Substructure hardening in duplex low density steel

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HIGHLIGHTS

• Significant strain hardening capability of low density steel due to substructure development
• Nano-size partitioning of the austenite (~530 nm) and ferrite (~500 nm) grains
• Progressive deviation of twin boundaries from Ʃ3-coincidence due to substructure development
• Substructure refinement possesses a high portion of the measured flow stress

GRAPHICAL ABSTRACT

ABSTRACT

The present work was conducted to evaluate the effects of substructure development on the strain hardening behavior of Fe–17.5Mn–8.3Al–0.74C–0.14Si lightweight steel. This was performed applying tensile testing method at ambient temperature. The significant strain hardening capability of the experimental steel is attributed to the cell structure formation and its progressive evolution to subgrains over a wide range of applied strain. The continuous subgrain refinement with the applied strain could lead to the nano-size partitioning of the austenite (~530 nm) and ferrite (~500 nm) grains. The size of substructure (mesh length) appears to be stabilized at true strains above 0.35, thereby reducing the rate of work hardening and inducing subgrain rotation to higher misorientations. The contribution of substructure refinement is significant and possesses a high portion of the measured flow stress (~550 MPa for austenite and ~70 MPa for ferrite at the true strain of 0.5).

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1. Introduction

Several mechanisms are proposed to justify the observed continuous strain hardening and excellent combination of strength/ductility (over 50,000 MPa.%) in low density steels [1–3]. Twinning induced plasticity and/or transformation induced plasticity effects are expected to somewhat disappear in lightweight steels due to their higher aluminum content [4]. In fact, the stacking fault energy is sufficiently high to suppress or delay twin formation or martensitic transformation. Accordingly, the work hardening phenomena in austenitic low density steels may be attributed to the intensified dislocation interactions. This speculation is challenged considering some report concerning the occurrence of γ/α′ transformation in duplex aluminum-bearing structures [5–10] where the retained austenite characteristics define its stability and may locally trigger the tripping effect. This finding has been well

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followed and discussed by Seok Su Sohn et al. [4,11]. They believe that the level of stacking fault energy is not the main parameter determining the austenite deformation mechanisms. The austenite grain size, its preferred crystallographic orientation and its morphology may also act as influencing factors. The simultaneous operation of multi-deformation mechanisms, i.e. martensitic transformation and twinning, in these cases is attributed to the optimal mechanical stability of austenite. Yoo et al. [12,13], Park et al. [14] and Gutierrez-Urrutia et al. [15] also relate the corresponding strengthening mechanisms, which account for the excellent strength-ductility combination of low density steels, to the substructures associated with planar glide such as Taylor lattice, Taylor lattice domain boundaries and crystallographic microbands. To this end, the glide plane softening in association with the occurrence of short range ordering is considered as the main cause of planar glide. They also believe that the short range ordering would relate to the formation of C-Mn octahedral clusters or originate from the presence of shearable nano-sized $\kappa$-carbides.

As is understood, the available literature present a complex picture of the strain accommodation in austenitic low density steels. This complexity may be increased due to the constraint coming from the presence of $\delta$ ferrite along with the presence of nano-sized $\kappa$-precipitate in duplex or triplex structures. In addition it is theoretically speculated that besides the aforementioned mechanisms responsible for strain hardening capability of lightweight steels, the substructure formation and refinement associated with wavy dislocation configuration (cells, cell blocks or subgrains) may also be involved. This concept has been assessed in the case of conventional twinning induced plasticity steels with medium stacking fault energy [16,17]. It was found that the early hardening stage was accompanied with planar (Taylor lattices) and wavy (cells, cell blocks) dislocation configuration, the transition of which would be dictated by the chemical composition and the amount of imposed strain. Concurrently, competing deformation mechanisms, namely twinning and microbanding were found to contribute in continuous strain hardening behavior of the investigated steels. The present authors believe that the degree of substructure development associated with cell and subgrain formation might be intensified in the case of low density steels with relatively higher stacking fault energies (>60 mJ/m²). In fact, the maximum potential and individual strengthening contribution of such mechanism is still unclear. The aforementioned capability is assessed and well explored in the present work through elaborating a free $\kappa$ precipitate duplex ($\alpha + \gamma$) microstructure,
which could be deformed under low strain rate scheme. These would further highlight the role of subgrain formation in strain hardening behavior of low density steels, which has not been considered so far, and finally gives a better understanding in further alloy development. The simplicity and clearness of whole substructure characterization is one of our key strategies, which in previous literature was mainly explored through transmission electron microscopy.

2. Experimental

2.1. Experimental specifications

A duplex lightweight steel holding the chemical compositions of Fe−17.5Mn−8.3Al−0.74C−0.14Si was adopted as experimental material. High aluminum content was purposefully added to reduce the alloy density. The corresponding apparent density (6.8 g·cm⁻³), which was measured using a proper densitometer (Mettler-Toledo XP205, Mettler-Toledo AG), shows a considerable reduction (~18%) in comparison to fully austenitic structure. The stacking fault energy (SFE) was calculated based on a sub-regular solution thermodynamic model proposed by Curtze [18] and Saeed-Akbari [19] which considers the effect of Mn, C and in particular Al content along with the effect of initial grain size. Accordingly, the experimental steel holds the SFE (Γ) value of ~85−90 mJ/m² at 298 K (25 °C), which is much higher than that of twinning induced plasticity steels, 20 mJ/m² < Γ < 40 mJ/m². In order to predict the phase fractions as a function of temperature, the thermodynamic calculations were conducted by Thermo-Calc software, the result of which is given in Fig. 1. As is expected, the experimental material is capable to have a duplex microstructure where the austenite is the predominant phase over a wide range of temperatures. The addition of aluminum promotes the precipitation of nano-sized (Fe, Mn)₃AlC carbides (so-called κ-carbides) which would control the steel mechanical behavior, in particular at higher deformation temperatures.

The material was received in as electroslag remelted condition. The as-remelted material was homogenized at 1473 K (1200 °C) for 1 h and then subjected to hot rolling at temperatures in the range of 1473–1373 K (1200–1100 °C) followed by air quenching. The material was then annealed at 1373 K (1100 °C)/10 min in order to have a dual phase structure including ferrite and austenite phases. The obtained representative duplex microstructure is depicted in Fig. 2a, b. The presence of ferrite phase along the austenite boundaries is clearly evident. The inverse pole figure indicates the random texture of the parent austenite while ferrite was relatively texturized. The corresponding X-ray pattern (Fig. 2c) confirms the majority of phase fraction belongs to austenite. The austenite and ferrite fractions were approximately 90% and 10%, respectively.

2.2. Mechanical tests

The flow behavior of the experimental steel was studied through tensile testing at room temperature under the strain rates of 0.1, 0.01, 0.001, and 0.0001 s⁻¹ using a Gotech Al−7000 universal testing machine. In order to well trace the development of the substructure, some of the tests were interrupted at intermediate true strain of 0.12, 0.24, 0.4 and 0.5. A proper contact extensimeter was utilized to precisely measure the displacement during tensile testing. Considering the fact that these alloys do not exhibit considerable necking during room temperature deformation, i.e. negligible cavitation phenomenon, the relation of true and engineering stresses could be well defined. The work hardening curves, hardening rate vs. true strain, were also plotted using a proper filtration.

2.3. EBSD characterization

The microstructural analysis was performed in the regions underneath the fracture surface of the tensile specimens on a plane containing the tensile axis direction using electron back-scattered diffraction (EBSD) method. The samples were mechanically polished and further electro-polished at room temperature in a solution of CH₃COOH (92%) and HClO₄ (8%) at an operating voltage of 32 V. The EBSD studies were conducted using a Zeiss LEO 1530 FEG SEM operated at 20 kV. The instrument was equipped with a fully automatic HKL Technology (now Oxford Instruments) EBSD attachment. Data acquisition and post processing were performed using both the HKL Channel 5 and Aztec software including the modified Kuwahara filter [20] for the orientation averaging.

3. Results and discussion

3.1. Hardening behavior

The room temperature tensile flow curves obtained under the various strain rates of 0.1, 0.01, 0.001 and 0.0001 s⁻¹ are depicted in Fig. 3a, b. The experimental material shows a continuous yielding with moderate yield stress of about 400 MPa followed by a rapid strain...
hardening behavior and finally a high fracture strength of about 1000–1200 MPa. Another important observation is the incremental trend of ductility values with decreasing the applied strain rate. This may well indicate that a dislocation based mechanism would control the strain hardening and plasticity behavior of the material. The corresponding work hardening curves, classified into three regions, are given in Fig. 3c. The occurrence of rapid hardening (Stage II) is clearly recognized in different conditions. The strain hardening coefficient values observed in these regions, which varies between G/35–G/40, are similar to those found for dislocation substructure hardening in Fe-Mn-C alloys (laying in the range of G/20–G/40 [21–23], where G is the alloy shear modulus, G = 70 GPa [24]). In addition, the length (L₀) and amplitude (H₀) of the rapid hardening region is well higher in the case of lowest applied strain rate. The rate of work hardening under the strain rate of 0.0001 s⁻¹ starts increasing from the true strain of 0.15 and surprisingly exceeds from the other curves obtained under the higher strain rates. Due to the positive effects of lower strain rate on the length and amplitude of hardening rate, it is speculated that the random distributions of dislocations would not be involved, although the slip processes may produce high densities of dislocations in the form of cell structure bounded with tangled walls on the active slip planes. In fact, the lowest strain rate would promote dislocations relaxation into boundaries by dynamic recovery to the degree at which the deformed microstructure (containing deformation cells with diffuse arrangements in their walls) evolve to subgrains with sharp and well-ordered low angle boundaries.

In order to justify the aforementioned speculation, the microstructure of the specimen deformed under the strain rate of 0.0001 s⁻¹, in which the probable microstructural changes are intensified, were precisely examined.

3.2. Substructure development within austenite

The deformed microstructures at different levels of strains were analyzed to evaluate the substructure evolutions (i.e., subgrain formation, refinement and its probable rotation) during tensile deformation. The austenite grain boundaries in the initial microstructure and the microstructures deformed up to the strains of 0.12, 0.24 and 0.4, are demonstrated by high resolution EBSD maps in Fig. 4a, c, e, and g.

Fig. 4. The high resolution orientation maps illustrating the austenite boundaries, besides the corresponding mean angular distribution in as-received condition (a, b) and in the microstructures deformed up to the strain of 0.12 (c, d), 0.24 (e, f) and 0.4 (g, h).
respectively. The selected interruption strains and their belonging microstructures are corresponding to the stages (I), (II), and (III) of hardening, as were indicated in Fig. 3c. The substructure development was also analyzed in a more quantitative way considering the misorientation’s distribution between neighboring grains/subgrains (Fig. 4b, d, f, and h). In as-received condition and at lower applied strain of 0.12, a large fraction of misorientations lays between 15 and 60°. In contrast, at higher imposed strain, the number of sub-boundaries increases and the frequency of their misorientations (ranging between 0 and 15°) increases significantly. In fact, the continuous and dynamic evolution of subgrains within the austenite grains is clearly rationalized. As is expected, the size of such dislocation substructures scales inversely with the applied resolved stress (indicated as labels in Fig. 4d, f, and h). The annealing twin maps are also superimposed on the deformed microstructures up to different strain levels. The Σ3 twin boundaries with straight-sided bands that run across the grains are identified as 60° rotation around a (111) axis and highlighted in red. The local pole figure of (111) planes around these bands indicates one-point coincident which represent their twinning nature. The rest of the high angle boundaries are presented in black. The high frequency of annealing twins (of about 57%) can be well traced in twin map of initial microstructure (Fig. 4a). As is observed, the frequency of twin boundaries is decreased as the deformation proceeds (Fig. 4c, e, g), which indicates their progressive deviation from the ideal coincidence (in terms of misorientation axis/angle) during straining. Plotting the distribution of the misorientation angles across the high angle boundaries (Fig. 6) indicates a significant decrease in annealing twin frequency. Such a change in misorientation angle (decreasing trend) may also be followed by some deposed general high-angle grain boundaries too. The misorientation broadening observed in as-received condition (Fig. 6a) is eliminated at higher imposed strains. Such behavior has been previously reported during the room temperature deformation behavior of conventional twinning induced plasticity steels possessing low stacking fault energy values [25,26]. This can be discussed considering the local interactions between the mobile dislocations and the twin or high angle boundaries and may introduce a significant influence on the hardening behavior of the material. However, in present case a thorough transmission electron microscopy is required to assess the characteristics of the evolved boundaries through distortion or possible detwinning process. This may be considered as the main concepts of the future works.

In order to more precisely examine the evolved substructure, a higher magnification of the selected area, which was deformed up to the true strain of 0.5 (corresponding to the stage of III hardening), is provided in Fig. 5. The blue, green, and magenta lines denote low-angle grain boundaries holding misorientations in the range of 0.7° < Θ < 2°, 2° < Θ < 5°, and 5° < Θ < 15°, respectively. The black lines represent the high-angle grain boundaries having misorientations of >15°. The red lines are Σ3. These results demonstrate several significant features. The initial austenite grains consist of a variety of cell structures after room temperature tensile deformation and the cell boundaries hold the misorientation angles of ~15°. As is observed, the cell walls (which are now the subgrain boundaries) have different misorientations which enable further sub-classification; the network of the cell structure at the vicinity of the grain boundaries is denser and the sub-boundaries possess higher misorientation angles in the range of 2° < Θ < 5° and 5° < Θ < 15° compared to the interior grains (0.7° < Θ < 2°). The measurements indicate an appreciable subgrain refinement; the average subgrain size at true strain of 0.5 is 530 nm (misorientation angle threshold has been considered to be 0.7°).

In low stacking fault energy twinning induced plasticity steels, which the cell formation and development is hard to be occurred, the diffused dislocations arrangements such as tangles or other rather random structures would be found even after larger strains [15]. In medium range of stacking fault energy, the substructure formation and development (refinement) associated with planar or wavy configuration generally contribute in the early stage of hardening (up to 0.1 true strain) [16,17]. In the present work, the progressive and effective sequential formation of the subgrains (with well-ordered and sharp low angle boundaries) and its refinement during tensile straining is considered as the exceptional microstructural feature of the present experimental material compared to the conventional twinning induced plasticity steels. This is introduced as the main responsible mechanism for the observed high strain hardening rate and also the higher capability of the material to be work hardened during room temperature straining. Another important point which should be considered here is the presence of deformation twins in the developed microstructures up to higher amount of strain corresponding to the stage (III) of hardening (Figs. 4b, 5). The deformation twins are much finer and inhomogeneous in appearance (compared with the annealing twins) and inherit the substructure configuration of the parent subgrains. It has been shown that these coherent twins would not act as strong barriers against dislocation movement and is not the main reason for the observed strain hardening in flow behavior of the experimental steel. In fact the deformation twinning contributed in strain hardening specifically at the latest stage. The reason for deformation twinning and its specific characteristics has been thoroughly described in our previous work [27].

### 3.3. Substructure development within ferrite

In order to trace the development of the substructure within the ferrite grains, the deformed microstructure of the interested area at true strain of 0.5 is depicted in Fig. 7a. For further clarification, the ferrite sub-boundaries after interrupted tests at true strain of 0.12, 0.24, and 0.4 along with the histograms of the misorientation angles are also given in Fig. 7b–e. The ferrite deformation mechanism is mainly consisted of dislocation slip owing to its high stacking fault energy and many active slip systems; this may effectively result in an intensified substructure development. The formation of sub-boundaries within ferrite is perceptible even at low strain of 0.12 where most of the sub-boundaries demonstrate misorientations lower than 15°. As is expected, the relative content of the larger-angle sub-boundaries gradually increases with strain.

The subgrain evolution in ferrite grains yields to significant refinement in the ferritic region; the measurements show an average subgrain size of 500 nm (at true strain of 0.5) in the ferritic region. Comparing the subgrain size of ferrite and austenite (Figs. 7 and 4) at the same strain level reveals a faster refinement kinetic within the ferrite phase. This can be justified considering the higher stacking fault energy.
and the higher rate of self-diffusion in ferrite, which makes the rearrangement and annihilation of dislocations comparatively easier [28].

3.4. Substructure contribution in strain hardening

At the early stage of deformation (stage I, Fig. 3c), a heterogeneous distribution of low density dislocation exist and the dislocation movement would proceed by little interaction with other dislocations. This is ended where a fairly uniform dislocation distribution of moderate density is developed (deformation cell formation). At this point the preferred dislocation configuration surround the cells of relatively low dislocation density, and the transition of the deformation cells to subgrains and further subgrain refinement gradually starts to invade the interior grains (stage II). The subgrain refinement continues too slowly (so is called subgrain size stabilization in this work) in stage III up to fracture.

To assess the influence of dislocation substructure evolution on the rate of strain hardening, its contribution to the flow stress is evaluated using Eq. (1) [21], where $M$ is the Taylor factor estimated from the corresponding texture, $G$ is the shear modulus (~70 GPa for austenite [24,29] and ~80 GPa for ferrite [30]), $b$ is the magnitude of the Burgers vector ($2.5 \times 10^{-10}$ m), $D$ is the substructure size and $K$ is constant.

$$\sigma = M\tau = MGKb/D$$ (1)

The Taylor factor was estimated for grains oriented $\langle 001 \rangle$ //tensile axis directions ($M = 2.5$ for austenite and 2.4 for ferrite, which are similar to those reported for cell forming materials [16]). $K$ was also calculated through plotting the variation of the average subgrain size in ferritic and austenitic region vs. resolved shear stress, as is demonstrated in Fig. 8a. The calculated values for ferrite $K_\alpha = 6.6$ and austenite $K_\gamma = 7.1$ are well higher than that of conventional twinning induced plasticity steels (which is ~3.7) with low to medium stacking fault energy [16,21], but fairly lays in the reported range for high stacking fault energy materials (~7–8) [31].

Considering the law of mixture, the contribution of substructure refinement within ferrite and austenite grains in strain hardening was
calculated and shown in Fig. 8b. The parent austenite boundaries, annealing twin boundaries and austenite/ferrite interfaces play the main role in strain hardening at early and intermediate stages of deformation. The contribution of substructure increases with increasing imposed strain, due to distortion of the annealing twins, and mainly because of the progressive formation/refinement of the subgrains. The calculated stress at the true strain of 0.5 (~550 MPa for austenite and ~70 MPa for ferrite) possesses a high portion of the measured flow stress. This analysis emphasizes the key role of substructure development in dictating the flow stress and the strain hardening.

3.5. Comparison

As a final point, the mechanical properties obtained from the present work and the results obtained from previous studies on Fe–Mn–Al–Ca alloys with different stacking fault energies and initial microstructures are summarized in Table 1. The valuable parameters such as weight reduction, specific strength, yield to ultimate strength ratio, and formability index have been extracted and given. Similar to high strength/formable low density steels which may benefit from twinning, tripping, planar glide and crystallographic microbanding, the present experimental steel demonstrates an outstanding mechanical properties: the yield strength of 400 MPa, ultimate tensile strength of 1200 MPa, elongation to fracture of 63%, weight reduction of 16%, specific strength of 110 MPa cm³/g, yield to ultimate strength ratio of 0.54, and formability index of 50,000. The unique characteristic of the present lightweight steel is mainly attributed to the rapid dislocation multiplication and their further interactions, which result in progressive subgrain evolution (formation, refinement and most likely rotation) within the ferrite and austenite grains. Simultaneously, the role of annealing twins connected with persistent substructure development would be influential.

4. Conclusion

The involved hardening mechanisms at room temperature deformation of a duplex (α + γ) low density steel were precisely studied emphasizing on the quantitative and qualitative characterization of substructure development. The following conclusion were drawn:

- The continuous and progressive formation of the subgrains within the austenite and ferrite grains was clearly recognized. The network of the cell structure at the vicinity of the grain boundaries was denser and the sub-boundaries possess higher misorientation angles (2° < θ < 15°) compared to the interior grains (0.7° < θ < 2°). The size of such dislocation substructures scaled inversely with the applied resolved stress. The measurements indicated an appreciable subgrain refinement, average subgrain size of 530 nm within austenite and 500 nm within the ferrite.
- The Taylor factor was estimated for grains oriented ⟨001⟩//tensile axis directions (M = 2.5 for austenite and 2.4 for ferrite) which were similar to those reported for cell forming materials. K constants (Kα = 6.6 and Kγ = 7.1) also were well higher than that of
conventional twinning induced plasticity steels (~3.7) but fairly lays in the reported range for high stacking fault energy materials (~7–8).

- The evolution of the subgrains continued up to fracture (but slower after size stabilization), which would apparently indicate the substantial role of such mechanism in strain accommodation during room temperature tensile deformation. This justified the obtained elongation to fracture values (~63%). The contribution of substructure refinement in strain hardening rate (~550 MPa for austenite and ~70 MPa for ferrite at the true strain of 0.5) was also significant and possessed a high portion of the measured flow stress.

Fig. 8. The variation of resolve shear stress with mean subgrain size within ferrite and austenite grains (a), and the contribution of the subgrain refinement to the overall flow stress (b).

<table>
<thead>
<tr>
<th>Chemical comos (constituent phases)</th>
<th>Density and weight reduction</th>
<th>-SFE (mJ·m⁻²)</th>
<th>Dominant deformation mechanisms</th>
<th>Rapid hardening</th>
<th>Y.S/UTS</th>
<th>Specific strength (MPa·cm³/g)</th>
<th>Formability index (MPa%)</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe–29Mn–5.2Al–0.06C–0.6Si (α + γ + DO3)</td>
<td>7.3 g/cm³, 6.5%</td>
<td>42</td>
<td>Twinning</td>
<td>Yes</td>
<td>0.62</td>
<td>70</td>
<td>27,000</td>
<td>[32]</td>
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<tr>
<td>Fe–21Mn–1.6Al–0.11C–2.5Si (γ)</td>
<td>7.7 g/cm³, 1.2%</td>
<td>21</td>
<td>Twinning + tripping</td>
<td>Yes</td>
<td>0.42</td>
<td>91</td>
<td>56,000</td>
<td>[33]</td>
</tr>
<tr>
<td>Fe–8.5Mn–5.6Al–0.3C (α + γ)</td>
<td>7.2 g/cm³, 8.5%</td>
<td>51</td>
<td>Twinning + twinning</td>
<td>Yes</td>
<td>0.68</td>
<td>101</td>
<td>56,000</td>
<td>[4]</td>
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<tr>
<td>Fe–20Mn–9Al–0.6C (α + γ)</td>
<td>6.84 g/cm³, 13%</td>
<td>70</td>
<td>Planar glide</td>
<td>No</td>
<td>0.64</td>
<td>117</td>
<td>37,000</td>
<td>[5]</td>
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<tr>
<td>Fe–27Mn–12Al–0.8C (α + γ + α + γ + DO3)</td>
<td>6.54 g/cm³, 17%</td>
<td>92</td>
<td>Planar glide</td>
<td>Yes</td>
<td>0.85</td>
<td>146</td>
<td>40,000</td>
<td>[34]</td>
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<tr>
<td>Fe–12Mn–5.5Al–0.7C (α + γ + α)</td>
<td>7.2 g/cm³, 8.5%</td>
<td>69</td>
<td>Planar glide (Taylor lattice) + twinning</td>
<td>Yes</td>
<td>0.89</td>
<td>171</td>
<td>33,000</td>
<td>[11]</td>
</tr>
<tr>
<td>Fe–27Mn–11.5Al–0.95C (α + γ)</td>
<td>6.55 g/cm³, 16.5%</td>
<td>80</td>
<td>Planar glide (Taylor lattice) + microbanding</td>
<td>Yes</td>
<td>0.85</td>
<td>141</td>
<td>46,000</td>
<td>[35]</td>
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<td>Fe–28Mn–9Al–0.8C (γ + α)</td>
<td>6.78 g/cm³, 14%</td>
<td>85</td>
<td>Planar glide (Taylor lattice) + microbanding</td>
<td>Yes</td>
<td>0.52</td>
<td>124</td>
<td>80,000</td>
<td>[14]</td>
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<tr>
<td>Fe–30.5Mn–2.1Al–1.2C (γ)</td>
<td>7.2 g/cm³, 8.5%</td>
<td>63</td>
<td>Planar glide (Taylor lattice) + wavy (cell, cell blocks) + twinning</td>
<td>No</td>
<td>0.26</td>
<td>188</td>
<td>88,000</td>
<td>[16]</td>
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<tr>
<td>Fe–20Mn–4Al–0.3C (γ)</td>
<td>7.4 g/cm³, 5.5%</td>
<td>43</td>
<td>Planar glide + wavy glide (cell blocks) + twinning + microbanding</td>
<td>No</td>
<td>0.51</td>
<td>81</td>
<td>29,000</td>
<td>[17]</td>
</tr>
<tr>
<td>Fe–18Mn–8Al–0.7C–0.15Si (α + γ)</td>
<td>6.8 g/cm³, 14%</td>
<td>85</td>
<td>Wavy glide (subgrain formation) + twinning</td>
<td>Yes</td>
<td>0.54</td>
<td>110</td>
<td>50,000</td>
<td>This work</td>
</tr>
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* Tripping refers to the austenite to martensite transformation during straining.

References


