1. Introduction

Low-alloy transformation-induced plasticity (TRIP) steels are currently widely used as materials for car body sheets in the automotive industry. They were developed during the 1990s and consist of a ferrite matrix, containing bainite, martensite and a fraction of a metastable retained austenite. Under plastic deformation, the retained austenite transforms to martensite, which means these steels have a good combination of high strength and high ductility. These high strength alloys are then easily shaped and allow manufacturers to reduce car weight and improve energy absorption as well. In recent years, the effect of heat treatment and alloying elements on the complex microstructures of TRIP steels has been the subject of research. The mechanical properties of these steels have been determined mainly by using uniaxial tensile tests, which yield strength and ductility values.

In the automotive industry TRIP steels sheets are typically formed by stamping, and one of the problems associated with deep drawing processes is the difference between the final shape of formed parts before and after the removal of tools. These variations are caused mainly by springback and the Finite Element Method (FEM) has been used to correct these part shape errors. Since FEM simulations need more realistic material properties, research has been undertaken in recent years into the variation of Young’s Modulus with strain. The introduction of elastic variations into FEM codes has led to an accurate prediction of springback and a reduction in shape errors in sheet metal stamping.

A study of the elastic response before and after tensile plastic strain was undertaken for two commercial low-alloyed TRIP steels. These steels, TRIP 700 (C–Mn–Al alloy) and TRIP 800 (C–Mn–Si) are commercial alloys used in sheet metal stamping. The behaviour of the instantaneous tangent modulus ($E_T$) versus stress during loading and unloading was measured for each degree of prestrain. Loading curves show a decrease in the $E_T$ of the deformed samples as compared with the undeformed state. Though at low stresses a highly linear response was measured for both steels, a decrease was obtained for TRIP 700 as strain increased, whereas TRIP 800 remained unchanged. During unloading, a progressive decrease in $E_T$ was obtained in all deformed states, with lower chord modulus values as the tensile plastic prestrain increased. The inelastic response observed is attributed mainly to microplastic strain caused by the displacement of mobile dislocations. Thus, the differences between the two TRIP steels studied are related to the microstructure and the different dislocation structures observed in them. A notable consequence of this study is a better accuracy in the prediction of springback passes due to a better understanding of these inelastic effects that stems from going beyond mere use of traditional Young’s modulus values.

KEY WORDS: TRIP steels; springback; Young’s modulus; dislocation structure.
This microplastic deformation results firstly from the short range of motion of mobile dislocations,\(^1^{,19}\) and secondly from the bowing of the dislocation line between pinning points following models proposed by Mott and Friedel,\(^2^{0,21}\) and by Granato and Lücke.\(^2^{2}\) This extra deformation, which occurs below the internal strength level of the material, is recoverable, and is affected by the dislocation density and the total length of dislocation line that is able to move or bow out. Thus, when \(\varepsilon_{\text{mp}}\) grows, a decrease in \(E\) is expected. The following Eq. (2), proposed by Granato and Lücke,\(^2^{3}\) summarizes this idea:

\[
\varepsilon_{\text{dis}} = \rho \cdot b \cdot \xi \quad \text{.........................(2)}
\]

where \(b\) is the magnitude of the Burgers vector, \(\rho\) is the density of the mobile dislocations and \(\xi\) is the average displacement of a dislocation line of length \(l\).

As springback appears after tools have been removed, it is the behaviour of Young’s modulus ‘during unloading’ that should be investigated to obtain a more accurate prediction of the recovery from strain after forming. Ghosh et al.\(^1^{,23}\) used uniaxial tensile tests to study strain recovery after plastic prestrain for a type of steel used in sheet metal stamping, and demonstrated the nonlinearity of the unloading part of the stress–strain curve. This ‘inelastic behaviour’ results in an extra deformation that is not predicted by the usual values of Young’s modulus, and again is resumed as microplastic strain. This deformation must be accounted for in FEM simulations in order to better predict the final shape of the pieces stamped. The nature of microplastic strain during unloading is mainly associated with the backward movement of the dislocations that pile up during deformation.

This paper describes the evolution with plastic prestrain of the elastic part of the stress–strain curve obtained by tensile test during loading and unloading, for sheets made from two different types of TRIP steel. This study focuses on all the deformation processes that take place until the appearance of the neck. Simultaneously to the deformation process, the dislocation structure was examined by TEM. The two types of low-alloyed TRIP steel sheets chosen are commercially known as TRIP 700 (a C–Mn–Al) and TRIP 800 (C–Mn–Si). While their microstructure and tensile behaviour has been studied,\(^9,10,24,25\) the effects of plastic prestrain on their Young’s moduli during loading are less well-known.\(^2^{9}\)

This study was undertaken to determine the influence of the dislocation structure on the elastic behaviour of previously deformed materials and the total amount of strain recovery that takes place after the deformation process. This knowledge can enable better results to be obtained in the prediction of springback when stamping sheets made from these two alloys.

2. Materials and Experimental Procedure

The chemical composition of the sheets examined is shown in Table 1. TRIP 700 is a C–Mn–Al low-alloyed steel and TRIP 800 contains C–Mn–Si. Microstructures of these multiphase steels were examined by optical microscopy (OM) and scanning electron microscopy (SEM). The specimens for OM were etched in a 2% nitric solution. The volume fraction of retained austenite was measured by X-ray diffraction using Cu-K\(_\alpha\) radiation. The integrated intensities of the (200)\(_\alpha\) and (211)\(_\alpha\) peaks and the (220)\(_\gamma\) and (311)\(_\gamma\) were used in the direct comparison method.\(^2^{6}\)

The steel sheets were 2 mm in thickness for both types of steel. Samples for the uniaxial tensile test in the longitudinal direction were cut from the as-received sheets. The length of the reduced section was 75 mm for TRIP 700 and 125 mm for TRIP 800. In order to maintain approximately the same strain rate in both steels (1 \(\times 10^{-4} \text{ s}^{-1}\)), the speed of the test machine was 0.5 mm/min for TRIP 700 and 0.8 mm/min for TRIP 800. Tensile test were conducted in an INSTRON 5585. Several specimens were deformed previously at different strains, and high precision electrical resistance strain gauges (EA-06-062AP-120 from Micro-Measurement Co.) were subsequently glued to the surface of each sample. Pre-deformed samples were stored for 48 h at room temperature before examination. Another set of samples was deformed step by step according to the following procedure: starting in the initial state, the elastic response is measured during loading, and plastic strain was then introduced. This plastic strain was limited to less than 3% strain to ensure the precision of the strain gauge. The unloading stress–strain curve was measured when the stress began to decrease. The strain gauge was then replaced and the cycle is repeated, as is shown in Fig. 1.

After being corrected for gauge factors, the true stress–true strain curves were plotted for loading and unloading, and then an ‘Instantaneous tangent modulus’ (\(E_1 = d\sigma/d\varepsilon\)) was calculated from these curves.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Al</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP 700</td>
<td>0.25</td>
<td>1.57</td>
<td>0.05</td>
<td>1.90</td>
<td>0.013</td>
<td>0.002</td>
<td>0.02</td>
</tr>
<tr>
<td>TRIP 800</td>
<td>0.25</td>
<td>1.68</td>
<td>1.75</td>
<td>0.03</td>
<td>0.02</td>
<td>0.002</td>
<td>0.01</td>
</tr>
</tbody>
</table>

\[ E = \frac{\sigma}{\varepsilon_x + \varepsilon_{\text{mp}}} \quad \text{.................................(1)} \]
During unloading, the $E_T$ exhibits one of two different kinds of curves (Fig. 2), depending on whether the crosshead stops for a few seconds between the end of loading and the beginning of unloading. When deformation doesn’t stop for a few seconds before the unloading, right as unloading starts the initial $E_T$ values are higher than expected because some forward plastic strain exists as the crosshead stops and starts to move in the opposite direction. If deformation is stopped for a few seconds before unloading begins, these high values are eliminated. Except at these initial values, the curves obtained from both procedures are similar.

Transmission electron micrographs (TEM) were taken from the tensile specimens. Thin foils were ground on 600-grit silicon carbide to a thickness of 250 $\mu$m and were then cut into discs 3 mm in diameter using a Gatan ultrasonic disc, in order to avoid the deformation of the specimens. Finally, TEM samples were electrochemically thinned on a Struers Tenupol II using 6% vol HClO$_4$ and 94% vol CH$_3$COOH. Three foils were examined from each deformed specimen on a Philips CM30 operating at 300 kV with a double-tilt stage. The dislocation structures were photographed with [001] zone axis and [200] two-beam condition, satisfying conditions for dislocation visibility.

3. Results

3.1. Microstructures of TRIP Steels

The microstructures of the two steels in as-received conditions are shown in Figs. 3 and 4, and a distribution of the volume-fractions of the different phases is presented in Table 2. TRIP 700 consists of a ferritic matrix in which second phases are not homogeneously distributed. The main ferrite grain size is 15 $\mu$m and the ratio between ferrite and the second phases is approximately 75 : 25. In general, bainite and retained austenite are usually located at the grain boundaries of ferrite grains. However, there are some zones in which grains of bainite, together with plates of retained austenite and ferrite, form a fine structure. Some traces of martensite can be observed in the retained austenite. Finally, some carbides and retained austenite appear, isolated into the ferrite grains.

The microstructure of TRIP 800 contains a higher volume fraction of second phases (the ratio between ferrite to second phases is 55 : 45) and their distribution is also more homogeneous. Micrographs show that it has a multiphase structure made up of bainite, ferrite and retained austenite.
In certain areas the structure is so fine that it is difficult to distinguish between ferrite and bainite. Nearly all the grain boundaries of ferrite have disappeared and small bainite or austenite grains are found there instead. The main clearly distinguishable ferrite grain size is 10 μm. As in TRIP 700, some austenite grains appear partially transformed into martensite.

### 3.2. TEM Examination

Figure 5 shows the evolution of the arrangement of dislocations in ferrite grains with cold tensile deformation for TRIP 700. In an initial stage prior to deformation, long dislocations can be seen, with few intersections between them. The main clearly distinguishable ferrite grain size is 10 μm. As in TRIP 700, some austenite grains appear partially transformed into martensite.

In certain areas the structure is so fine that it is difficult to distinguish between ferrite and bainite. Nearly all the grain boundaries of ferrite have disappeared and small bainite or austenite grains are found there instead. The main clearly distinguishable ferrite grain size is 10 μm. As in TRIP 700, some austenite grains appear partially transformed into martensite.

<table>
<thead>
<tr>
<th>Steel</th>
<th>V_α</th>
<th>V_B</th>
<th>V_M</th>
<th>V_Y</th>
<th>(\sigma_{0.2}) (MPa)</th>
<th>(\sigma_{\text{max.}}) (MPa)</th>
<th>(\epsilon_{\text{eff}})</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP 700</td>
<td>75</td>
<td>17</td>
<td>1</td>
<td>6-7</td>
<td>420</td>
<td>785</td>
<td>0.18</td>
</tr>
<tr>
<td>TRIP 800</td>
<td>55</td>
<td>35</td>
<td>2</td>
<td>8</td>
<td>520</td>
<td>1020</td>
<td>0.17</td>
</tr>
</tbody>
</table>

### Table 2. Mechanical properties and volume-fractions (%) of the microstructural constituents composing TRIP 700 and TRIP 800 steels. \(\alpha\): ferrite; \(\alpha_B\): bainite; \(\alpha_M\): martensite and \(\gamma\): austenite.

Fig. 5. TEM Micrographs showing the evolution of dislocation arrangement in ferrite grains with strain in TRIP 700. (A1 and A2) 0% strain, (B) 4%, (C) 12% and (D) 16%.

Fig. 6. TEM Micrographs showing the evolution of dislocation arrangement in ferrite grains with strain in TRIP 800. (A) 0% strain, (B1) 4%, ferrite and bainite, (B2) inside a ferrite grain (C) 12%, ferrite and bainite and (D) 16%, inside a ferrite grain.

The evolution of the dislocation structure with strain is shown for TRIP 800 in Fig. 6. While there is a larger presence of bainite, the ferrite has a smaller grain size. This creates a fine structure in which bainite sheaves and ferrite density increases, as was expected. An important fact is that dislocations are uniformly distributed along the grains, with few intersections between them and an absence of tangles. This development of dislocation structure in ferrite grains is similar to other previously reported for pure iron. The presence of hard phases (martensite and carbides) in grain boundaries did not produce a fast increase of an entangled structure or the formation of high density dislocation walls (HDDWs).

As the strain advances, more dislocations are produced, which causes the number of dislocations to increase in ferrite grains while the dislocation lines are getting shorter. However, the most important factor is that the structure changes slowly, since tangles or HDDWs are rarely encountered in the centre of grains. This situation remains the same for the most deformed samples examined by TEM (0.180 strain), in which single dislocation lines can be resolved inside some ferrite grains and from time to time in some grain boundaries.
grains are closely connected, which makes it difficult to find isolated ferrite, especially when deformation begins. In the initial state there are no differences inside the ferrite grains compared to TRIP 700. At 4% strain, the dislocations in ferrite tend to form tangles and dislocation forests, which grow from bainite or austenite–martensite interfaces. At the same time, a high dislocation density is observed inside the plates of the bainite sheaves, and some high density dislocation walls are observed. Neither austenite nor cementite was found between these plates, the latter probably due to the high Si content in TRIP 800. As deformation progresses, more tangles are formed and it is difficult to observe free dislocations in the ferrite grains. At 16% strain, a rough cellular structure forms in the centre of the ferrite grains. Unlike TRIP 700, band structure is not observed in the longitudinal sections.

3.3. Mechanical Properties

Figure 7 shows the true stress–true strain curves obtained from uniaxial tensile tests for the two steels studied. The main mechanical properties determined from these curves are illustrated in Table 2. Both steels experience the TRIP effect to the same extent, as they have a similar content in initial retained austenite and their rate of transformation to martensite is very alike (Fig. 8). During the first part of deformation process the total amount of austenite transformed is high, but the rate slows as the strain increases. Finally, both steels present an equal amount of untransformed austenite (around 2–3%). The two steels studied exhibit a similar uniform elongation and TRIP 800 shows higher strength levels than TRIP 700, which is a result of the slight difference in their martensite contents, the greater volume fraction of bainite rather than ferrite in TRIP 800 and the strengthening effect of silicon in ferrite.24,30)

3.4. Tensile Tests: Loading

Figure 9 illustrates the evolution of the instantaneous tangent modulus, $E_T$, during loading at different plastic prestrains for the two steels studied. In the undeformed state both steels exhibit similar behavior, in which $E_T$ shows very little variations until stresses of 250–300 MPa are reached, at which point it drops suddenly and plastic strain begins. In the first zone, the apparent Young’s modulus ($E_A$) can be calculated using linear regression; the values found were 195 and 205 GPa for TRIP 700 and TRIP 800 respectively. Figure 10 shows the typical behaviour of the $E_T$ versus true stress in the loading of a previously deformed sample. As can be seen, the stress–strain response deviates from linearity at stresses well below those that occur at the beginning of yielding. As a result, three different stages can be distinguished in these curves: a first stage (Stage I), which entails the first zone in which the $E_T$ has a constant values; a second stage (Stage II), in which the $E_T$ decreases at a steady rate; and finally, a final stage (Stage III), in which large scale plasticity occurs. In Stage I, when loading starts, the $E_T$ remains almost constant. However, depending on the sample, small decreases in the $E_T$ can be observed. Here, an apparent Young’s modulus ($E_A$) is obtained in the same manner as for the undeformed samples. The $E_A$ is recalculated by linear regression at every data pair ($\sigma$–$e$), and for all the curves the coefficient of linear regression was above
0.999. The results for the calculations of the $E_A$ in Stage I are shown in Fig. 11. For TRIP 700, the $E_A$ diminishes quickly in the first stage of deformation, which is followed by a slight and constant drop as strain increases. The final decrease in the $E_A$ reaches a level of 5–6%. On the other hand, TRIP 800 experiences less variation, and for the pre-deformed samples, their $E_A$ values fluctuates around the initial value in the undeformed state.

In Stage II, a clear decrease in the $E_T$ is observed. This drop started to occur, in most cases, when the stress reached 100–140 MPa. Between this point and the beginning of the plastic range, nearly all of the prestrain levels studied show a constant rate of decrease, independently of the type of steel. For TRIP 700, this rate is $\approx 0.120 \pm 0.011$ (GPa/MPa), while for TRIP 800 it is lower: $\approx 0.103 \pm 0.014$ (GPa/MPa). Stage III entails the large-scale plastic deformation that occurs when the internal strength is exceeded. At this point a sudden drop is observed in the $E_T$.  

### 3.5. Tensile Tests: Unloading

The unloading behavior of the $E_T$ as related to the level of plastic prestrain for TRIP 700 and TRIP 800 is shown in Fig. 12. Again, $E_T$ decreases, but unlike during loading, there is no initial area of a linear response. The shape of the continuous decrease differs for both steels. For TRIP 700, at high stresses, there is a slow rate of decline, but as the stress decreases a progressive increase is observed in the rate. Finally, at low stresses, $E_T$ drops at much higher rates. These two areas are similar to the areas defined by Ghosh, and can be observed independently the amount of plastic prestrain, even in the initial state (0.001 strain). However, for TRIP 800, a more stable slope is defined throughout the unloading process, except for during the initial undeformed state, in which the two areas are clearly distinguishable.

The general shape of the unloading stress–strain curves is shown in Fig. 13. Due to the non-linearity of the stress–
strain curve, a ‘chord modulus’ ($E_c$) had to be calculated. In this study this parameter represents the slope between the points at maximum and minimum load of the stress–strain curve. The variation of $E_c$ with plastic prestrain is illustrated in Fig. 14, and shows a big drop for small prestrain. However, as the prestrain increases, the decrease in $E_c$ is less pronounced. In this case, the curves for TRIP 700 and TRIP 800 are very similar, and at a prestrain of 0.06 a mean value of approximately 155 GPa is found for both steels.

In Fig. 13, a straight line has been drawn from the maximum stress to the zero stress point to represent the slope of the apparent Young’s modulus, $E_A$, during the initial unloading of the undeformed steels (195 GPa for TRIP 700 and 205 GPa for TRIP 800). This shows an ideal elastic recovery after plastic deformation. For the two steels studied and for any prestrain level tested, the recovery was clearly greater than expected. Total recovery is then divided into linear and non-linear recovery. Linear recovery is related to the elastic strain and depends on the initial $E_A$ of each steel, while non-linear recovery represents the microplastic strain caused by the deformation process. The percentage of non-linear recovery out of the total recovery obtained during the unloading after plastic deformation is plotted in Fig. 15. This percentage increases quickly at the beginning and stabilizes at approximately 0.008 strain for both steels. Non-linear recovery comes to represent 20–24% of the total recovery. No significant variations in this case are found between the two steels and the percentages obtained agree with those of high strength steels.\(^{13}\)

4. Discussion

4.1. Inelastic Effects on Loading

For the two TRIP steels studied in the undeformed state, $E_T$ remains almost constant throughout the test until the beginning of macroplastic yielding. However, the evolution of the $E_T$ during loading for the previously predeformed samples (Fig. 10) shows some deviations from the expected linear behaviour. Firstly, at stresses well below those that occur at the beginning of yielding, a clear and constant drop in the $E_T$ occurs in both steels (Stage II). Secondly, in Stage I, in which the $E_T$ remains constant as stress increases, some differences were found between the two steels studied. For TRIP 700, increasing the amount of plastic prestrain led to a smooth decrease in the apparent Young’s modulus ($E_A$), whereas for TRIP 800 the apparent Young’s modulus appears to be unaffected (Fig. 11). In order to explain these two effects, some pertinent factors had to be investigated. The first of these was the transformation of metastable austenite to martensite with deformation. Since martensite has a lower Young’s modulus than a ferritic or austenitic matrix,\(^{26}\) the TRIP effect has been related to a decrease in Young’s modulus. As is shown in Fig. 8, the total amount of retained austenite transformed at the end of deformation process is small and quite similar for both steels (around 5% for TRIP 800 and 4% for TRIP 700) and it is important to consider that the changes of Young’s modulus reported for different steels after heat treatment to fully martensitic structure are smaller than 5%.\(^{31,32}\) Moreover, although both steels undergo the strain-induced transformation, the decrease in the $E_A$ is found only in TRIP 700. Within the TRIP 700, important changes in the $E_A$ are observed at low strains in which the percentage of martensite transformed is only around 1%. Therefore, although the strain-induced transformation of martensite could lead to an unquantified decrease in $E_A$ values, such small differences are not believed to explain why TRIP 700 and TRIP 800 exhibit different behaviours.

Furthermore, textural change during plastic deformation is usually linked to changes in Young’s modulus. However, small tensile strains, as were used here (0.04, 0.08) cause few changes to the orientation distribution function (ODF), and volume fractions calculated for certain special directions (mainly belonging to $\alpha$ and $\gamma$-fibres) remain practically unaffected.\(^{16,18}\) As the main changes in $E_A$ appear during these small deformations, no relationship with texture can be established.

The deviations found in the elastic behavior for both TRIP steels in the predeformed state can be explained by the appearance of an extra strain. This strain would not belong to an extra elastic strain but instead to a microplastic strain ($\varepsilon_{mp}$). As discussed in the introduction, $\varepsilon_{mp}$ is directly related to the dislocation structure.\(^{19-22}\) Since this extra strain can be produced by mobile or pinned dislocations, depending on the model. As can be derived from Eq. (2),

![Fig. 14. Variation of the chord modulus ($E_c$) during unloading as plastic prestrain increases for TRIP 700 and TRIP 800. The apparent Young’s modulus ($E_A$) values, calculated during loading of undeformed samples, are included to comparison.](image1)

![Fig. 15. Variation of the percentage in non-linear or plastic recovery from the total recovery during unloading with plastic prestrain for TRIP 700 and TRIP 800 steels.](image2)

\(^{1931}\) © 2005 ISIJ
\( \varepsilon_{\text{mp}} \) increases as the dislocation density (\( \rho \)) rises and there is also an increase in the dislocation displacement that depends on the dislocation line length (\( l \)) capable of bowing out or moving. It is worth noting that most of the materials tested had been pretrained. The plastic deformation produces an increase in the dislocation density in both steels\(^{24,29,33} \) and a decrease in the pinning points in the dislocations that are moved during deformation process, as for some time after deformation they are free from interstitial atoms.\(^{34,35} \) These conditions are suitable for microplastic strain to appear.

To move on, we will first focus our efforts on analyzing the effects found in Stage I. At low stresses, as the number of dislocations that can break away from their impurities or dislocation nodes is likely to be small, bowing out between pinning points would appear to be the prevailing mechanism behind \( \varepsilon_{\text{mp}} \). Although in both TRIP steels \( \rho \) increases with plastic strain, the ease with which extra deformation occurs when a small stress is applied is not the same for the two steels. In TRIP 700, the amount of \( \varepsilon_{\text{mp}} \) appears to be greater than in TRIP 800 due to the different ways in which their dislocation structures evolve. In TRIP 700, in which ferrite grains are prevalent and dislocations are distributed rather uniformly without forming tangles, there are a lower number of pinning points and the values of \( l \) are large. For TRIP 800, tangles of dislocations are created inside the ferrite grains from the beginning of strain and at higher strains (16\%) HDDWs are formed and even a cellular structure develops (Fig. 6). Here it is more difficult for dislocations to move at low stresses due to the high number of intersections with other dislocations.\(^{18} \) This results in the dislocation line \( l \) being small in length. The situation of dislocations in bainite sheaves seems to be very similar as they would be trapped in the borders of sub-units. In summary, while the amount of \( \varepsilon_{\text{mp}} \) generated in TRIP 700 can explain the decrease in observed \( E_A \), the absence of this additional strain led to constant values of \( E_A \) during the deformation process for TRIP 800.

The change observed in \( E_T \) curves during loading in Stage II can be attributed to the beginning of the release of dislocations from pinning points. As stress increases, more dislocations are activated, so the \( E_T \) continues to fall. The energy needed for this process seems to be similar for both steels as the phenomenon begins to take place at similar stresses. Differences in the slopes of the approximate linear decrease in the \( E_T \) under increasing stress (\(-0.120 \text{ GPa}/\text{MPa}\) for TRIP 700 vs. \(-0.103 \text{ for TRIP 800}\) could be related to the different densities of obstacles to the motion of mobile dislocations at a small scale. Since in TRIP 800 the dislocation structure in ferrite and bainite entails a higher density of obstacles, the \( \varepsilon_{\text{mp}} \) would be lower.

4.1. Inelastic Effects on Unloading

With regard to the unloading behaviour, the \( E_T \) was once again found to be nonlinear for the two steels studied (Fig. 12). This aspect can also be explained in terms of microplastic strain.\(^{13,23} \) During tensile plastic deformation many dislocations are created and pushed, which causes pile-ups, tangles and HDDWs to form. Many of these dislocations, especially those forming part of pile-ups, are repellant in character and only the stress applied keeps them very close to one another. When the stress drops, these dislocations go back, which produces an extra strain and thus a decrease in the \( E_T \) measured. At lower stress values, more non-pinned dislocations are activated and therefore the total extent of \( \varepsilon_{\text{mp}} \) is highest when the unloading stress tends towards zero. Here, again, as in loading, some differences have been found between the TRIP 700 and 800 steels. The \( E_T \) curves for the undeformed state (0.001 strain) are very similar for both steels, with a slow decrease at the beginning of unloading and a faster one when low stresses are achieved. However, the \( E_T \) curves for the deformed TRIP 700 maintain that shape, whereas in TRIP 800 this acceleration in the activation of mobile dislocation at low stresses disappears, the steel showing a more linear behaviour during unloading. In this case, the faster growth of tangles and dislocation forests in ferrite and bainite observed in TRIP 800 creates a higher number of intersections between dislocations. These act as nodal pinning points and their presence could mean that the number of dislocations activated at low stresses would be smaller than in TRIP 700. Therefore, the drop is less pronounced at low stresses.

Another aspect to be considered in the behaviour of \( E_T \) during unloading (Fig. 12) is the fact that the mean values of \( E_T \) decrease for the two TRIP steels as the samples are deformed. This effect is directly related to the values of the ‘chord modulus’ (\( E_c \)) in Fig. 14 or to the total amount of extra plastic recovery obtained when the sample is unloaded (Fig. 15). The \( E_T \) decreases at a quick rate at the beginning and more slowly once a certain amount of strain has been reached. This can be explained by the fact that the drop in the \( E_T \) is related to the backward motion of dislocations: as the material is deformed, \( \rho \) increases and more dislocations contribute to the extra strain; therefore, the \( E_T \) decreases. However, the rate of increase of free dislocations is high at the beginning but is reduced as plastic strain advances.\(^{18,33,36} \) Therefore, the fall in the \( E_T \) values and the increase in the percentage of non-linear or plastic recovery during unloading should stop at high degrees of deformation. These values are approximately 0.06 for TRIP 700 and 0.08 for TRIP 800.

4. Conclusion

The study of the microstructure of TRIP 700 and TRIP 800 steels reveals that TRIP 700 has a high volume fraction of ferrite (75\%), which is almost free from hard phases in grain boundaries; while in TRIP 800, ferrite, bainite and martensite-austenite form a complex multiphase structure. In TRIP 800, ferrite only represents 55\% of the phases presented and has a smaller grain size than in TRIP 700.

The differences found at the microstructural level affect the dislocation structure of ferrite in both steels. In TRIP 700, tensile deformation leads to a gradual increase in the number of dislocations in ferrite grains, these being, however, dislocations that are distributed rather uniformly, as dislocations forests and a cellular structure are not frequently observed at the centre of ferrite grains or in certain grain boundaries, even at the highest prestrains. In TRIP 800, tangles and HDDWs are observed at low strain levels in ferrite and bainite. At higher strains, even a cellular structure has been created inside the ferrite grains.
Dislocations in TRIP 800 seem to be more firmly trapped due to the high density of intersections between dislocations.

The study of the elastic stress–strain response of the two steels studied shows a decrease in the instantaneous tangent modulus ($E_T$) during loading and unloading as tensile deformation progresses. This decrease in both steels is attributed to the presence of microplastic strain. As microplastic strain depends on the presence of mobile or low-pinned dislocations, in accordance with the models established in the Introduction, the fact that the observed density of dislocation intersections is lower in TRIP 700 than in TRIP 800 is related to the higher decrease in the $E_T$ found in TRIP 700 during Stage I of loading.

In the prediction of springback in these and other reported cases, an understanding of the effect of strained microstructure on loading and unloading behaviour is needed in order to a better approximate these phenomena. Handbook values of Young’s moduli obtained from the undeformed state are not valid for predicting springback, as plastic recovery during unloading can produce a 20% increase over the recovery expected given those values.

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