Localized Deformation due to Portevin–LeChatelier Effect in 18Mn–0.6C TWIP Austenitic Steel

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The thermo-mechanical properties of low stacking fault energy austenitic Fe18Mn0.6C steel exhibiting twinning-induced plasticity were investigated during uniaxial tensile deformation using infrared thermography. Over a wide strain range, the plastic deformation was by the movement of very few well-defined localized deformation bands. The formation and propagation of Portevin–LeChatelier (PLC) bands lead to type A and type B serrated stress–strain curves, exhibiting a negative strain rate sensitivity. The PLC band properties were analyzed in detail: strain, strain rate and mobile dislocation density within the bands were determined. The microstructures of the un-deformed and deformed Fe18Mn0.6C TWIP steel were studied by transmission electron microscopy. The possible dynamic strain aging processes causing the localized deformation are reviewed.

KEY WORDS: TWIP steel; thermo-mechanical properties; dynamic strain aging.

1. Introduction

Improved safety standards, reduced automotive body-in-white weight, and manufacturing processes requiring a superior formability, have led to a strong interest in advanced high strength steel and “super tough”, high manganese steel characterized by TWinning-Induced Plasticity (TWIP). The TWIP effect is believed to cause the observed high flow stress (600–1 100 MPa) and exceptional elongation (60–95%).^{1–3)} Extensive research has already been reported on high Mn TWIP steels with slightly different compositions, but very similar properties: Fe25Mn3Al3Si,¹⁾ Fe22Mn0.6C,^{2,3)} Fe27Mn0.02C,⁴⁾ Fe30Mn3Al3Si,⁵⁾ Fe12Mn1.1C,⁶⁾ Fe12Mn1.2C,⁷⁾ Fe13Mn1.2C,⁸⁾ Fe32Mn12Cr0.4C⁹⁾ and Fe23Mn2Si2Al.¹⁰⁾

Deformation twinning in low Stacking Fault Energy (SFE) austenitic steels leads to high strain hardening and higher ductility, and it is well accepted that the mechanical properties of TWIP steel are mainly due to a combination of deformation twinning and dislocation motion in conditions of pronounced planar glide. In low SFE austenitic steels, the increased partial dislocation separation results in the ease of twin nucleation. During deformation, the grains are progressively subdivided by the twinning process, and the internal twin boundaries increase the strain hardening. Although the actual twinning strain is limited and the twin formation itself may actually cause softening, the twin boundaries reduce the dislocation slip distances progressively, and promote dislocation accumulation and storage, much as grain boundaries.^{2–4)} This dynamic Hall–Petch effect may not be the only cause for the observed strain hardening of TWIP steel. In fact, the mechanism leading to a

high strain hardening in TWIP steel is still a matter of debate. Research on Fe12Mn1.2C steel has shown that the pronounced strain hardening may not necessarily be caused by twinning only, as additions of Al and high temperature testing, both of which increase the stacking fault energy and thereby suppress deformation-induced twinning, do not influence the strain hardening much.⁶⁾

Alternative explanations stress the fact that a pronounced planar glide can also lead to a high strain hardening. In high Mn TWIP steel containing a considerable concentration of interstitial C, the strong attractive interaction between C and Mn leads to local ordering. This local order is a non-random distribution of interstitial C atoms in which there is a very high probability that a C atom occupies an octahedral interstitial position and forms an octahedral cluster for which the number of Mn atoms on the six nearest-neighbor positions is higher than the site occupancy expected on basis of the atomic Mn concentration.¹¹⁾ This clustering leads to a higher lattice resistance to dislocation glide, as the passage of a partial dislocation will in general change the local position of both substitutional and interstitial atoms. During the deformation of low SFE alloys, the passage of the leading partial of a dislocation results in a disordering of the C–Mn clusters. The trailing partial dislocation restores the f.c.c. lattice but not the original short range order. This disordering process results in a gradual reduction of the stress required to move each of the following dislocations on the same glide plane. This will tend to confine slip to a single slip plane, and the more pronounced planar glide will result in a higher strain hardening. Planar glide connected to solid solution strengthening and short range ordering has been discussed by Gerold *et al.* for N-al-

loyed steel.¹²⁾

In TWIP steel, twinning may interact with the ordering, as the homogeneous twinning shear also alters the degree of short range order as it results in a different pair correlation and interstitials in different positions. This mechanism is referred to as pseudo-twinning.^{13–15)} It leads to a meta-stable structure which may transform to a true twin structure as a result of thermal activation.

The flow curve of high Mn TWIP steels containing interstitial carbon are often characterized by serrations, the Portevin–LeChatelier (PLC) phenomenon, and a negative strain rate sensitivity. Both result from a microscopic Dynamic Strain Aging (DSA) process. DSA results in an increase in flow stress and strain hardening, but a decrease in post-uniform elongation and a reduction of area at fracture. When both DSA and a negative strain rate sensitivity occur simultaneously, Portevin–LeChatelier (PLC) bands will be observed, and the stress–strain curve will have characteristic serrations.¹⁶⁾

DSA occurs when an aging process, related to solute atoms, is fast enough to occur during deformation, *i.e.* when the dislocation velocity is similar to the solute mobility, as dislocations move by means of thermally activated jumps, with a characteristic waiting time between two jumps. As pointed out by Cuddy *et al.*¹⁷⁾ no long range diffusion is required and the solutes involved in the aging process may actually remain immobile in the lattice. Dastur *et al.*⁶⁾ have suggested C–Mn pairs or point defect clusters which reorient themselves in the stress field at the core of the dislocations are able to pin the dislocation strongly. It may actually be the C–Mn octahedral clusters rather than C–Mn pairs, and the resulting short range ordering effect, which causes dislocation pinning.

The DSA phenomenon is not limited to high Mn TWIP steel, as DSA has also been observed in 316L austenitic stainless steel for which it was shown that plastic deformation changed from wavy slip to a pronounced planar slip during DSA.¹⁵⁾ Kibey *et al.*⁸⁾ have pointed out that both C and N affect the dislocation structure and the deformation mechanism. Whereas N decreases the SFE, C increases the SFE. Activation of twinning in steels alloyed with N is due to the lowering of the SFE, resulting in an increased ease of twin nucleation. In low N stainless steels, twinning is suppressed, and planar dislocation glide is preferred. Note that the SFE is both composition and temperature dependent, with increasing temperature leading to an increased SFE.³⁾

DSA is usually not observed immediately after the start of straining, and a critical strain ϵ_c must first be achieved before the serrated flow is observed. This is believed to be due to the fact that sufficient vacancies must first be created to enhance the solute diffusion to dislocations. Serrated flow curves may not always be observed, as ϵ_c is strain rate dependent: as the strain rate increases, ϵ_c is increased. In certain cases however the serrated flow curve were clearly observed in published stress–strain curves but the fact was not mentioned or discussed by the authors, although it is of prime importance in understanding the very specific mechanical properties of high Mn-alloyed steels.

Serrated flow curves were observed by Allain³⁾ who reported on the properties of low SFE Fe22Mn0.6C TWIP steel with an ultimate tensile stress of 1 400 MPa and a true

uniform elongation of ~ 0.5 . Allain *et al.*²⁾ have published flow curves for Fe22Mn0.6C steel (SFE=12–35 mJ/m²) clearly showing the serrated flow of DSA, but they did not discuss the implications of this feature in their paper. Similar observations were made by Karaman *et al.*⁷⁾ for Fe12Mn1.2C steels (SFE 23 mJ/m²) for which twinning produces a higher strain-hardening than the one observed in austenitic steels not experiencing twinning. Fe12Mn1.2C single crystals, oriented along [001] and [111] for multiple slip, had serrated flow curves when deformed in compression. Hong *et al.*¹⁸⁾ reported on the Fe32Mn12Cr0.4C TWIP steel in the -150°C to $+150^\circ\text{C}$ range. No DSA and no negative SRS were observed in this case. The observations of Grassel *et al.* for Fe25Mn3Si3Al TWIP steel¹⁾ were similar. In their case, this may be due to the very low interstitial C content (300 ppm). Vercammen⁵⁾ studied Fe30Mn3Al3Si TWIP steel with a tensile stress of 975 MPa and a true total elongation of 0.41. He did not discuss the serrations present on the measured flow curves, but he reported that no local necking occurred prior to fracture, a strong indication for low or negative strain rate sensitivity. **Table 1** reviews the observation of negative strain rate sensitivity and serrated stress–strain curves in TWIP steel in the recent literature on high Mn steel.

It is clear that there are no detailed studies on the mechanical properties of Fe18Mn0.6C TWIP steel and studies of the TWIP mechanism in other high Mn steels have only become available recently. These studies show that there are still many questions related to the mechanism of plastic deformation in these steels. In particular, the mechanism for the very high strain hardening is still unclear. The aim of the present contribution was to report on recent investigations focusing on the observation of a pronounced DSA over a wide strain range in Fe18Mn0.6C. This phenomenon has been observed previously, but it has not been given the

Table 1. Overview of the stress–strain curve characteristics of high Mn steels.

Alloy	Serrated stress-strain curve	Negative strain rate sensitivity	Dynamic strain aging and localized deformation bands	Ref.
Fe18Mn0.6C	Yes	Yes	Yes	*
Fe22Mn0.6C	Yes	Yes	Yes	[2,3]
Fe12Mn1.1C	Yes	Yes	Yes	[6]
Fe12Mn1.2C	Yes	Yes	Yes	[7,11]
Fe13Mn1.2C	Yes	Yes	Yes	[8]
Fe25Mn3Si3Al	No	No	No	[1]
Fe27Mn0.02C	No	No	No	[4]
Fe30Mn3Al3Si	No	No	No	[5]
Fe23Mn2Si2Al**	No	No	No	[10]
Fe32Mn12Cr0.4C	No	No	No	[9]

*: Present work

** : Contains Nb micro-alloying additions

necessary attention, in particular in models that have been published recently on the TWIP mechanism. There is currently no theoretical model which takes this mechanism into account. The present contribution focused on thermo-mechanical properties in order to understand the complex interaction between twinning and dislocation glide in TWIP steel during mechanical loading and analyze the serrated stress-strain behavior of Fe18Mn0.6C TWIP steel. In addition, the results show the very important localization of the heating of the material during deformation. Current models of the mechanical behavior of TWIP steel do not take thermal effects into account. Fully austenitic Fe18Mn0.6C TWIP steels were tested by means of tensile tests combined with *in-situ* high sensitivity infrared thermography which gave the temperature distribution of the specimen during the whole test. The flow curve of 18Mn0.6C TWIP steel had a very high strain hardening and plastic deformation was entirely by localized PLC band propagation during uni-axial tension.

2. Experimental

The experimental approach focused on capturing the dynamic nature of the deformation process of Fe18Mn0.6C TWIP steel. Cold rolled and annealed Fe18Mn0.6C TWIP sheet steel was cut to flat ASTM A376 standard specimen for tensile tests. The surfaces of specimens were carbon-sprayed before testing to get a high and homogeneous thermal emissivity. The tensile tests were carried using a screw driven ZWICK Z100 machine with a crosshead displacement speed of 25 mm/min. A mechanical strain gauge with a 50 mm gauge length was used to measure the specimen strain. A CEDIP Silver 420M camera was used to monitor to specimen temperature by high resolution infrared (IR) thermography. The camera registered the distribution of infrared radiation from the specimen surface and created digital thermographic images of the samples with a 100 Hz frequency. The temperature space resolution of the camera was 3–5 μm and the temperature sensitivity was 20 mK. The infrared camera was placed in front of the tensile specimen to acquire the temperature distribution across the specimen and measured the localized heating associated with PLC deformation bands. The data acquisition system allowed for simultaneous capture of the flow stress curve and thermographs. In addition, the microstructures of Fe18Mn0.6C TWIP steel before and after tensile testing were studied by conventional transmission electron microscopy (CTEM) using a PHILIPS Tecnai operated at 200 keV.

3. Results

3.1. Stress-Strain Curves

Figure 1 shows a typical room temperature stress-strain curve of the Fe18Mn0.6C TWIP steel. For true strains less than 0.45, step-like discontinuities in stress were observed, which were separated by plateau regions in which the deformation proceeded with a small increase of the stress. At a true strain of 0.45 the stress-strain curve shows random serrations. This behavior is due to the initiation and propagation of localized PLC bands. **Figure 2** illustrates the

Fig. 1. (a) Stress-strain curve of Fe0.6C18Mn TWIP steel. (b) Enlargement of the first boxed segment of the stress-strain curve exhibiting type A serrations, characterized by a steep rise in stress alternating with plateau-like features. (c) Enlargement of a typical type A serration. The steep rise corresponds to the initiation of a PLC band outside the strain gauge measuring range. The plateau segment corresponds to the passage of the PLC band in the strain gauge measuring range. The increase in stress in the plateau segment, suggests strain hardening in the region outside the PLC band. (d) Enlargement of the type B serrations, characterized by an unstable serration pattern.

time-dependence of the stress, strain and the temperature at the center of the specimen which was recorded by IR thermography. The stress showed a discontinuity each time a single deformation band was initiated at one end of the specimen. The time-dependence of the strain shows that there was only a very limited deformation when the propagating deformation band was outside the measuring range of the strain gauge.

During the plastic deformation, the temperature at the specimen center was increased each time the PLC band passed the measuring point. In Fig. 2, it can be seen that the temperature increase was $\sim 5^\circ\text{C}$ for the first deformation band and increased to $\sim 20^\circ\text{C}$ for the last PLC band. In the time between the passages of two successive PLC bands, the temperature decreased slightly. The final temperature prior to fracture is typically about 110°C .

It is therefore clear that in the stress-strain curves of Fig. 1, the initiation of a single deformation band caused the step-like increase in stress, while the propagation of the PLC band in the measuring gauge range produced the large plateau-like strain increments at more or less constant stress. In addition, the deformation was associated with a pronounced local heating caused by the PLC bands. The successive deformation bands tended to be nucleated at the same end of the specimen. A possible explanation has been

Fig. 2. Stress (a), strain (b), and temperature at the center of the sample (c) as a function of deformation time. The initiation of PLC bands causes an unstable stress pattern, and no increase in the measured strain. The passage of a PLC band in the measuring gauge range increases the strain, and is associated with a small increase in stress. A pronounced increase in temperature occurs each time a PLC band passes through the center of the sample.

provided by Wijler *et al.*,²⁰⁾ who suggested that this is due to the fact that the specimen shoulders act as stress concentrators. The random serration in the stress-strain curve which were observed prior to fracture were caused by the formation and propagation of multiple PLC bands. The observed serrations are commonly referred to as “type A” serrations, which are caused by the repetitive continuous propagation of localized deformation bands, moving smoothly from one end of the specimen to the other along the specimen length. In the plateau range of the serrations, the tangent modulus of the stress-strain curve is very small, and therefore smaller than the stress magnitude. This implies that the Considere criterium for macroscopic necking is met. Typical type A serrations are due to *e.g.* Luders band propagation in low carbon steels, and in the latter case the stress-strain curve is flat during band propagation as all the deformation is concentrated in the band only. In the present case “modified” Type A bands were observed: the stress-strain curve was not flat during band propagation and it showed a small continuous increase in the plateau region. This suggests that there was some strain hardening in the region outside the PLC band.

At high strains, the multiple un-correlated discontinuities in the stress-strain curve are observed. These serrations are usually referred to as “type B” serrations.

3.2. Type A PLC Band Characteristics

As a PLC band moves through the specimen, the rest of the specimen undergoes little or no deformation, except for an elastic deformation. Due to the finite width of the band and the finite width of the slip distance of the dislocations in the band, the plastic flow can spill over to the adjacent undeformed region, and the band propagates through the sample. The temperature of the specimen is raised by the deformation band due to the highly concentrated strain, and this made it possible to record thermographs of the individual narrow mobile shear bands moving through the sample as shown in **Fig. 3**. The PLC bands were typically about 4 mm wide and made a specific angle to the tensile axis. The thermal image of the deformation band is wider than the actual physical width of the band. Deformation band characteristics such as the band strain rate and the disloca-

Fig. 3. IR thermographs of a propagating PLC band. In (a) the band is initiated at the bottom of the specimen, and moves to the opposite end of specimen, straining the whole specimen. The strain gauge only registers an increase in strain when a band passes through the gauge measurement range. (b) is an enlargement of the thermograph of a deformation band at a true strain of 0.29, the temperature increase, ΔT , is 5.0°C. (c) is an enlargement of the thermograph of a deformation band at a true strain of 0.38, $\Delta T=11.4^\circ\text{C}$.

tion density, which are calculated using the thermal width of the PLC bands, should therefore be considered as lower bound values. These band features did not change during the specimen deformation. The bands were inclined at 53.1° to the tensile axis, and this angle remained stable during the whole testing. This angle is close to 54.74° , the characteristic angle for localize necking of an isotropic material prior to fracture, which is also a discontinuity in deformation. The small deviation observed in the present case may be due to the normal anisotropy of the tested TWIP steel.²¹⁾ The angle is a clear indication of the fact that the strain along the band is prevented by the undeformed regions adjoining the band.

The velocity of the PLC band, v_B , was obtained by using the following equation:

$$v_B = \frac{L}{t_{out} - t_{in}}, \quad \text{mm/s}^{-1} \quad (1)$$

Where L is the measuring gauge length, t_{in} and t_{out} are the time at which the band enters and leaves the strain gauge measurement range, respectively. The PLC band velocity as a function of time and strain is shown in Fig. 4. It shows that the velocity of the bands decreased linearly with time and strain. It implies that when a band passes through the specimen, the material is strengthened and a large amount of barriers to plastic flow are generated which suppress the movement of the successive band. The results of Fig. 4 also reveal that the specimen fractured when the motion of the bands was fully suppressed, *i.e.* when the band velocity became zero.

Determination of $\Delta\epsilon_B$, the strain in the PLC band, was done using two methods. Using the local temperature rise ΔT and assuming adiabatic heating of the material in the deformation band, the measurement of the deformation band strain over a wide strain range could be obtained. As austenitic steels are known to have a low thermal conductivity, the assumption of adiabatic conditions is a reasonable one. $\Delta\epsilon_B$ is given by:

$$\Delta\epsilon_B \approx \frac{\Delta T \cdot \rho \cdot C_p}{\sigma_B \cdot \beta} \quad (2)$$

Where ΔT is the local temperature rise, ρ , the density, C_p , the specific heat, σ_B , the stress, and β , a factor taking the heat conversion efficiency into account. β was taken equal to 0.9.

As this first method does not allow for the measurement deformation band strains at very low strains, a second method was used to determine the band strain for strains less than 0.2. It is based on the crosshead displacement velocity, using²²⁾:

$$\Delta\epsilon_B \approx \frac{\dot{\delta}}{n \cdot v_B} \quad (3)$$

Where $\Delta\epsilon_B$ is the deformation band strain, $\dot{\delta}$, the crosshead displacement rate, v_B , the band velocity and n , the number of deformation bands. In the present case n was equal to 1. The effective strain rate in the deformation band, $\dot{\epsilon}_B$, must necessarily be very high due to the localized nature of the deformation. The local strain rate was calculated using the following formula²³⁾:

Fig. 4. PLC band velocity as a function of time (a) and strain (b). The velocity decreases linearly with strain and fracture occurs when the band velocity reaches zero. The continuous decrease of the band velocity results in an increase of the band strain $\Delta\epsilon_B$ (c) and the dislocation density within the band (d). Open symbols: deformation band parameter calculated using the temperature rise in the band. Close symbols: deformation band parameter calculated using crosshead displacement speed.

$$\dot{\epsilon}_B \approx \frac{\dot{\delta}}{d_B} \quad (4)$$

Where d_B is the deformation band width, and $\dot{\delta}$, the crosshead displacement rate $\dot{\epsilon}_B$ was found to be 0.096 s^{-1} . As the applied strain rate was 0.001 s^{-1} , this implies a considerable local strain rate increase by a factor of 100.

In the plateau segment of the stress-strain curves a deformation band moves through the sample; the localized strain is due to the mobile dislocation within the deformation band. This mobile dislocation density, ρ_m , was estimated using the following equation²⁴⁾:

$$\rho_m \approx \frac{\Delta\epsilon_B}{b \cdot d_B}, \quad \text{m}^{-2} \quad (5)$$

Where $\Delta\epsilon_B$ is the deformation band strain, b , the dislocation burgers vector length (0.256 nm), and d_B , the band width. This equation assumes that the band velocity is equal to the dislocation velocity. This is a reasonable assumption, which has been made previously by Sandstrom *et al.*²⁵⁾ for low carbon steel. The measured fraction of mobile dislocations (10^{10} – 10^{11} m^{-2}) appears to be rather low during serrated yielding. This is consistent with the observation of a serrated flow curve caused by an effective pinning of the dislocations.

The results shown in Fig. 4 clearly show that as the deformation proceeds, the band velocity decreases as it moves

Fig. 5. Strain hardening of Fe_{0.6}C₁₈Mn TWIP steel as a function of strain (a) and stress (b), and the instantaneous hardening as a function of strain (c). The indicated hump-like feature corresponds to “stage B”: the initiation of deformation twinning. The strain hardening rate shows instabilities due to discontinuities in the stress–strain curve. By using a smoothed stress–strain curve the mean strain hardening can be obtained. Note the absence of localized necking.

into increasingly deformed material. As a uniform deformation rate is maintained, the band strain increases and more stress is needed for the band to propagate.

The strain hardening behavior of Fe₁₈Mn_{0.6}C TWIP steel is reviewed in Fig. 5. The presence of plateaus in the stress–strain curves causes the strain hardening curves to have many discontinuities, which were removed by a suitable averaging procedure. It has been reported that low SFE polycrystalline f.c.c. metals exhibit different stages of strain hardening response, which are also observed for the low SFE Fe₁₈Mn_{0.6}C TWIP steel. According to Kalidindi²⁶⁾ the strain hardening regime for deformation twinning is initiated in stage B, which is indicated in Fig. 5, when pile-ups of coplanar slip dislocation act as a trigger for deformation twinning. In the present case, stage B occurred in the 450–550 MPa stress range, which corresponds to the 2–4% strain range. The occurrence of the stage B strain hardening regime implies that prior to twinning, some dislocation interactions are required to create initiation sites for the twins.

3.3. Microstructural Analysis

The microstructures of un-deformed and deformed

Fig. 6. TEM micrographs of un-deformed Fe_{0.6}C₁₈Mn TWIP steel, in which large grains and annealed twins are the main features (a). Dissociated dislocations and wide stacking faults are also observed. Micrograph (b) shows contrast from the partial dislocations and micrograph (c) shows the contrast due to the stacking fault between the widely separated partials.

Fig. 7. TEM micrograph of deformed Fe_{0.6}C₁₈Mn TWIP steel. Large amounts of twins, often oriented in parallel, are observed in the microstructure after tensile testing. The width of the deformation twins is approximately 60 nm. The observed twin boundaries are not flat and most twins are internally faulted by micro-twin formation.

Fe₁₈Mn_{0.6}C TWIP steel were studied by TEM and the main observations are presented in electron micrographs of Figs. 6 and 7. In the un-deformed samples, large grains ($\sim 30\ \mu\text{m}$) and wide annealed twins ($\sim 500\ \text{nm}$) were observed. The dislocation density before testing was very low. Widely dissociated partial dislocations and corresponding wide stacking fault were however observed. This is indicative of the low stacking fault energy of Fe₁₈Mn_{0.6}C. From

the available literature data, based mainly on theoretical calculations, the stacking fault energy of the Fe18Mn0.6 TWIP steel used for the present research was about 20–30 mJ/m².

As shown in Fig. 6, the observed stacking fault width was also very variable, even within the same dislocation segment. The deformed specimen contained large amounts of narrow deformation twins. Figure 7 shows that the twin boundaries are non-planar. The twins are also internally faulted. The mean width of larger deformation twins was about 60 nm. A high density of micro-twins was formed within the major twins. No evidence for the formation of h.c.p. ϵ -martensite or b.c.c./b.c.t. α' martensite was found.

4. Discussion

In the present work the observation of serrated stress–strain curves revealed the presence of a pronounced DSA phenomenon in Fe18Mn0.6C TWIP steel. Up to now, many researches have been done on high Mn TWIP steels with different compositions and mechanical properties. The occurrence or absence of serrated stress–strain curves for high Mn TWIP steel has been discussed in the introduction. From the present results, it is clear that plastic deformation of Fe18Mn0.6C TWIP steel exhibits a complex combination of different phenomena. Their characteristic high rate of strain hardening is caused by deformation twinning and/or a pronounced planar glide. The pronounced DSA is related to solute–dislocation interactions. In the following paragraph it will be argued that, in the case of Fe18Mn0.6C TWIP steel, these phenomena are very likely interacting strongly with each other.

Fe18Mn0.6C TWIP steel had plateau-type serrations in its stress–strain curve. It is well known that this serrated flow is due to DSA and associated with negative strain rate sensitivity. From the experimental observation, the localized deformation band always formed at the lower part of the specimen, and this usually lead to a small drop in stress which arrested the movement of the band. The aging of dislocations in the band front occurred until the stress was raised to its normal level. After the stress reached a normal level, the deformation band started to propagate through the specimen and the strain increased. The initiation and propagation of localized deformation bands produced a serrated flow curve. When the stress rises rapidly at low temperature, not much aging occurs and the band front propagates smoothly along the measuring gauge, producing type A serration. At higher temperature, the dislocations in the front become increasingly pinned and can only break away at a higher yield point. These repeated yield points appear as type B serration.

The origin of the high rate of strain hardening of Fe18Mn0.6C at room temperatures is as yet not known. The strain-induced transformation of austenite to ϵ or α' martensite does not play any role as the austenite was found to be stable during straining. Different theories have been proposed. They are reviewed schematically in Fig. 8. In the low SFE theory, it is assumed that the width of the stacking faults will reduce the frequency of cross slip to such an extent that planar slip will prevail and cause the high strain hardening rate (Fig. 8(a)).

Since a low SFE facilitates deformation-induced twinning during plastic straining, Bouaziz and Guelton⁴⁾ suggested that deformation twinning had a strong influence on the work-hardening rate of TWIP steel. They argue that the density of fine twins increases during deformation and gradually reduces the effective grain size. The mean free glide distance of dislocations will steadily be reduced as the twin boundaries act as effective barriers to their motion. According to Bouaziz, this “dynamical Hall–Petch effect” is the cause of the observed strain hardening. In this model, twins are regarded as impenetrable obstacles for dislocations (Fig. 8(b)) and contribute to strain hardening. TEM observations done in the course of the present work have however shown that the grains are not as effectively segmented by twins in the manner described by Bouaziz. In fact, most deformation twins tended to be strongly aligned parallel to each other in the grains in which they formed.

Gerold and Karnthaler,¹²⁾ on the other hand, argue that short range order (SRO) or short range clustering (SRC) in solid solution are the main cause of planar slip in f.c.c. materials. According to them, the stacking fault energy may not be the only parameter influencing the strain hardening. They argued that planar slip can actually occur in high stacking fault energy alloys. During planar slip, dislocations cannot easily avoid obstacles to their glide, and this results in an increased strain hardening (Fig. 8(c)). In their model, the origin of planar glide in concentrated solid solutions exhibiting SRO or SRC is due to the fact that dislocation glide removes the order or the clustering (Fig. 8(d)). The glide of a first dislocation on a glide plane requires a higher stress than the following dislocations on the same glide plane. The higher stress is provided by trailing dislocations, generated by a dislocation source, on the leading dislocation. The high stress on the leading dislocation increases its velocity and plastic deformation, as the trailing dislocations face a reduced lattice resistance after the passage of the first dislocation. The phenomenon, a fast deformation localized on a single glide plane, is known as “glide softening”. The stress level is reduced and the change in internal stress distribution leads to the activation of neighboring regions. The deformation is localized on the microscopic level. The original model of Gerold and Karnthaler ignored the dissociation of the dislocations in low SFE alloys.

In FeMnC alloys the SRO or SRC is very likely of a statistical nature, and related to the formation of octahedral clusters by substitutional Mn and interstitial C atoms (Fig. 8(d)). Owen and Grujicic¹¹⁾ have proposed a local-order model, in which the ordering is measured by the likelihood that a C atom will occupy the interstitial position in an octahedral cluster of metal atoms with n , an integer between 0 and 6, Mn nearest neighbors. They were able to calculate the stress opposing dislocation motion for a single isolated dislocation and for a sequence of dislocations moving on the same slip plane, and provided an adequate description of DSA in high Mn alloys.

Finally, Dastur and Leslie⁶⁾ have argued that twinning is not the main cause of work hardening in high Mn alloys. They have proposed an extension of the Cottrell theory of DSA in low carbon steels to explain the phenomenon of rapid strain hardening in Mn steels. According to them, the

Fig. 8. (a) Low SFE model: a low SFE leads to widely dissociated partial dislocations, which cannot cross slip. This pronounced planar slip results in a high strain hardening rate. (b) Dynamic Hall-Petch model of Bouaziz-Guelton: at low SFE, twinning is promoted. The intersection of twins forms a cell structure. The twin boundaries act as obstacle for dislocation glide. (c) Planar slip model: in concentrated solid solution alloys short range order (SRO) is possible and dislocations will destroy the SRO on their slip plane. Whereas the leading dislocation experiences a high resistance to its motion, subsequent dislocations can glide more easily in the same slip plane. This favors planar slip and result in a high rate of strain hardening. (d) Illustration of the reduction of the SRO during the passage of partial dislocation on their glide plane: the octahedral clusters are sheared and the C atoms are transferred to tetrahedral interstitial sites. Left: before slip. Right: after slip. (e) Dastur-Leslie model: dynamic strain aging in FeMnC alloy is caused by the re-orientation of Mn-C cluster in the dislocation stress field during dislocation motion.

high rate of strain hardening is associated with serrated stress-strain curves characteristic of DSA, and negative strain rate sensitivities. In this model, the attractive interaction between C and Mn atoms leads to the formation of Mn-C pairs. These atomic pairs can interact with dislocations by their fast reorientation at the core of moving dislocations (Fig. 8(e)). The weak points of their model are that it requires a low activation energy to allow for diffusional jumps at room temperature, it ignores the dissociation of the dislocations, and it does not describe the actual aging process in any detail.

Most mechanical models for localized plastic flow assume a softening on the macro-level. That this is not needed has been discussed by Sittner *et al.* who have shown that simultaneous deformation and transformation processes can explain this behavior in polycrystalline materials.²⁷⁾

In low SFE f.c.c. metals exhibiting twinning deformation, both the twinning strain and the dislocation glide are caused by the slip of similar $1/6\langle 112 \rangle$ type partial dislocation. In interstitial f.c.c. alloys a $1/6\langle 112 \rangle$ type shear disturbs the original f.c.c. ABCABC stacking sequence and traps the interstitial C atoms in transient interstitial position as illustrated in Fig. 8(d). In the case of a single partial dis-

location, a C atom will be transferred from an octahedral position to a tetrahedral position, and back to an octahedral position when the f.c.c. lattice is restored by the trailing partial dislocation. For a twinning dislocation the situation is different for two reasons: (a) there is no trailing partial dislocation restoring the f.c.c. lattice, and (b) there is a partial on every successive slip plane. In both cases the C atoms are transferred to higher energy tetrahedral positions, from which they can jump to reposition themselves in an octahedral position, considering the relatively small activation required for this. In the case of dislocation motion, if the C atoms can reposition themselves in an octahedral site before the passage of the trailing partial, this will require a higher stress. Clearly DSA will only be observed if the stacking faults are wide enough.

In the present work the local increase in temperature will result in a higher stacking fault energy, but this increase is apparently not enough to suppress deformation twinning. The fact that the deformation twinning cannot contribute much to strain and that twinning cannot easily be related directly to DSA, would suggest that in Fe18Mn0.6C, deformation twinning does not cause the pronounced strain hardening. Instead the formation of C-Mn octahedral clusters

and the associated strong planar glide would appear to offer an explanation both the observed strain hardening and the DSA.

5. Conclusion

The present study has shown that the austenitic Fe18Mn0.6C steel is characterized by a high strain hardening rate which results in a combination of high strength and uniform elongation. The material has very specific deformation properties, combining deformation-induced twinning and a pronounced strain localization. The properties of the deformation bands were analyzed in detail using IR thermal imaging, which revealed that the deformation was achieved by the propagation of one or two isolated PLC bands over most of the strain range. The motion of the deformation bands could be related to the observation of serrations in the stress-strain curve.

The main conclusions of the present contribution are as follows:

(1) The tensile deformation of the Fe18Mn0.6C TWIP steel is achieved by the initiation and propagation of narrow isolated mobile PLC bands. This results in a stress-strain curve with clear characteristic discontinuities in the stress and plateau-like segments.

(2) The temperature of the specimen was raised by the deformation band due to the highly concentrated strain, which made it possible to record thermographs of the individual deformation bands. The deformation bands were typically about 4 mm wide and made an angle of 53.1° to the tensile axis. These geometrical band features did not change during the specimen deformation. The band strain and velocity however were strain dependent. At low strains, the initial deformation band strain was ~ 0.005 and the band velocity was ~ 36 mm/s. Close to fracture the PLC band strain increased to ~ 0.09 and the band velocity was reduced to ~ 4 mm/s.

(3) Pronounced local thermal effects were associated with the narrow PLC bands. At low strains, the temperature increase due to the initial PLC band was $\sim 5^\circ\text{C}$. Close to fracture the temperature increase due to the final PLC band was $\sim 20^\circ\text{C}$.

(4) TEM images of undeformed material showed widely spaced partial dislocations, with a characteristic non-uniform separation. Highly deformed specimens showed a large amount of deformation twins which were internally faulted.

(5) Any modeling of the mechanical behavior of Fe18Mn0.6C TWIP steel must take strain localization caused by DSA into account and the condition of homoge-

neous deformation, as can be found in recent literature on the mechanical properties of TWIP steel, must be revised to include inhomogeneous deformation.

(6) From a technological point of view the PLC phenomenon, and hence DSA, must be avoided. This ensures stable material behavior during sheet forming processes as DSA gives rise to non-homogeneous plastic flow and may lead to surface defects on formed parts.

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