mu$m were observed after recrystallization during standard continuous annealing. The kinetics of the recrystallization process is relatively high, which is due to the existence of a high density of nucleation sites. This is e.g. the case of the impact of the phase interface between the twins and grain boundaries or twins
belonging to two different twin systems. The recrystallized volume fraction $X_V$ can be expressed by anizothermal Johnson-Mehl-Avrami concept:

where $n$ is the Avrami coefficient, $B_{CR}$ is pre-exponential factor, which is a function of the chosen deformation during cold rolling, $Q$ represents the activation energy for recrystallization. When monitoring material commonly referred to as FeMn TWIP 1000 (or 1100), it was found that $n = 0.81$, $Q = 309.5 \text{ kJ mol}^{-1}$, $B_{CR} = 4.7 \times 10^{16}$ at 50% cold deformation of the matrix.

Special attention must be paid to the response of the material at the impact stress (crash performance). In this connection, two load variants were evaluated (pressure, or bend) at the impact of 55 km-hod$^{-1}$ (pressure), or 29 km-hod$^{-1}$ (bending. In the case of the pressure (impact) load, TWIP alloy exhibits a significant increase in resistance in comparison with conventional deep drawing steels. The ratio of “tolerated” force for the compared material variants is 77 kN/41 kN for the benefit of high manganese alloy. When testing in impact bending, the ratio is 38 kN/24 kN. In the case of hardening FeMnC TWIP material by 8% (variant 1), or 17% (variant 2), it is possible to achieve a further increase in resistance (the technical level of the material). At the compressive stress, the “tolerated” force increases from the original level of 77 kN (without deformation) to 85 kN (variant 1) and 105 kN (variant 2). The bending gave the following values: 38 kN (no deformation) and 44 kN (8% deformation) and 58 kN (14% deformation). In this context it should be noted that all the above data were determined for the plate material FeMnC TWIP 1.5 mm.

![Fatigue resistance of FeMn TWIP steel vs HSLA 320 and DP600 behaviour](image)

**Fig. 3.12** Fatigue resistance of FeMn TWIP steel vs HSLA 320 and DP600 behaviour
Significant influence is ascribed to its **fatigue properties**. As found in the load in ratio - 1, TWIP material generally exhibits a very favourable fatigue response, e.g. in comparison with steel TRIP800 - see Fig. 3.12.

### 3.4. TRIPLEX steel

#### 3.4.1. Conditions for TRIPLEX existence

It is a system characterized by both a high Mn content (about 26-30%) and extremely high aluminium alloying (about 8-12%). The average content of C is 1%. Further, the alloy must meet the following parameters.

- a) microstructure is mostly formed by (austenitic) matrix
- b) ferrite (about 10 %)
- c) k-carbides (about 6 to 8%)
- d) stacking fault energy ranges approximately from 80 to 140 mJ.m\(^{-2}\)

K-carbides precipitate in TRIPLEX matrix during annealing at high temperatures. They are separated via spinodal disintegration upon aging at temperatures between 500-600 °C.

![Graph showing density and relatively reduction in density vs. Al content](image)

**Fig. 3.13** Density and relatively reduction in density vs. Al content in Fe-xMn-yAl-zC system with increasing Mn content
3.4.2. Selected parameters of TRIPLEX steel

Decrease in specific weight of the concerned quaternary alloy with higher Al content. Fig. 3.13 shows dependence of specific weight of the Fe-xMn-yAl-zC system on Al content, with graded Mn content (taking into consideration the contents of 14 to 28%). The lower linear dependence represents the total decrease in specific weight with increasing aluminium concentration. Upper dependence shows a reduced density resulting from FCC type base lattice dilatation.

Linear dependence of specific weight on the content of Al in the evaluated alloy (Fig. 3.13) shows that it is possible to achieve a minimum level of about 6.5 g.cm\(^{-3}\) (at the highest considered Al content, up to about 12.5%). This condition is equivalent to the percentage decrease in specific weight \(([\Delta \rho/\rho]) 100 \approx 17\%\). The decrease in weight in the case (as described above) of the coexistence of the FCC and BCC lattices (phases) in the base solid solution of Fe matrix (Mn, Al) is due to a lower average molecular weight of alloying elements of studied alloys and significant decrease in lattice density of the individual cells. This results from an effective extension of the lattice, which corresponds to a larger atomic radius in the matrix of dissolved Al carbon \((r_{\text{Al}} = 0.147\text{nm})\) compared with a smaller radius of Fe \((r_{\text{Fe}} = 0.126\text{nm})\). It should also be noted that Mn atoms in solid solution contribute to a reduction in weight due to the larger atomic radius of Mn \((r_{\text{Mn}} = 0.134\text{nm})\). This also means that alloying of Mn leads to a complementary increase in lattice parameter for FCC matrix unit cells. Upper linear relationship in Fig. 3.13 represents a partial reduction in specific weight value \((\Delta \rho/\rho) 100 = 10\%\) if the aluminium content in the alloy is 11.5%. This phenomenon is caused only by lattice dilation.

The microstructure of analysed alloy Fe-26Mn-11Al-1.15C (TRIPLEX) consists of a matrix with the FCC crystal lattice characterized by the occurrence of the annealing twins and about 10% of the ferritic phase. The structure is also formed by the \(\gamma'\)-type k-carbides, showing that their crystalline configuration corresponds to the L1\(_2\) structure (this is a variant of an ordered FCC lattice). A cell of K-carbide defined by chemical constitution \((\text{Fe, Mn})_3\text{AlC}\) has dissolved carbon atoms arranged in its centre - see Fig. 3.14. Medium lattice parameter \(a_0 = 0.3837\text{nm}\), while its size depends primarily on the content of Al in k-carbides. In terms of the stability of the structure of the FCC type TRIPLEX alloy, an important role is played by the thermodynamic characteristics of the base matrix, namely in terms of the relationship between the basic FCC lattice and the martensitic phase transformation product (\(\varepsilon\)-martensite formation). The previously performed analysis showed that the considered
phase change corresponds to a positive value of free enthalpy $\Delta G_{\text{FCC} \rightarrow \varepsilon} = 1755 \text{J.mol}^{-1}$, indicating a relatively high stability of the FCC phase due to its disintegration into $\varepsilon$-HCP martensite. SFE level compared with the ternary system Fe-Mn-C (TWIP) is practically much higher (see the above-mentioned SFE figure = 110 mJ.m$^{-2}$).

![Diagram of the k-karbide unit cell](image)

**Fig. 3.14** Unit cell of k-karbide $(\text{FeMn})_3\text{AlC}$ with positions of C, Al and $(\text{Fe},\text{Mn})$ atoms in arranged lattice

The matrix with this SFE level will not be deformed by twinning mechanism (twinned deformation can be expected if stacking fault energy level is less than 30 to 40 mJ.m$^{-2}$). The analysis of the dependence of the actual stress and deformation (or the intensity of strain hardening, which is defined by the ratio $d\sigma_{sk}/d\varepsilon_{sk}$) showed that the strain hardening exponent $n$ achieved in TRIPLEX alloy is usually higher than similar value found in conventional austenitic steels, e.g. type 304 (compared to those found at the normal temperature). In this case, the ratio of the strain hardening exponent values $n$ is considered up to the level of 0.58 vs. 0.45. At higher temperatures (about 200 °C), even a significant increase of the initial strain hardening exponent $n$ to the level of 0.70 is achieved. Increased hardening suggests that it is likely to be superposition firming effect, which in its consequences leads to increasing restrictions of constriction (neck) formation in the tensile test.

In this case, two hardening mechanisms will apply. One of them is the initial formation of k-carbide precipitates (characterized as a preliminary stage) induced by voltage, in conjunction with the initial degradation of the FCC matrix (with a high content of Mn and Al). An additional hardening mechanism is an interaction between carbon atoms with edge dislocations interstitially dissolved in the matrix by, which is accompanied by the formation of significant instabilities in the stress-strain relation. The results of X-ray analysis performed on the material of the tensile tests carried out at a temperature of 250 °C (solid solution with a
high content of Al) showed the complexity of the microstructure which corresponds to the particular state of the matrix from – Fig. 3.15.

![Fig. 3.15 X-ray diffraction of austenite, ferrite and k-carbide (γ') – arranged crystallographic FCC structure](image)

In addition to the basic structure of the FCC, incidence of ferritic reflections corresponding to partially ordered phase (super lattice of the $\text{D}_0^3$ type) and $\gamma$ reflections with the $\text{L}_{12}$ type structure, which can be assigned to k-carbide, were detected. TEM analysis showed high occurrence of uniformly arranged shear bands on a plane $\{111\}$ in the structure. This characteristic can be attributed to the action of a homogeneous shear deformation, which leads to an overall plastic ductility, abbreviated as SIP-effect (shear band induced plasticity). This mechanism can be classified in a separate category, unlike some ideas that were based on the development of twinning, or induced martensitic phase transformation (see positive free enthalpy value of the conversion of FCC to $\varepsilon$-martensite and SFE greater than 100 mJ.m$^{-2}$). As is evident from the past experience, the distribution of a coherent precipitation of k-carbide of nano-dimensions has a significant influence on the formation of a uniform distribution of shear bands and hardening of the TRIPLEX matrix. The coexistence of three phases in the quaternary alloy influenced the choice of the working name of this type of material as the TRIPLEX alloy.

3.5. **Mutual comparison of properties of high manganese steels**

Physical mechanical and technical response to stress variations is not diametrically distinguishable. Despite this fact, the microstructure constitution of these alloys is
fundamentally different. In fact, the presented data show that the modification of the chemical constitution (even in case of ensuring the FCC microstructure) can lead different deformation characteristics. This either results in the preferential existence of twinned deformation or to the formation of stress induced martensite.

The decisive criterion of the discussed different deformation behaviour is the SFE level, whose size can be reliably ensured by the selection of the chemical composition of the solid solution of the alloy matrix. On the other hand, the TRIPLEX type alloys are due to their chemical constitution (in addition to the high content of Mn and increased Al additive to 10 to 12%) characterized by a specific form of the deformation response, as this corresponds to an increased level of stacking fault energy up to about 110 mJ.m^{-2}. This means that in this case, deformation process by twinning or via induced martensitic phase transformation cannot be expected. In the case of the quaternary Fe-Mn-Al-C system, the core deformation process is the realization of the mechanism associated with the formation of homogeneous shear bands. This leads to achieving enhanced plastic behaviour based on the dislocation slip that is associated with the formation of homogeneous shear bands.

The decisive parameter is the formation of a uniform arrangement of these bands with the auxiliary action of the uniform distribution of nano-carbides, so called k-carbides of the L1_2 type structure (the size of 20 to 30nm), coherent with the FCC matrix. Conversely, in the case of TWIP alloy, depending on the level of stacking fault energy, either intensive twinned deformation (in case when the stacking fault energy is higher than mJ.m^{-2} – up to the size 30, or max. 40 mJ.m^{-2}) or the deformation associated with the formation of the HCP martensitic structure occurs. This applies if the SFE level is lower than the above mentioned 18 mJ.m^{-2} however it is usually the case when SFE ranges from 12 to 15 mJ.m^{-2}. Twinned deformation is accompanied by the formation of micro-twins arranged in layers, which is an obstacle for dislocation movement. Micro-twins are about the size of several tens of nanometres, while the layered structure is about several tens μm.

An important role with respect to hardening at the current high plasticity is played by the emergence of the system comprising two types of configuration twins. The second type of the emerging twin arrangement that is only induced at higher deformation (more than 15%) effectively participates in the process of strain hardening. Regarding the hardening, the formation of ε-martensite is a part hardening process. Given that dislocation slip process is more difficult, hardening becomes more demanding. It is also important to comply with the phase stability in the FCC matrix within the temperature interval of alloys used.
experimentally (in practice). This concerns both the FCC lattice as well as suppression of uncontrolled formation of carbide phases. For example, with TWIP alloy, it is the range of the given SFE level, or possibly the constitution type with the addition of Al, which limits the realization of induced martensitic phase transformation.

As regards the response to various types of stress, the level of strength obtained with both variants of TWIP alloys (TWIP, TRIPLEX) is about **1100 to 1200 MPa** at the favourable plastic response. Under optimum conditions of the deformation process with TWIP and TRIPLEX at the normal temperature, **ductility is higher than 50 %**. There are cases where ductility was to 70%. The values of strain hardening exponent - n - correspond to this level with TRIPLEX to about **0.55 to 0.60**. It is interesting that this value is on average about 40% higher than that obtained at the stress-induced phase transformation from the austenitic CrNi steel (type 304).

Both these variants of high-strength alloys **strongly absorb energy**, e.g. when simulating conditions implemented in a car crash. The absorbed energy is in the case of alloy TWIP and TRIPLEX substantially higher (at TWIP alloys even slightly higher than those obtained with TRIPLEX alloy). In the case of high manganese alloy is the absorbed energy at a level from 0.45 to 0.50 J.mm\(^{-3}\), while in the TRIP steel, this value is 0.23 J.mm\(^{-3}\), i.e. half in comparison with TWIP and TRIPLEX alloys, as shown also in Fig. 3.16. The discussed behaviour practically corresponds to the ratio of other variables as well, such as plastic response and strain hardening achieved with the discussed types of materials.

![Diagram of specific energy absorption (E\(^{\text{spec.}}\)) for different material types, TWIP and TRIPLEX under crash simulation including](image-url)

**Fig. 3.16** Diagram of specific energy absorption (E\(^{\text{spec.}}\)) for different material types, TWIP and TRIPLEX under crash simulation including
Compared to conventional mechanisms of plastic deformation, the TWIP deformation mechanism represents a preferred process. It is both a plastic deformation associated with dislocation slip, and a deformation associated with the development of plastic response during the induced martensitic transformation in a locally distributed retained austenite (TRIP effect). Both of these mechanisms are less efficient than the process of twinning, which is a continuous hardening mechanism. The dislocation slip does not lead to sufficient intensity of hardening that would lead to a high strength as it is in the case of twinning. Regarding the TRIP mechanism, this does not lead to particularly high ductility. TRIP effect is exhausted when the entire volume fraction of the retained austenite transforms into martensite.

3.6. Applications of TWIP and TRIPLEX steels

For both observed types of high-manganese alloys, a highly favourable combination and balance of properties such as formability, strength level, ductility and strain hardening parameters are achieved, which in summary allows a weight reduction of vehicles while improving the level of resistance against the effects of automobile accidents. With TRIPLEX steel, promising applications of cryogenic technology (equipment for storing and transporting liquefied gases) and the weight-saving design in a broad sense complement this overall evaluation. TRIPLEX steel also gives a possibility of its use as a replacement of austenitic Cr-Ni steels under conditions where corrosion resistance is not critical, such as the high-strength FCC non-ferromagnetic material (rings of rotating electric machines – generators). The above-mentioned facts suggest promising benefits resulting from the application of high-manganese alloys in engineering practice in the implementation of cutting-edge technical and technological design requirements - see Fig. 3.17.

Fig. 3.17 Application of high manganese steels in automotive industry
Summarization of chapter terms

In the end of this chapter main terms that you should master are repeated
Hadfield’s steel, TWIP steel, TRIPLEX steel, stacking fault energy (SFE), twinning, shear induced plasticity, K-carbides, specific energy of absorption

Questions

- Could you divide high manganese steels?
- What is typical for Hadfield’s steel?
- Do you know chemical basis of Hadfield’s steel?
- How can be treated Hadfield’s steel?
- Do you know basic chemical composition and properties of TWIP steel?
- What the deformation process is realized in TWIP steel?
- Do you know level of stacking fault energy for TWIP steel existence?
- Which elements form TRIPLEX steel?
- Do you know structure of TRIPLEX?
- Do you know level of stacking fault energy for TRIPLEX steel existence?
- Could you explain the SIP effect?
- Explain impact of K-carbides on TRIPLEX steel properties.
- Do you know differences between TWIP and TRIPLEX?
- Describe possible applications for Hadfield’s steel, TWIP and TRIPLEX.

Literature: