On the Simultaneous Improving of Strength and Elongation in Dual Phase Steels via Cold Rolling

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Abstract: The ferrite-pearlite microstructure was cold-rolled to form dual phase (DP) steels, the percentage reduction of which varied. To do so, the steels were annealed in two steps and then the workpiece underwent water quenching. Accordingly, a decrease was observed in the average size of the ferrite grains, from above 15 µm to below 2 µm, subsequent to the thermomechanical processing. By an increase in the reduction percentage, the volume fraction of martensite grew. The balance between strength and elongation also improved nearly 3 times, equivalent to approximately 37,297 MPa% in DP in comparison to 11,501 MPa% in the ferrite-pearlite microstructure, even after 50% cold-rolling. Based on Hollomon and differential Crussard-Jaoul (DC–J) analyses, the DP steels under investigation deformed in two and three stages, respectively. The modified C–J (MC–J) analysis, however, revealed that the deformation process took place in four stages. The rate of strain hardening at the onset of the deformation process was rather high in all DP steels. The given rate increased once the size of the ferrite grains reduced; an increase in the volume fraction of martensite due to larger percentage of reduction also contributed to the higher rate of strain hardening. The observation of the fractured surfaces of the tensile specimens indicated ductile fracture of the studied DP steels.

Keywords: dual phase steel; strength-elongation balance; cold rolling; strain hardening behavior; fracture mechanism

1. Introduction

Dual phase (DP) steels, which are made up of hard martensite and soft ferrite grains, are formed by the common low alloy steels. These steels exhibit a high performance in terms of elongation and strength [1], thus receiving a lot of attention in particular in automotive industry as to the development of steel sheet [2]. Producing economical DP steel sheet, having enhanced mechanical qualities and grain-refined microstructures, is the staple challenging task in this regard.

If the yield and tensile strengths increase, the sheet thickness decreases, thereby lowering the weight. Simultaneously, the safety features at the time of car accidents would enhance. To build up the strength of the DP steels, either alloying elements are added or carbon content is increased [3–5]. However, the former is costly and the latter limits the ability of the steels to be welded. Thus, scholars have recently attempted to develop novel processing techniques to enhance the mechanical qualities. Accordingly, a number of small-scale processing routes [6–11], have been founded to enhance the mechanical features through refining the microstructure in the absence of any alloying elements. The studies indicated that fine ferrite-carbide aggregates could be employed as the initial microstructure while forming ultrafine grained (UFG) DP steels, where the size of the grains reduced and the balance between strength and elongation improved [9–11]. Nonetheless, the widespread applications of the
given techniques in the industry are accompanied with a number of limitations, including the size and shape of the sample, rapid rates of heating/cooling and great deformation at high temperatures.

With the use of the thermomechanical controlled processing (TMCP) technique, the industrial manufacturing of the given steel sheet seems feasible, where the whole sequence of processing, namely the reheating temperature, the rolling schedule and the cooling rate, is controlled [12]. That is to say, appropriate amounts of rolling and subsequently cooling are applied to refine the structure of the steels. Moreover, the nucleation sites of the ferrite are enhanced during cooling by the use of controlled rolling, in which the size of the grain reduces and the strain of the austenite lowers. In this technique, dislocations accumulate, annihilate and rearrange themselves and the grains recrystallize and grow, thereby evolving the microstructure of the sheet. To control the rolling, the inclusion of alloying elements, the level of reduction, the amount of strain, the temperature and the time intervals between reductions should be examined [13–15].

Although hot rolling and large deformation can effectively contribute to the formation of UFG in the DP microstructures, show two major limitations. First, this structure entails the application of costly machinery having huge capacity for loading. Second, it functions at elevated temperatures. Cold rolling, however, can facilitate the manufacturing of UFG DP sheets due to its ready-to-use equipment, where no particular facilities are required. Therefore, in the current investigation, cold rolling was employed to produce UFG DP structures, while enhancing both elongation and strength simultaneously. Moreover, two annealing steps were followed to lift the industrial restrictions. Furthermore, a number of mechanical qualities, including strain-hardening behavior, were considered which have not extensively been studied so far. Thus, the effects of microstructural factors on the deformation behavior of UFG DP steels were systematically examined through three separate analyses.

2. Materials and Methods

2.1. Materials

The AISI 5115 alloy was provided, which could be forged under hot conditions. The given alloy had the nominal chemical composition of 0.17 C-0.4 Si-1.15 Mn-0.95 Cr (wt. %). The specimens whose dimensions equaled 50 × 30 × 6 mm$^3$ were obtained from the hot-forged sheets.

2.2. Processing Procedure

The process of forming ferritic- martensitic DP steels out of a low carbon steel is shown in Figure 1a. The ferrite-pearlite steel was initially 50% and 80% cold-rolled, the rolling speed of which equaled 350 rpm, reducing nearly 0.5 mm per pass. Subsequently, annealing of the cold-formed sheets was done using an electric furnace (Temperature = 600 °C; Time = 20 min) followed by water quenching. Furthermore, to protect the sample from further decarbonizing while being heated, the cast iron swarf was used. This process yielded ferrite/carbide aggregates, selected as the initial structure prior to the last step. The specimens were eventually heated to the intercritical region and kept at 820 °C for 10 min, preceded by water quenching. The dilatometer equipment (DAMA Pajouh Arvin Co., Tehran, Iran) helped to specify the lower and upper critical temperatures that is, $A_{c1}$ and $A_{c3}$ temperatures, respectively (Figure 1b). During the dilatometry test the solid cylindrical specimen with the diameter of 2.4 mm and length of 10 mm was heated to 950 °C at a rate of 1 °C/s, held at this temperature for 20 min, followed by cooling to the room temperature. The dilatation curve consisted of two main points corresponding to the two intercritical temperatures.
which was based on the Hollomon equation [18]:

\[ \theta \]

where \( \theta \) is the misorientation angle, \( u \) is the unit length and \( b \) is the Burgers vector magnitude.

The Santam-STM-50 machine (SANTAM Engineering Design Co. Ltd., Tehran, Iran) helped to do the uniaxial tensile test, performed based on the ASTM-E8 standard on the UFG DP steels (gauge length = 15 mm; width = 5 mm; thickness = 1.2 mm). Moreover, the speed of crosshead remained constant, equivalent to 1 mm/min. To delineate the strain hardening behavior of the given steels, the following widely used mathematical analyses were applied:

(i) Hollomon analysis [18]:

\[ n = \frac{d(ln\sigma)}{d(ln\varepsilon)} \]

which was based on the Hollomon equation [18]:

\[ \sigma = k\varepsilon^n \]  

(ii) Differential Crussard–Jaoul (DC–J) analysis [19,20]:

\[ \ln(d\sigma/d\varepsilon) = (n - 1)\ln\sigma + \ln(k\varepsilon) \]  

2.3. Characterizations

JEOL-JSM-840A scanning electron microscopy (SEM, JEOL Ltd., Tokyo, Japan) and Union Versamet-2 optical microscopy (OM, Union Optical Co., Tokyo, Japan) did the microstructural observations. To measure the grain size and volume fraction of the phases, microstructural image processing (MIP) software (MIP4Student, Nahamin Pardazan Asia, Mashhad, Iran), working on the basis of linear intercept method [16], was employed.

Electron backscattered diffraction (EBSD) analyses were done by a field emission SEM (JSM7001F, JEOL Ltd., Tokyo, Japan) equipped with an EBSD detector combined with the TSLOIM analysis software. The accelerating voltage in the SEM analysis was 15 kV. To determine the grain size, dislocation density and texture, the structural studies including image quality (IQ), inverse pole figure (IPF), grain average misorientation (GAM), Kernel average misorientation (KAM) and geometrically necessary dislocations (GNDs) distribution maps was performed using EBSD analysis. The lattice curvatures which reflect from the local changes in the lattice orientation can be considered to calculate GND densities, according to the method of ref. [17]. The KAM values which are directly regained from EBSD data can be defined as a measure for the local misorientations and related to the GND density (\( \rho_{\text{GND}} \)) as follow [17]:

\[ \rho_{\text{GND}} = 2\theta/ub \]  

\( \theta \) is the misorientation angle, \( u \) is the unit length and \( b \) is the Burgers vector magnitude.

Figure 1. (a) Thermomechanical processing developed to produce ultrafine grained–dual phase (UFG–DP) steels and (b) dilatometry test experiment upon continuous heating by the rate of 1 °C/s. \( Ac_1 \): start and \( Ac_3 \): finish temperature of austenite formation during heating; WQ: water quenching; CR: cold-rolling.
which was based on the Ludwik equation [21]:

\[ \sigma = \sigma_0 + k\epsilon^n. \]  

(iii) Modified C–J (MC–J) analysis [22]:

\[ \ln(\frac{\sigma}{\epsilon}) = (1 - m)\ln\sigma - \ln(\sigma_0/k'), \]  

which was based on the Swift equation [23]:

\[ \epsilon = \epsilon_0 + k'\sigma^m, \]

where \( \epsilon \) and \( \sigma \) are the true strain and stress, respectively, \( n \) and \( 1/m \) are the exponents of strain hardening, \( k \) and \( 1/k' \) are the strength coefficients and \( \sigma_0 \) and \( \epsilon_0 \) are material constants. As specified in the Hollomon analysis as to the \( \ln\sigma - \ln\epsilon \) plot, \( n \)-value can be derived from the slope of the line. The slope of the curve \( \ln(\frac{\sigma}{\epsilon}) \) vs. \( \ln\epsilon \) in the DC–J analysis is equal to \( (n - 1) \). According to the MC–J analysis, the slope of the plot \( \ln(\frac{\sigma}{\epsilon}) \) – \( \ln\sigma \) yields \( (1 - m) \), where \( m \) is the inverse of the strain-hardening exponent. Subsequent to running the tensile tests, the examination of the fracture surfaces of the specimens was done with the help of SEM.

3. Results and Discussion

3.1. Microstructural Evaluations

3.1.1. Optical Microscopy and Scanning Electron Microscopy

Figure 2 indicates the as-received state of the ferrite-pearlite microstructure, where the volume fraction and the average grain size of the ferrite was nearly 70% and 15 µm, respectively. Figure 3 shows the microstructure of the 50% and 80% cold-formed steels. A wavy pattern was observed in the ferrite matrix, curved along the pearlite colonies and stretched in the rolling direction. As can be seen, the pearlite colonies were deflected toward the cold-forming direction. That is to say, ferrite was deformed together with pearlite, quite in line with previously reported findings [24,25]. The microstructural properties of the cold-formed duplex structure included high density of ferrite/pearlite interfaces that could be considered nucleation sites for austenite formation. Any reduction in the rolling process would enlarge the ferrite/pearlite interfaces, thus potentially increasing the density of the desired nucleation site and the volume fraction of austenite (martensite at room temperature). The UFG ferrite matrix and the distributed nearly spheroidized cementite particles with average size of about 400 and 90 nm, respectively, formed after annealing at 600 °C for 20 min followed by water quenching, is shown in Figure 4.

![Figure 2. Optical microscopy (OM) micrograph of the as-received steel.](image-url)
As indicated in Figure 5 subsequent to the intercritical annealing of the 50% and 80% cold-formed steels, the DP structures were observed. The SEM micrographs indicated a uniform phase distributed throughout ferrite (shown by F in the figures), filled with martensite particles (shown by M in the figures). The volume fraction of martensite ($V_m$) and the average grain size of ferrite ($d_f$) in the DP microstructures under investigation are summarized in Table 1. Any further of the rolling reduction would enhance $V_m$ (see Figure 5 and Table 1). The increase in the amount of rolling reduction would

![Figure 3](image-url)  
Figure 3. (a,b) OM and (c,d) scanning electron microscopy (SEM) micrograph of the (a,c) 50% and (b,d) 80% cold-rolled ferrite-pearlite initial microstructure; F: ferrite; P: pearlite.

![Figure 4](image-url)  
Figure 4. SEM micrographs of UFG ferrite/nanocarbide aggregates formed by annealing of 80% cold-rolled ferrite-pearlite microstructure at 600 °C for 20 min followed by water quenching. F: ferrite; C: carbide.

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result in the larger volume fraction of austenite, transforming to martensite during the next water quenching process. This finding was consistent with the previously reported findings, where increase in the rolling reduction resulted in the greater $V_m$ \cite{15,26}. That is to say, the nucleation sites and stored energy required for the formation of austenite in the microstructures before the last intercritical annealing would enhance with any further increase in the rolling process, thus increasing the driving force for the phase transformation and $V_m$.

![SEM micrographs of UFG ferrite/nanocarbide aggregates formed by annealing of 80% cold-rolled ferrite-pearlite microstructure at 600 °C for 20 min followed by water quenching. F: ferrite; C: carbide.](image)

**Figure 4.** SEM micrographs of UFG ferrite/nanocarbide aggregates formed by annealing of 80% cold-rolled ferrite-pearlite microstructure at 600 °C for 20 min followed by water quenching. F: ferrite; C: carbide.

The declining pattern observed in $d_f$ once rolling increased was in line with the previous studies \cite{13,15}. On the one hand, with further amount of the rolling reduction, a finer structure was found in the DP sheets due to the substantial deformation previously occurred in the structure. On the other hand, any further increase in $V_m$ stunted the growth of the ferrite grains, thereby lowering $d_f$ \cite{27}.

### Table 1. Microstructural and mechanical characteristics of the as-received and DP steels under different conditions.

| Sample      | $d_f$ (µm) | $V_m$ (%) | YS (MPa) | UTS (MPa) | UE (%) | TE (%) | UTSTUE (MPa%) | Toughness (MPa) | YS/UTS |
|-------------|------------|-----------|-----------|-----------|--------|--------|----------------|-----------------|--------|
| as-received | 15.21      | -         | 517 ± 13  | 742 ± 10  | 15.5 ± 0.5 | 26.5 ± 0.5 | 11,501         | 90 ± 5          | -      |
| 50% CR      | 2.83       | 84        | 895 ± 10  | 1510 ± 15 | 24.7 ± 0.3 | 36.2 ± 0.3 | 37,297         | 260 ± 5         | 0.58 ± 0.01 |
| 80% CR      | 1.67       | 90        | 1050 ± 10 | 1608 ± 15 | 20.7 ± 0.3 | 32.5 ± 0.5 | 33,285         | 210 ± 5         | 0.59 ± 0.02 |

CR: cold-rolled; $d_f$: ferrite grain size; $V_m$: martensite volume fraction; YS: 0.2% offset yield strength; UTS: ultimate tensile strength; UE: uniform elongation; TE: total elongation.

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#### 3.1.2. Electron Backscattered Diffraction

The IQ, IPF, GAM, KAM and GNDs distribution maps of 50% and 80% cold-formed DP steels are given in Figure 6. Martensitic phase in IQ map seems darker than ferrite phases because of its higher
lattice distortion. The boundaries with misorientation above $15^\circ$ that is, high angle grain boundaries (HAGBs) are represented by blue lines. Low angle grain boundaries (LAGBs) (boundaries with misorientation between $2^\circ$ and $5^\circ$) are also represented by red lines. Figure 6a,f show that the resultant microstructures after intercritical annealing, specifically for the 50% cold-formed DP steel, contained larger ferrite grains compared to martensite islands and there are almost no LAGBs in the ferrite grains. In the cold-rolled structures, recovery and continuous recrystallization of ferrite during intercritical annealing resulted in LAGBs and HAGBs, respectively. The microstructural parameters of HAGBs fraction, average misorientation angle and average boundary spacing including all boundaries were quantitatively indicated from the EBSD analysis and are illustrated as a function of the rolling reduction percentage in Table 2. As can be seen, increasing rolling reduction resulted in decreasing fraction of LAGBs and increasing fraction of HAGBs. The higher fraction of HAGBs and average misorientation angle in the 80% cold-formed DP steel, show lower growth rate and correspond to the lower boundary spacing and finer grain size. The orientation of grains is shown by IPF maps in Figure 6b,g. The stereographic triangles show correspondence between the colors and the crystal orientations. Based on these figures, it can be concluded that these samples have random texture distributions.

The GAM map can be used to analysis the plastic deformation because it characterizes the plastic strain distribution [28]. The KAM is generally used to determine the GNDs density since it defines as the average misorientation of a given point with all its neighbors in EBSD results. In Figure 6c,h the ferrite grains was indicate by blue color since this phase has the lowest value of the GAM while the martensite phase with highest value of the GAM was indicated by green color. This difference between martensite and ferrite in GAM value can be attributed to their different mechanical properties.

Indeed, the deformation of ferrite phase begins at lower strains and sooner than martensite which it increase the GAM values in the martensite phase. Moreover Figure 6c,h implies that all grains in the 80% cold-formed DP steel experience higher plastic strain compared to that of the 50% one. In the other words, as the rolling reduction increases, the stored plastic strains increases. The KAM maps in Figure 6d,i show that at the ferrite/martensite interface, the KAM value is higher than the grain’s inside. This can be linked to the GNDs at the interface (Figure 6e,j) which are mainly formed to keep the lattice continuity during austenite transformation [9,29].
with the large concentration of geometrically necessary dislocation (GNDs) led to this continuous pattern in the yielding behavior, which could be created during quenching from the austenite region (Figure 6) [30,31].

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The newlformed DP steels could potentially substitute the commercial grades in the automobile body manufacturing. Indeed, the considerable internal strains together with the large concentration of geometrically necessary dislocation (GNDs) led to this continuous pattern in the yielding behavior, which could be created during quenching from the austenite region (Figure 6) [30,31].

| Sample | HAGBs Fraction (%) | Average Misorientation Angle (°) | Average Boundary Spacing (µm) |
|--------|--------------------|---------------------------------|------------------------------|
| 50% CR | 56                 | 29.4                            | 3.14                         |
| 80% CR | 64                 | 32.4                            | 1.98                         |

3.2. Tensile Properties

A discontinuous pattern was found as to the yielding behavior of the as-received sample. However, the yielding behavior gradually changed to a usual continuous pattern subsequent to the intercritical annealing, which is typical of DP steels (see Figure 7). Indeed, the considerable internal strains together with the large concentration of geometrically necessary dislocation (GNDs) led to this continuous pattern in the yielding behavior, which could be created during quenching from the austenite region (Figure 6) [30,31].

![Figure 6.](image1.png)

**Figure 6.** (a,f) Image quality (IQ), (b,g) inverse pole figure (IPF) (c,h) grain average misorientation (GAM) (d,i) Kernel average misorientation (KAM) and (e,j) geometrically necessary dislocations (GNDs) distribution maps of (a–e) 50% and (f–j) 80% cold-formed DP structures.

**Table 2.** Microstructural parameters determined from the electron backscattered diffraction (EBSD) data for different DP steels.

![Figure 7.](image2.png)

**Figure 7.** (a) Engineering and (b) true stress-strain curves of the as-received and cold-formed DP structures.
Further increase in the rolling reduction increased the yield strength (YS) and ultimate tensile strength (UTS), as can be observed in Figure 7 and Table 1, which can chiefly be due to the smaller \( d_f \), larger \( V_m \). Next to intercritical annealing, the YS and UTS of the 80% cold-formed sample (Temperature = 1093 K (820 °C); Time = 10 min) reached nearly 1050 and 1608 MPa, respectively. That is to say, the behavior of the sample improved more than 100% in comparison to the behavior observed in the as-received sample (YS: 517 and UTS: 742 MPa). The harder martensite, present in the sample, yielded greater strength \[32\]. Furthermore, the grains refined in the DP steels greatly contributed to the enhancement of strength and elongation at the same time (Table 1). Nakada et al. \[26\] found that strain localization in the ferrite grains did not propagate to the adjacent ferrite since the grains were enclosed by the martensite grains having a chainlike, networked structure. This could probably help to preserve elongation even upon enormous levels of strength. Calcagnotto et al. \[9\] could well elaborate on this mechanism, stating that any refinement in the grains could lower the difference observed in the tensile features of ferrite and martensite phases, thereby yielding greater interface cohesion, delaying the plastic instability and reaching more uniform elongation (UE).

The strength-elongation balance (UTS \( \times \) UE) and toughness (the area under the engineering stress-strain curve which is an indication of the required energy to break the material) can exhibit the energy absorption capability, chief parameters in vehicle crashworthiness. The values obtained from the DP steels were higher than the values derived from the as-received sample, which could be attributed to the considerable strength and elongation in the sample (see Table 1). The 50% cold-formed sample enjoyed the greatest UTS \( \times \) UE balance and toughness, having considerable advantages for industrial manufacturing. The UTS of the 50% cold-formed DP steel was about 7% lower than that of the 80% one. In contrast, the UE of the 50% cold-formed DP steel was about 19% higher than that of the 80% one. This is nearly consistent with the 12% higher value of strength-elongation balance in the 50% cold-rolled sample in comparison with the 80% one (Table 1). The newly formed DP steels could potentially substitute the commercial grades in the automobile body manufacturing.

### 3.3. Strain Hardening Behavior

Figures 8 and 9a,b indicate the findings as to the application of the Hollomon analysis to the true stress-strain data of the steels under investigation. The as-received sample behaved linearly; however, the relationship between \( \ln \sigma \) and \( \ln \varepsilon \) was found to be nonlinear during the two-stage strain hardening which has so far been suggested for the DP steels, where a specific structural deformation took place at each step \[33–35\]. The strain hardening exponent of the DP steels was found to be larger than that of the ferrite-pearlite. Shin and Park \[36\] and Calcagnotto et al. \[9\] also found the use of the intercritical annealing treatment to be effective in altering the cementite phase to the martensite one, thus considerably enhancing the amount of strain hardening. Based on the results released so far \[37,38\], any increase in the \( V_m \) would result in smaller work hardening exponent in the DP steels. Furthermore, the strain hardening behavior in the DP steels was under the control of \( d_f \). That is, reduction in the \( d_f \) lowered the \( n \)-value of the DP steels \[9\]. Thus, the variations observed in \( n \) were the main reason for the changes made in the \( V_m \) and \( d_f \) when rolling increased. Although the \( \ln \sigma - \ln \varepsilon \) plots shown in Figure 9a,b looks very good and provide reasonable values of \( n \) between 0.1 to 0.2 (see Table 3), which are very typical for steels but the Hollomon analysis is not able to show a correct trend for the variations of \( n \) with \( V_m \) and \( d_f \).

| Sample  | \( n_I \) | \( n_{II} \) | \( \varepsilon_{tr} \) |
|---------|---------|-------------|------------------|
| 50% CR  | 0.111   | 0.160       | 0.063            |
| 80% CR  | 0.118   | 0.195       | 0.040            |

Table 3. The variations of strain hardening exponent (\( n \)) and transition strain (\( \varepsilon_{tr} \)) between the deformation stages of the investigated DP steels based on the Hollomon analysis.
The examination of the transition strain between deformation stages ($\epsilon_{tr}$) was also of paramount importance. Mazinani and Poole [39] found that by lowering the amount of carbon ($C_m$) of martensite the strength decreased, while the plasticity enhanced. That is, the strength of martensite plus $C_m$ reduced with an increase in $V_m$. Hence, as stated in Jiang et al. [34], with further reduction of rolling and enhancement of $V_m$, martensite plastic deformation would initiate sooner and $\epsilon_{tr}$ would tend to smaller strains.

The DC–J model rested on the Hollomon model, in which a new factor, that is, $\sigma_0$ in Equation (5), corresponding to the yield strength of steel, increased the reliability of the model as to the plastic region. The $\ln(\sigma/d\epsilon) - \ln\epsilon$ plots of the steels under investigation are shown in Figure 9c,d. Table 4 also summarizes the findings related to the DC–J analysis, indicating that the DP steels followed three steps to be deformed. This could be due to various mechanical features appeared in each phase in the DP steels once different amounts of strains were applied. The given deformation steps involved the formation of GNDs [40], the deformation of the ferrite grains with no trace of martensite plastic deformation [41,42] and the plastic deformation of both phases [34,43], respectively. The slope of the stages also rose as the deformation process went on. The measured values in Table 4 are in the range of that reported in literature [44]. Nonetheless, once investigating the 80% cold-formed DP steel through DC–J analysis, negative values were found for $n$ in the third stage of strain hardening, which seemed unjustifiable. That is to say, the DC–J analysis could not differentiate the strain-hardening pattern of the DP steels whose microstructure varied in terms of coarseness.

Figure 9e,f demonstrate the $\ln(\sigma/d\epsilon)$ vs. $\ln\sigma$ plots of the formed DP steels in various thermomechanical conditions. Table 5 also presents the findings as to the MC–J analysis. Here, contrary to the Hollomon and DC–J analyses, it was found that the DP steels followed four steps of strain hardening, namely the motion of the GNDs and statically stored dislocations [40,45], the occurrence of the ferrite matrix plastic deformation while retaining martensite elasticity, the ferrite deformation constrained by martensite islands [41,42] and eventually plastic deformation of both phases [46–48], respectively. The ability for strain hardening decreased as the process proceeded, while the slope increased. The plastic deformation of ferrite constantly entails the existence of hard martensite. Figure 10 show SEM micrographs on the perpendicular plane to the fractured tensile specimen surface (rolling direction-normal direction (RD-ND) section) of the 50% cold-formed DP steel. Both phases elongated approximately along the loading direction. If the amount of stress transfer from ferrite to martensite is large enough, martensite plastic deformation occurs.

The higher the $1/m$ value the greater the ability for strain hardening. Overall, the findings revealed that $V_m$ and $d$ trued affected the deformation patterns of the DP steels. Furthermore, $V_m$ evidently improved once $\epsilon_{tr}$ decreased. Son et al. [49] and Colla et al. [50] also found that among the given analyses only the MC–J was easily influenced by the microstructure, indicating that the DP steels deformed during multiple steps.
The DC-J model rested on the Hollomon model, in which a new factor, that is, \( \sigma_0 \) in Equation (5), corresponding to the yield strength of steel, increased the reliability of the model as to the plastic region. The \( \ln(d\sigma/d\varepsilon) - \ln\varepsilon \) plots of the steels under investigation are shown in Figure 9c,d. Table 4 also summarizes the findings related to the DC-J analysis, indicating that the DP steels followed three steps to be deformed. This could be due to various mechanical features appeared in each phase in the DP steels once different amounts of strains were applied. The given deformation steps are referred to as I, II, and III.

**Figure 9.** (a,b) the Hollomon analysis, (c,d) the differential Crussard-Jaoul (DC–J) analysis and (e,f) the modified C–J (MC–J) analysis of the cold-formed DP steels. \( \varepsilon_{tr} \) is the transition strain between deformation stages.

**Table 4.** The variations of strain hardening exponent \( (n) \) and transition strain \( (\varepsilon_{tr}) \) between the deformation stages of the investigated DP steels based on the DC–J analysis.

| Sample | \( n_I \) | \( n_{II} \) | \( n_{III} \) | \( \varepsilon_{tr(I)} \) | \( \varepsilon_{tr(II)} \) |
|--------|-----------|-------------|-------------|----------------|----------------|
| 50% CR  | 0.997     | 0.967       | 0.055       | 0.112          |
| 80% CR  | 0.917     | −0.852      | 0.050       | 0.131          |
an increase in the amount of martensite, when rolling reduction increased, resulted in greater strain hardening once the strains were not large. The considerable internal strains created by the quenching from austenite region, the high density of GNDs resided in the ferrite/martensite interfaces (Figure 6) and eventually the difference in the plastic behavior of ferrite and martensite all increased the rate of strain hardening once the strains were not large.

| Sample | $1/m_I$ | $1/m_{II}$ | $1/m_{III}$ | $1/m_{IV}$ | $\varepsilon_{tr(I)}$ | $\varepsilon_{tr(II)}$ | $\varepsilon_{tr(III)}$ |
|--------|---------|------------|-------------|------------|-----------------|-----------------|-----------------|
| 50% CR | 0.987   | 0.951      | 0.866       | 0.501      | 0.042           | 0.079           | 0.132           |
| 80% CR | 0.784   | 0.406      | 0.201       | 0.061      | 0.038           | 0.061           | 0.119           |

**Figure 10.** (a) Low and (b) high magnification SEM micrographs on the perpendicular plane to the fractured tensile specimen surface (RD-ND section) of the 50% cold-formed DP steel.

The level of fluctuations occurring in strain hardening rate against the true strain over the entire uniform strain range in the DP steels under investigation is indicated in Figure 11. The inset of Figure 11 shows the similar trend for the 50% and 80% cold-formed samples. As previously stated [9,37], the DP steels initially demonstrated a high level of strain hardening. The considerable internal strains created by the quenching from austenite region, the high density of GNDs resided in the ferrite/martensite interfaces (Figure 6) and eventually the difference in the plastic behavior of ferrite and martensite all increased the rate of strain hardening once the strains were not large.

**Figure 11.** The variation of strain hardening rate as a function of true strain for 50% and 80% cold-rolled DP steels.

Based on the microstructural properties summarized in Table 1, a decrease in grain sizes and an increase in the amount of martensite, when rolling reduction increased, resulted in greater strain hardening in the DP steels at the onset of deformation. This was in line with previous findings.
hardening in the DP steels at the onset of deformation. This was in line with previous findings, where an increase in the initial strain hardening observed with reducing the grain size and boosting the amount of martensite [37]. The initial density of dislocations would enhance when \( d_f \) decreased and \( V_m \) increased, thus boosting the initial strain hardening because of dislocations pile up [41,46].

3.4. Fracture Mechanism

To find out the fracture mechanism, SEM images were used to observe the fracture surfaces of the DP steels (see Figure 12). The fracture surfaces exhibited homogeneous dimples without any cleavage properties, indicating ductility in the DP steels [9]. This ductility occurred within three consecutive steps, namely void nucleation, void growth and void coalescence, resulting in a fracture surface with uniform dimples. The microvoid nucleation resulted from the martensite breaking or the ferrite-martensite interface decohesion. These microvoids, which could be raised in the ferrite matrix with greater ductility, led to ductile fracture surfaces. Moreover, these nucleation mechanisms were largely dependent upon the \( d_f \), \( V_m \) and \( C_m \). As a brittle phase, martensite mostly inclined towards cleavage fracturing. However, the deformation of martensite may enhance when \( C_m \) reduces [39,51]. Conversely, Calcagnotto et al. [9] demonstrated that the rate of cleavage fracturing might decrease when the grain size reduced as the number of built-up of dislocations within ferrite grains and consequently shear stress applying to martensite particles reduced. That is, the small \( d_f \) and \( C_m \) corresponds to slighter difference in strength between ferrite and martensite, which may lead to reduced strain localization at interfaces, creating lower densities of void formation during straining, thereby boosting the strength-elongation balance or energy absorption capability, as well as ductile patterns of fractures [52].

![Fracture surfaces of DP steels](image.png)

**Figure 12.** Fracture surfaces of (a,b) 50% and (c,d) 80% cold-rolled DP steels.

4. Conclusions

The dual-phase steels were made by means of thermomechanical processing while taking into account the industrial constraints. The process included low cold-forming mechanism, followed
by two annealing stages. The DP steels were formed with 50% and 80% cold-rolling. The tensile properties of the steels were remarkable (~1510 and 1608 MPa tensile strength and ~25 and 21% uniform elongation, respectively) in comparison with the as-received condition (~742 MPa tensile strength and ~16% uniform elongation). Moreover, three strain-hardening models were employed to examine the strain-hardening behavior of the DP steels, namely Hollomon, DC–J and MC–J. The results indicated a better fit of the MC–J analysis to the experimental data, revealing that the steels deformed in four consecutive steps. The initial level of strain hardening rose when the volume fraction of martensite enhanced and the grain size of ferrite reduced. In addition, the plastic deformation of the hard martensite regions increased when the difference between the strength of ferrite and martensite reduced due to a decrease in the grain size of ferrite and in the amount of carbon of martensite, thus resulting in ductility in fracture surfaces.

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