1. Introduction

Dual phase steels are known since 1963 for their optimum combination of strength and ductility. For this reason, in recent years these steels have met a growing interest in the applications of the automotive industries, where the high strength is important to reduce the weight of the components and the good formability can improve the quality and the production rate of the final cold forming operations.

These steels are obtained by quenching steel sheet from intercritical temperature to produce their microstructure that consists in a dispersion of a hard second phase in a matrix of ferrite. The presence of a second phase (usually martensite) is required to obtain the typical mechanical dual-phase behavior: a continuous yielding, a low yield to tensile-strength ratio, a high uniform total elongation and a high work-hardening rate.

The physical and mechanical properties of the metals and of their alloys are closely related to the possible presence of preferential crystallographic textures produced during the manufacturing process that might induce the anisotropy in their mechanical and physical features. From the mechanical point of view, in the case of the dual phase steels the attention is particularly focused on the yield strength and formability. One of the more rapid and reliable techniques to determine the formability and the mechanical anisotropy is the determination of the plastic strain ratio through tensile tests along the main directions featuring the rolled sheet. The relation between the formability and the induced anisotropy has been deeply studied in the low strength dual phase steels for the deep drawing operation, but less attention has been paid about this aspect for high strength dual phase steels characterized by a high value of the martensite volume fraction.

Although the stretch formability of dual phase steels is excellent in relation to their strength level, the deep drawability tends to be less impressive for two reasons. Firstly, it is difficult to develop the appropriate crystallographic texture necessary for high normal anisotropy because of the alloying additions which are frequently used. Secondly, some authors have shown that even when the texture is suitable, the hard martensite phase perturbs the ferrite deformation. The formability depends mainly on the conditions of the ferritic component. The martensite is so hard that it does not deform plastically and therefore, in first approximation, its texture should not be important to the plastic anisotropy directly. However, the presence of this hard phase should modify the plastic behavior and anisotropy of the softer ferrite.

The rolling texture of the body centered metals, as ferritic steels, becomes steadily sharper as the imposed plastic deformation increases. The intensity of each component of texture influences the $r$-coefficient and the mechanical anisotropy. The presence of the components featured by the planes $\{111\}$ parallel to the rolling plane increases the $r$-coefficient and the components $\{100\}$ decrease this value.

The aim of this work is to understand the different mechanical behavior of two kinds of dual phase steels with different high martensite content. The present research has been articulated through these following different steps:
1. micro-structural characterization of two grades of the dual phase steels featured by a difference of about 200 MPa of the yield strength;
2. measurement of $r$-coefficient of the cold rolled steel sheets along the three characteristic directions ($0^\circ$, $45^\circ$, $90^\circ$).
and 90° with the respect to the rolling direction) to determine the Lankford coefficient \( r_{\text{m}} \) and the planar anisotropy \( r_{\text{ww}} \); 
3. determination of the textures and their relation with the mechanical properties before and after the intercritical heat treatment; 
4. definition of the relation among the relative intensity of texture components and the value of the Lankford coefficient.

The textures created during the rolling of the sheets depend on the crystal structure as well as on the induced reduction and on the rate of its application. Moreover, the heat treatment influences the final texture induced in the rolled sheet as a function of the former parent texture. So, the texture was measured before and after the intercritical heat treatment.

2. Experimental Procedure

Two different steel grades have been used for the analysis (Table 1). The two steels are featured by an average minimum tensile strength of 1 000 MPa (Steel A) and 800 MPa (Steel B).

The steel sheets have undergone the same manufacturing process: hot rolled plastic deformation down to 3.35 mm thickness, then they were cold rolled in two passes, the first one to 1.8 mm and the second one to the final thickness 0.7 mm. The final annealing was performed in a continuous line at the temperature of about 800°C. In this thermal range the steels modify their microstructure and austenite compares at the ferritic grain boundaries. The hot band was quenched by a cold gas and then by water. The cooling speed has been set at 1 000°C/s. Finally, the sheets are also annealed up to 300°C and cooled through an inert gas atmosphere.

The tested specimens were taken before and after the heat treatment and they were prepared by mechanical polishing and chemical metallographic etching (Nital 2% 10 s). These steels were analyzed along the rolling and the transverse direction, to evaluate the microstructural differences. This was carried out through a light microscope and SEM. The measures of the volume fraction of the present phases was obtained on four different images realized by the light microscope and taken on each sample by a software for image analysis (Image ProPlus®).

The tensile tests were carried out by the test machine MTS Alliance RT/100®. The tests were performed on specimens of Steels A and B, obtained from the steel sheet according with ASTM E8, before and after the heat treatment in the direction 0°, 45°, 90° respect to the rolling one. To obtain the \( r \)-values the tests were stopped at 2% and 5% strain. The collected mechanical properties were: Young Modulus (YM), Yield Stress (YS), Strain Hardening Exponent (n), Tensile Strength (TS), Uniform Elongation (UE), Total Elongation (TE), reduction in area after tensile fracture (N).

These properties were obtained in according to ASTM E8 and ASTM E111 for the Young Modulus.

The tests for the specimens before the heat treatment were carried out only at 1 mm/min speed whereas for the specimens after the heat treatment the tests were performed at the speeds of 1 mm/min, 25 mm/min and 36 mm/min in order to measure the coefficient m (rate sensitivity). The plastic strain ratio or \( r \)-coefficient:

\[
 r = \frac{E_w}{E_t} = \frac{\ln(w/w_0)}{\ln(t/t_0)} \tag{1}
\]

was determined as prescribed in ASTM E517 from tensile specimens sampled along 0°, 45°, 90° respect to the rolling direction. To evaluate the possible dependence of \( r \)-coefficient on the applied deformation rates the mechanical tensile tests were performed at 1 mm/min, 25 mm/min and 36 mm/min. The average \( r \)-coefficient was calculated using the following equation:

\[
 r_{\text{m}} = (r_0 + 2r_{45} + r_{90})/4 \tag{2}
\]

The possibility and the characteristics of the earing phenomenon were evaluated through the determination of the coefficient of planar anisotropy which is defined as

\[
 \Delta r = (r_0 - 2r_{45} + r_{90})/2 \tag{3}
\]

The characterization was fulfilled by the fractography procedure (according to ASM\(^{13}\)) and by the Vickers hardness tests. The Vickers hardness tests were carried out with a load of 300 gf by the penetration of the diamond tip and in agreement with the standard ASTM E384.

Finally, FLD tests were performed. The blank width are 75 mm, 80 mm, 100 mm, 120 mm, 160 mm, 190 mm, 260 mm. The length are 260 mm (in rolling direction).

The crystallographic analyses were performed by two techniques: with the Schultz type X-ray diffraction with 5 degrees steps and with the EBSD (Electron Back Scattering Diffraction). In the first one, the used texture goniometer is a X’Pert Philips®. At least three pole figures were needed, to determine all texture characteristics. A powder spectrum was realized in order to establish the experimental accessible pole figures. The spectrum (carried out through Cu radiation) revealed the \{200\}, \{110\}, \{211\} and \{310\} diffraction peaks of ferrite. It should also show the martensite diffraction peaks, but it has not been possible to note the peaks corresponding to the martensite lattice. This phenomenon is due to the little difference of the diffraction peaks related to the ferrite and to the martensite. The quantity of carbon present in the studied steel is low enough to make distortion of the martensite lattice is not relevant and thus the phase diffraction peaks of the ferrite and of the martensite overlaps. The powder (random specimen) was needed also to correct the defocusing effect and to provide a reference intensity level to compare the intensities produced by the textured materials.

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Table 1. Chemical composition (wt%) of specimens.

| (% wt) | C | Si | Mn | P | S | N | Cr | Ni | Cu | Mo | Al | Nb | V | B |
|-------|---|----|----|---|---|---|----|----|----|----|----|----|---|---|
| Steel A | 0.154 | 0.51 | 1.5 | 0.009 | 0.002 | 0.0044 | 0.03 | 0.05 | 0.01 | 0.02 | 0.046 | 0.018 | 0.007 | 0.0004 |
| Steel B | 0.134 | 0.19 | 1.49 | 0.017 | 0.004 | 0.0024 | 0.03 | 0.04 | 0.01 | 0.05 | 0.015 | 0.01 | 0.0002 | 0.0002 |
For all specimens the background level was measured and it was subtracted from the revealed intensity of the diffraction. Measured pole figures corrected from defocusing and background are incomplete pole figures.\(^{31}\) The orientation distribution function (ODF) was obtained through the combination of the incomplete pole figures \{200\}, \{110\}, \{211\}, \{310\}. This analysis was done on samples strained at 2\% and 5\% and on unstrained samples before and after the heat treatment.

The Inca Crystal EBSD system was mounted on Oxford Instrument scanning electron microscope (SEM) operated at 20 kV. This analysis was performed in order to evaluate the possible distortion of the measurements related to the crystallographic orientations due to an incorrect detection of the martensite.

3. Results

3.1. Metallographic Characterization

The micrographs show a structure with the presence of two phases: thin ferritic grains surrounded by martensite. The presence of grains elongated along the rolling direction has been pointed out and this structural organization of the two main structural constituents could be one of the source of the induced mechanical anisotropy (Fig. 1).

The two steel grades show a different quantity of the structural constituents (Table 2) (Fig. 2).

It is evident that the structure belonging to Steel A shows a larger volume fraction of martensite than Steel B and this structure appears also to assume the thinner morphology of the structural constituents. The dimensions of the ferritic grains is not uniform and it is not possible to obtain a significant average value of this quantity, because the standard deviation assumes very high value if compared with the average one.

3.2. Mechanical Properties

The mechanical properties of the two steels were evaluated for different cross head speeds, but the difference among the obtained values are negligible and then only the results for the cross head speed of 25 mm/min are represented Tables 3, 4.

For both steels the Young Modulus assumes the maximum value along the direction perpendicular to the rolling one. The difference among the data related to this direction and the ones at 0° and 45° from the rolling direction can be quantified in average value of 10 GPa. Steel A shows a greater value of stiffness than Steel B.

The stiffness variations along the three directions are the consequence of the different orientation of the grains in the materials. Steel A shows a greater stiffness than Steel B because of the larger quantity of martensite.

The yield point of Steel A is discontinuous and is evaluated through the autographic diagram method, while the yield of Steel B is continuous and then the yield stress were determined by the offset method.

Steel A is stronger than Steel B because of the greater quantity of the contained alloying elements, the related amounts of martensite and the finest microstructure. The best mechanical features are always along the transverse direction.

The strain hardening exponent (\(n\)) was obtained in two different ranges of deformation: \(n_1\) between 1\% and 3\%, \(n_2\) between 3\% and 5\%.\(^{16}\)

|        | Steel A | Steel B |
|--------|---------|---------|
| ferrite (%) | 19.9 (1.5) | 36.9 (1.7) |
| martensite (%) | 80.1 (1.5) | 63.1 (1.7) |

Fig. 1. Micrographs of the studied steel: (a) steel A, (b) steel B along the main directions.

Fig. 2. SEM Micrographs of Steel A (a) and Steel B (b).
The greater values of the strain hardening exponent are those of Steel B in all cases. This implies that this steel, mechanically weaker, shows a greater increase of the yield stress than Steel A after the strain hardening related to the plastic deformation mechanism. The higher hardening coefficient the more homogeneous is the distribution of the strain along the different directions during the plastic deformation and this provides a better formability.

Also the strain rate sensitivity was calculated and the obtained values are coherent with that present in literature. The tensile strength shows the better behavior of Steel A than Steel B. The transverse direction shows the higher values. The values of tensile strength confirm the data obtained from the yield stress. The increase of test speed causes a light increase of tensile strength, consequently a better performance.

Uniform elongation is an interesting feature for the deep drawing steel, actually this well represents the formability limit of the materials. These values include both the elastic and plastic components of the elongation and they can be determined directly from the stress–strain tests. The values of the uniform elongation show a great spread around the mean value. This can be explained considering that the specimens were taken from different parts of the sheet and then the samples obtained from the centre of the sheet have better characteristic than the ones taken from the boundary zones.

The observations of the fracture surfaces show a general
ductile behavior featured by a great quantity of dimples and
the presence of some zones with micro-voids which seems
to be more frequent and pronounced on the fracture surface
of Steel B (Fig. 3).

The microstructural observations are coherent with the
macroscopical mechanical behavior which shows a ductile
fracture that can be deduced from the following features:
great slips due to the plastic deformation and considerable
absorption of energy (high plastic deformation in the stress
strain curve). The frequency of the micro-voids is probably
promoted by the role of the martensite which confines the
deformation of the ferrite in more localized regions induc-
ing a significant coalescence of the ductile dimples formed
in the ferritic grains.

The measurements of Lankford coefficients were carried
out with a definite strain 2% or 5%. The experimental data
show the greatest values of the $r$-value along the direction
at 45° from the rolling one. The relation between the mean
value of $r$ and the deformation speed is not well estab-
lished.17)

The average $r$-coefficients were calculated, applying the
Eq. (2). Steel A shows greater values than steel B, so the
first one shows a better deep drawability.

Ears along the 45 degrees from the rolling direction after
the deep drawing tests have turned out from the tests per-
formed previously (Fig. 4).

The planar anisotropy was calculated applying the Eq.
(3), for the two steels, for the different amounts and speeds
of deformation. The planar anisotropy coefficient is a mea-
sure of the magnitude of the $r_{45°}$ value compared with the
other direction coefficients.18) The values show the tenden-
cy of Steel B to point out the earing phenomenon which is
less evident for Steel A.

The forming limit diagrams of the two steels have also
been determined (Fig. 5) and show a better formability of
the Steel B which is due to the larger fraction of the ferrite
present within this steel and to the high value of $n$.

The lowest limit of strain shown on FLD corresponds to
the state of plain strain (FLD_0).19)

3.3. Determination of Crystallographic Texture

From the ODF (analysis) performed by XRD analysis the
main component orientations featuring the cold rolled sheet
steel were obtained. The ODF ($\varphi=45°$, notation by Bunge,
Figs. 6 and 7) (Table 5) for the four specimens show the
greater components present in the different steel grades.

Components \{111\}(\{101\) is very spread and it is possible
to see a weak evidence of their presence. Before the heat
treatment in Steel B the component that shows the maxi-

![Fig. 4. Earing of deep drawing cups, on the left Steel A, on the
right Steel B. The arrow shows the 45° direction with the
respect to the rolling direction.](image)

![Fig. 5. FLD of the Steel A (a) and Steel B (b).](image)

![Fig. 6. ODFs ($\varphi=45°$ cross section) of Steel A before the heat treatment (a) and after the heat treatment (b). The values
are in random units.](image)
mum intensity is \{6\ 13\ 14\}\((5\ 5\ 1)\). For Steel A and Steel B after the heat treatment the component with the maximum intensity is \{115\}\((5\ 5\ 1)\). However, these data should be normalized, the first one to the component \{122\}\(2\ 1\) and the second one to the component \{001\}\(1\ 1\).

After the heat treatment, in Steel A, all the components are unchanged but \{121\}\(1\ 1\) component shows a significant decrease. For Steel B, after the heat treatment, the components \{100\}\(0\ 1\) and \{111\}\(1\ 1\) and \(g\)-fiber appears to be nearly unchanged, but the component \{121\}\(1\ 1\) decreases significantly.

The component \{100\}\(0\ 1\) shows a greater intensity in Steel A than in Steel B. Moreover, the intensity of the component \{111\}\(1\ 1\) and of the \(g\)-fiber are greater in Steel A before the heat treatment, while is the same for both materials after the heat treatment.

From the analysis of the specimens 2% and 5% strained, it is possible to note that, for both materials, the texture components are unchanged but for a weak increase of the intensity of \{111\}\(1\ 1\) as the strain increases.

The texture analysis developed through EBSD on the martensite structures have shown a very similar pattern of the orientations which can be clearly pointed out by the three inverse polar figures (Fig. 8).

In the Steel A the textures of the martensite appear to be slightly sharper. This proves that the parent texture of the steel before heat treatment and martensite formation has to be very similar, the texture of the martensite after heat treatment appears to be a constant factor in the two steels, the structure of the martensite influences the steel behavior only through the volume fraction and the morphological relation occurring among the ferrite grains (and their texture) and the martensitic zone.

### 4. Discussion

From the results of the metallographic analysis, the materials have a ferritic–pearlitic structure before the heat treatment, while it becomes ferritic–martensitic after annealing. Steel A is characterized by an average volume fraction of 80.1% of martensite while Steel B by 63.1% of this value. In Steel A the ferritic grains appear finer than in Steel B. In the material before the intercritical annealing treatment, the presence of elongated grains along the rolling direction is evident, and still after the thermal treatment this morphological characteristic is observable (Fig. 1).

The Young Modulus revealed in the two steels appears extremely low, especially in the ones tested before the heat treatment. However, this behaviour can be explained by the presence of the component \{001\}\(0\ 1\) as suggested also by the studied performed by other authors. The trend of the Young Modulus as a function of the rolling direction is very similar to the one induced by \{001\}\(0\ 1\) texture and its fundamental influence can explain the lowest values observed in the steel before the heat treatment, when the textures featured by \{001\} planes and developed by cold rolling has not been already transformed.

The mechanical tests have shown that, for each material, the best mechanical strength is related to the transverse direction. Along 45° there is a better formability and worse strength properties. The strain hardening exponents \(n\), 0.15 for Steel A and 0.16 for Steel B are relatively large if compared to the ones belonging to the other steels featured by a tensile strength of the same order of magnitude. These values could explain the good formability of the two grades because they impose a uniform and distributed deformation in the sheet before each step of increase of the applied stress. The fractografy technique shows a more ductile frac-
ture for Steel B than for Steel A certainly caused by a lower content of martensite in the first one (Fig. 4).

The $r_m$ values are higher in Steel A ($\approx 0.95$) than in Steel B ($\approx 0.85$). It is important to underline that $r$-coefficients are maximum, for each material, along the direction rotated of 45 degrees from the rolling one. Steel B shows values of the $r$ coefficient higher than Steel A only along 45°, so a better formability of Steel B turns out. Planar anisotropy coefficients ($\Delta r$) are approximately 0 for Steel A while it is $-0.25$ for Steel B. It justifies the isotropic behavior and the tendency to no earing of the Steel A experimentally observed. The negative value of planar anisotropy ($\Delta r$) for Steel B is in agreement with the presence of ears along the 45° after the cup test (Fig. 4).

After the determination of $r$-coefficients, yield surfaces have been computed (Fig. 9). Steel A and Steel B show a different behavior than a mild steel that shows a Von Mises limit diagram which shows a best attitude to face a biaxial state of stress.

FLD$_0$ value represents the most critical strain state on the FDL. High value of FLD$_0$ means better formability. The FLD$_0$ for the Steel B is 0.17 and for the Steel A is 0.12.

The behavior of the investigated steels and its variation can be well explained on the basis of the crystallographic measurements. From the analyzed polar figures and ODF, the steels show the following texture variations from the step before the heat treatment to the as-cold rolled conditions:

- Steel A:
  - constant intensity of $\{100\}$
  - intensity reduction of $\{121\}$
  - intensity reduction of the component of $\{111\}$ and of $\gamma$-fiber;

- Steel B:
  - constant intensity of $\{100\}$
  - intensity reduction of $\{121\}$
  - constant intensity of $\{111\}$ and of $\gamma$-fiber.

Because the thermal treatment is short, it is reasonable that a decrease of components $\{100\}$ does not take place as one would expect in the case of a long time annealing.

Fig. 8. Inverse pole figures for martensite of Steel A (a) and Steel B (b).
The same consideration could be applied to the component featured by \{111\}, because the second phase present in these steels could prevent its development. Reduction of component \{121\} is a typical consequence of the annealing treatment.

From the comparison between the two steels, it is clear that:
- component \{100\}(011) is stronger in Steel A than in Steel B. This trend is evident before and after annealing;
- component \{111\}\{0\}1 and \gamma\text{-fiber} texture show a higher intensity in Steel A before annealing, while it is the same for both materials after it.

In Steel B there is a greater presence of \{100\}(011) than in Steel A, so it causes lower \(r\)-coefficients in Steel B (smaller value) than that in Steel A (higher value). In the latter, the component \{100\}(011) has a smaller intensity, so it confirms a better formability for this steel.

Component \{121\}\{1\}01 is a typical cold rolling texture and it is present in both materials. It does not have a particular effect on \(r_m\) values, but it increases \(r_{2\alpha}\). So determining an influence on \(\Delta r\) to which earing is connected.

The collected experimental data allow to create a correlation with those obtained from the mechanical tests, particularly with the planar anisotropy coefficients.

Slip plane of \{100\} gives the worst drawing quality of deep-drawing sheet, while component \{111\} is the ideal texture for deep-drawing sheet, because the correct texture gives the proper orientation of slip system so that the strength in the thickness direction is greater than that in the plane of the sheet.

In order to improve the formation of \gamma\text{-fiber} (inducing a better formability) and to reduce the presence of component \{100\}(011), it should be useful to modify the heat treatment parameters; an increase of the time of the heat treatment can be a possible solution.

The \(r\)-coefficient is also influenced by the reduction path during rolling. High thickness reductions improve the presence of components \{100\}(011) in the plane of the sheet, so lower thickness reduction could be useful to increase the \(r\)-coefficient.

5. Conclusion

In the present study the microstructural characterization and the relation of texture and deformation properties of two different high martensitic dual phase sheet steels are investigated.

(1) The micro-structural analysis show ferritic-martensitic structure after annealing: Steel A is characterized by an average volume fraction of 80.1% of martensite while Steel B by 63.1% of this value, in Steel A the ferritic grains appears finer than in Steel B.

(2) The values of normal anisotropy coefficients \(r_m\) are not high if compared with the traditional behavior of the typical mild steel but the high values of the stain hardening exponents \(n\) (0.15 for Steel A and 0.16 for Steel B) give to these steels good deformation property. Moreover, the FLDF values show the better formability of Steel B than Steel A.

(3) The values of normal anisotropy coefficients \(r_m\) are related to the components of texture present in the two sheets. In Steel A the significant presence of components with \{111\} parallel to the rolling plane and the lower presence of the ones with \{100\} than in Steel B justify the higher value of \(r_m\) for Steel A.

(4) The values of planar anisotropy \(\Delta r\), related with the earing phenomenon, is approximately 0 for Steel A while it is \(-0.25\) for Steel B. For the last one the ears take place in the direction 45° to the rolling one after the deep drawing tests and they are related with the intensity of the components texture present in the materials.

Acknowledgement

The authors would like to thank SSAB Tunnplat AB, in particular Eng. Björn Carlsson, for the technical support. Helpful advices of Prof. Bevis Hutchinson of the Swedish Institute of Materials Research during the course of investigation are greatly appreciated. The authors also express their appreciation to Prof. Marcello Barico of the Università degli Studi di Torino for his availability for the texture measurements and to Ing. Montagnoli Fabrizio for his care during experimental tests.

Nomenclature

- \(\varepsilon_w\): Deformation of the tensile specimen along width
- \(\varepsilon_t\): Deformation of the tensile specimens along thickness
- \(w_0\): Initial width of the tensile specimen (mm)
- \(w\): Final width of the tensile specimen (mm)
- \(t_0\): Initial thickness of the tensile specimen (mm)
- \(t\): Final thickness of the tensile specimen (mm)
- \(r\): Lankford coefficient
- \(r_m\): Average of the Lankford coefficients among the main directions
- \(r_{2\alpha}\): Lankford coefficient in a direction rotated by \(\alpha\) angle from the rolling one
\Delta r : \text{Coefficient of planar anisotropy} \\
\varphi_1, \Phi, \varphi_2 : \text{Angle in the ODF analysis with Bunge notation} \\
n : \text{Strain hardening exponent}

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