

# Mechanical Stability of Retained Austenite in the Nanostructured, Carbide Free Bainitic Steels during Tensile Testing and Cold Rolling Process

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## Abstract

Terms of service of the steels with retained austenite in the microstructure, is severely depended on the stability of austenite during the course of deformation. The present work aims to evaluate the mechanical stability of austenite in nanostructured, carbide free bainitic steels during tensile testing and cold rolling process. To achieve the microstructure with retained austenite, the steel with chemical composition of (in wt.%) 0.76 C-1.95 Si-1.4 Mn- 0.12 Ni- 0.24 Mo-0.69Al isothermally transformed at temperatures of 200, 250 and 300 °C for 6, 16 and 72 hours, respectively, after austenitizing at 950 °C for 30 min. Austenite stability was evaluated by means of microhardness testing, x-ray diffraction pattern analysis and color metallography. Further study was carried out by comparing the incremental work hardening exponent obtained from the stress-strain curve. According to the results, austenite transforms into martensite in a progressive manner; blocky austenite at the early, but filmy austenite at the later stage of deformation. Samples transformed at the 300 °C have a good capability of retained austenite to martensite transformation. This is as a result of the higher mechanical stability of the austenite in this microstructure.

*Keywords:* Mechanical stability; Retained Austenite; Tensile testing; Cold rolling; Nano bainite.

## 1. Introduction

Nanostructured, carbide free bainitic steels exhibit a perfect combination of high strength and ductility due to their multi phase microstructure including bainitic ferrite and retained austenite. The mechanical properties evaluation revealed ultimate tensile strength of about 2.2 GPa, hardness values in the range of 600–670 HV and a noticeable uniform elongation in the range of 5–30% <sup>1-10</sup>. These steels have been gained interest in automobile components such as chassis, engine compartments, bumpers and side doors, as well as a range of applications in construction, offshore and defense industries <sup>11</sup>.

Much of the strength and hardness of nanostructured, carbide free bainitic steels come from the very small thickness of bainitic ferrite plates <sup>1-10</sup>. Also, unique mechanical properties of the steels can be resulted from the

combined deformation behavior of retained austenite and bainitic ferrite phases. It is reasonable to assume that retained austenite as a softer phase will control the levels of ductility of the steel. Further enhancement of the ductility can be achieved by the transformation of retained austenite into the martensite during the course of deformation <sup>12</sup>, the so called transformation induced plasticity (TRIP) effect.

The decomposition of the retained austenite into the martensite which can be resulted further ductility in nanostructured bainitic steels is a subject which is not visible in other nanostructured materials. In nanostructured materials, after the yield point, due to the limited activity of dislocations, local instability in plastic deformation occurs. Therefore, nanostructured materials show a lower ductility. This problem has been solved at the nanostructured bainitic steels by the austenite to martensite transformation.

The stability and transformation rate of the austenite is decisive to obtain a good ductility in these steels. Austenite microstructure characteristics, including morphology and its carbon content, are the determinants of residual austenite stability. Other factors that can affect the TRIP of retained austenite, is the dislocation density and strength of the adjacent phases i.e. bainitic ferrite. However, due to the larger and unique effects of the retained austenite on the mechanical properties of nano-

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structured bainitic steels, the present work aims to evaluate the variation of the volume fraction of retained austenite induced by the applied tensile and pressure stresses during two different mechanical processes of tensile testing and cold rolling process. Also other microstructure terms of the steels like work hardening exponent during tensile testing, hardness variations and other subjects will be studied.

## 2. Experimental Procedure

The chemical composition of the steel is listed in Table 1. The calculated TTT diagram by the use of MUCG83 thermodynamic model<sup>13)</sup>, is shown in Fig. 1. MUCG83 thermodynamic model developed by Bhadeshia et al, predicts the TTT diagram from the chemical composition of the steel. According to the figure, bainite and martensite start temperatures, are 435 and 157 °C, respectively. TTT diagram parameters are also listed in Table 2. As it is shown in this Table, the diffusional and bainite nose times are different from

Table 1. Chemical compositions of the steel (wt. %).

C	Si	Mn	Cr	Ni	Mo	Co	Al	Fe
0.76	1.95	1.4	1.42	0.12	0.24	0.14	0.69	Bal.

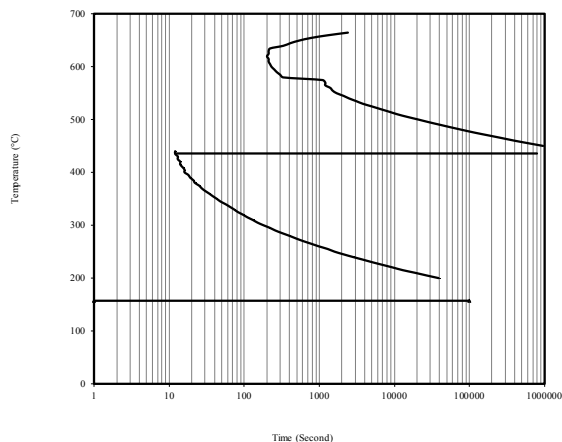


Fig. 1. Calculated TTT diagrams by the use of MUCG83 thermodynamic model.

Table 2. TTT diagram parameters.

TTT Diagram parameters	Steel
Diffusional Nose Time (s)	200
Diffusional Nose Temperature (°C)	620
Bainitic Nose Time (s)	12
Bainitic Nose Temperature (°C)	435
Martensite Start Temperature (°C)	157

each other; therefore bainite reaction can occur without any restrictions resulting from the diffusional transformation.

The tensile tests specimens were prepared according to the ASTM E08 standard test method with a gage length of 25 mm<sup>14)</sup>. The samples of cold rolling process were prepared with a dimension of 8 mm in width and 10 cm in the length with a little slope in the front part of the samples to make the start of the rolling process easier. Detailed information of the chemical composition of the steel and processing is similar to the work of Yoozbashi et al<sup>9)</sup>.

The tensile and compression samples after austenitizing at 950 °C for 30 min, were transformed isothermally at temperatures of 200, 250 and 300 °C for 6, 16 and 72 hours, respectively. Then samples were cooled in the air. Tensile tests were carried out at the room temperature with a cross-head speed of 0.1 mm/min. The specimens used for compression test, were reduced 0.4 mm in the thickness direction at each stage of the cold rolling process.

At each stage of deformations, the microhardness test, x-ray diffraction (XRD) pattern analysis and color metallography were carried out on the specimens. The microhardness measurements were carried out under a 100 g<sub>r</sub> load, from the surface to the center of the specimens in the thickness direction.

To determine the volume fraction of retained austenite (VRA), quantitative XRD analysis was carried out on the samples according to the ASTM E975-84 standard<sup>15)</sup>. Also, the carbon content of the retained austenite was calculated from the angular position of the peaks from the (200), (220) and (311) planes of austenite. Detailed information on the measuring of carbon content of retained austenite has been given in the Yoozbashi et al<sup>9)</sup>.

To reveal the martensite formation from the retained austenite decomposition during the deformation process, pre-etch with Nital 2% and subsequent etch with sodium metabisulfite 20 wt. % was used to tint and color the samples. Then the samples were examined using polarized light to obtain better results. Sodium metabisulfite has been used in a number of concentrations, as a strong coloring etchant for irons and steels. It tints martensite into brown, bainite into blue, carbides, phosphides, and residual austenite, unaffected<sup>16-18)</sup>.

## 3. Results and Discussion

### 3.1. Tensile testing

Fig. 2 shows the tensile curve of the samples transformed at three different transformation temperatures. It can be seen that a higher strength has been achieved for samples transformed at 200 °C. Samples transformed at 300 °C, have shown more ductility in comparison with the other samples. The lower yield strength of the samples transformed at 300 °C is a result of the higher thickness of bainitic ferrite and its lower volume fraction according to the lever rule.

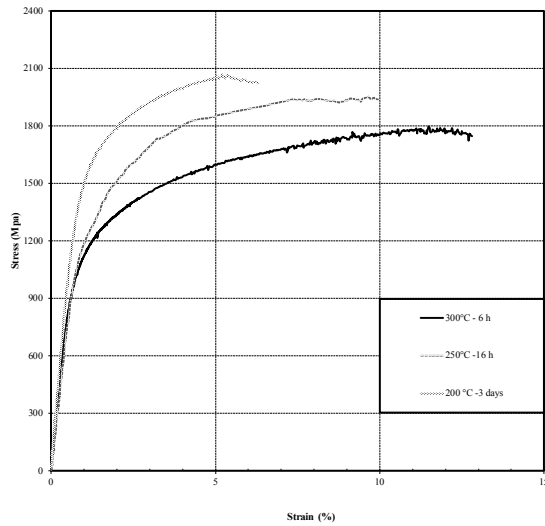


Fig. 2. Tensile curves of samples transformed at three different transformation temperatures.

A method to analyze the mechanical stability of retained austenite during tensile testing is to determine the incremental strain hardening exponent ( $n_{incr}$ )<sup>19</sup>. This parameter is derived from the common power-law relationship that relates true stress ( $\sigma$ ) and true strain ( $\epsilon$ ),  $\sigma = k\epsilon^n$  and is defined as:

$$n_{incr} = \frac{d \ln \sigma}{d \ln \epsilon} \quad \text{Eq. (1)}$$

Work hardening represented in the form of  $n_{incr}$  as a function of true plastic strain.  $n_{incr}$  vs true plastic strain of the stress-strain curve given in the Fig. 3a, has been plotted in the Fig. 3b. The straight line represents the points where,  $n_{incr} = \epsilon_p$ , which is called the instability criterion. Instability criterion, where uniform true strain is equivalent to  $n_{incr}$ <sup>19</sup>. The difference in strain hardening can be attributed to the difference in which retained austenite transforms to martensite during plastic deformation.

From the Fig. 3b, it is evident that strain hardening rate of the specimens is higher in the early stage of tensile deformation. It is anticipated that higher rate is due to the lower mechanical stability of blocky shape austenite<sup>20, 21</sup>. Retained austenite in two morphologies involving thin films between bainitic ferrite's subunits and blocks separating the sheaves exists in the microstructure. Because of lower carbon content, the blocky shape austen-

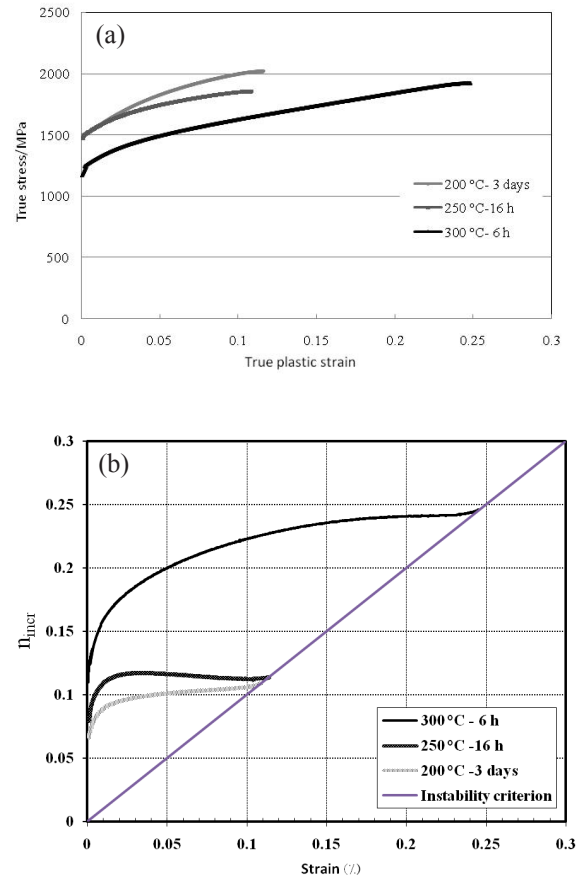


Fig. 3. The tensile curves (a) and incremental strain hardening exponent ( $n_{incr}$ ) as a function of true plastic strain (b). The straight line represents the instability criterion,  $n = \epsilon_p$ .

ite transforms into martensite at the early stage of deformation.

As shown in Fig. 3b, the samples transformed at 300 °C, experienced higher strain hardening rate compared to the samples transformed at 250 and 200 °C. Higher mechanical stability of retained austenite in specimens transformed at this temperature is due to its higher carbon content and higher volume fraction of the retained austenite in the microstructure. The volume fraction of retained austenite and its carbon content, measured from the XRD pattern analysis, have been given in Table 3.

For the specimens transformed at 200 and 250 °C, after an initial immediate increase,  $n_{incr}$  has a relatively constant slope to the necking point, while the samples

Table 3. The volume fraction of retained austenite and its carbon content for three different transformation temperatures.

Transformation temperature (°C)	Volume fraction of retained austenite (%)	Carbon content of retained austenite (%)
300	34	1.53
250	21	1.25
200	19	1.1

transformed at 300 °C, have a continuous moderate increase after an initial higher rate. In the case of the 200 and 250 °C specimens, there is a rapid transformation of austenite to martensite in comparison to the samples transformed at 300 °C that makes it impossible to get full advantages of the ductility increase which is the consequence of the complete and gentle transformation of austenite to martensite. For the samples transformed at 300 °C, this transformation gradually and increasingly continues to the instability criterion and as a result a significant elongation of 27% has been achieved in the tensile test of these samples.

**3. 2. Cold rolling process**

Fig. 4a-b shows the hardness of the surface and center of the samples via engineering strain percentage in each stage of cold rolling deformation. From the Fig. 4, it is evident that the hardness of the specimens increases with decreasing the transformation temperature. This is as a result of higher bainitic ferrite volume fraction and fine thicknesses of bainitic ferrite at lower transformation temperatures.

The enhancement of hardness in each step of plastic deformation can be resulted from the transformation of retained austenite to martensite and also from the strain hardening effects. This behavior implies that austenite transformation into the martensite and also, strain hardening occur progressively during deformation. It can be

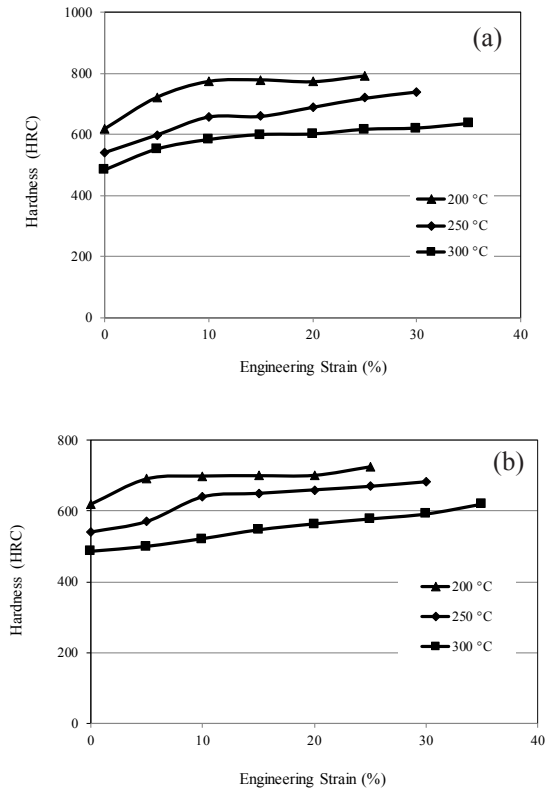


Fig. 4. Variation of the hardness of the (a) surface and (b) center of the samples as a function of engineering strain percentage in each stage of cold rolling deformation.

seen that the hardness increasing rate is higher at the early stage of the rolling process. Similar to a tensile test, this is related to the fact that blocky shape unstable austenite transforms into the martensite at the early stage of deformation. This process is not dominant in Fig. 4b, since the pressure increases gradually in the center of the samples.

To check the stability of the retained austenite, the variations of hardness increasing ratio ( $\frac{h_t - h_0}{h_0}$ ) via engineering strain has been plotted in Fig. 5a-b for surface and center of samples, respectively, which  $h_0$  and  $h_t$  are initial hardness and hardness at each stage of deformation, respectively.

From Fig. 5, it can be concluded that the transformation of the austenite into the martensite has been occurred faster in the samples transformed at 250 °C. This behavior has also occurred on the surface and the center of samples. This is in agreement with the results obtained from the incremental strain hardening exponent. In the early stages of cold rolling deformation, samples transformed at 200 °C have shown a higher percentage of transformation in the surface and core of the samples. In the later stages, transformation rate has been de-

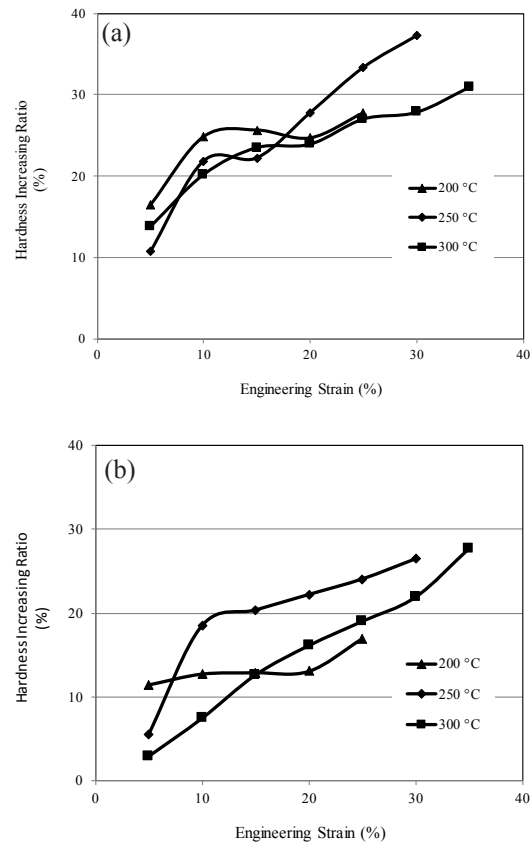


Fig. 5. Variations of the hardness increasing ratio ( $\frac{h_t - h_0}{h_0}$ ) of the (a) surface and (b) center of the samples as a function of engineering strain percentage during cold rolling deformation.

creased. This behavior arises essentially from the fact that the retained austenite surrounded by strong bainitic ferrite phase at the early stages, therefore retained austenite decomposition has been retarded. This type of austenite has been located between the bainitic ferrite subunits, remained intact until the late stage of deformation. This Austenite has been named as a filmy austenite.

As mentioned previously, filmy austenite due to the higher carbon content, remains intact until the final stages of deformation. Also, according to Fig. 5, the specimen transformed at 300 °C has shown a uniform and progressive hardness enhancement in both of the surface and the center of samples during the process of plastic deformation which leads to optimal use of the austenite into martensite transformation. So this leads to a greater increase in ductility.

Moreover, the transformation of austenite into the martensite during cold rolling deformations has been investigated by color metallography. Color metallography of samples transformed at 300 °C and then deformed by cold rolling at two different true plastic

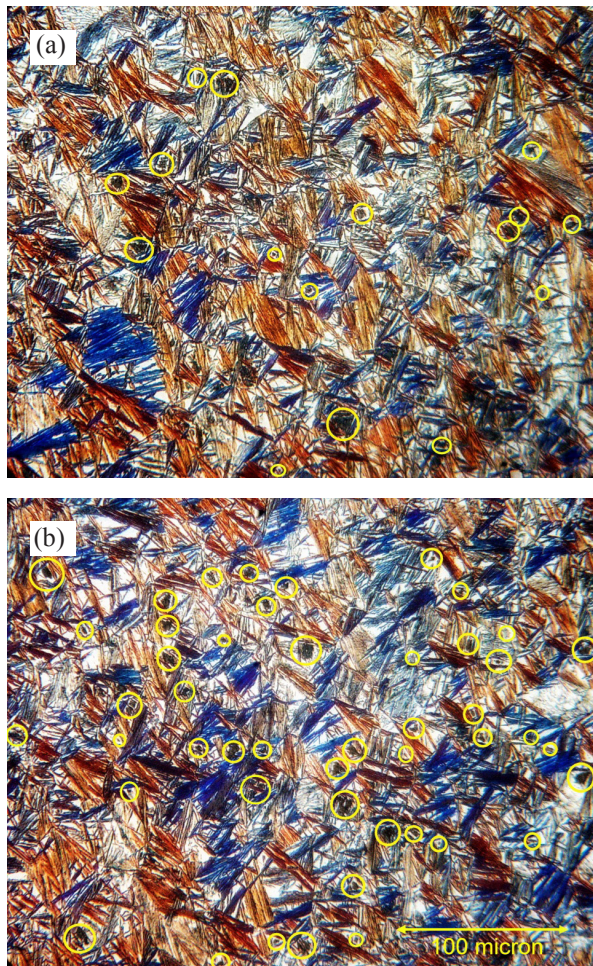


Fig. 6. Color metallography of the samples transformed at 300 °C and then deformed by cold rolling at two true pressure strain levels of 2% (a) and 7% (b). The yellow circles show the area of martensite.

strains has been shown in Fig. 6a-b. It can be seen that, with changing the strain from 2 to 7 %, the volume fraction of martensite phase has been increased. The areas marked with yellow circles represent the areas in which the blocks of austenite have been transformed into martensite during cold rolling deformation. Usually, martensite appears in the form of zigzag in high carbon

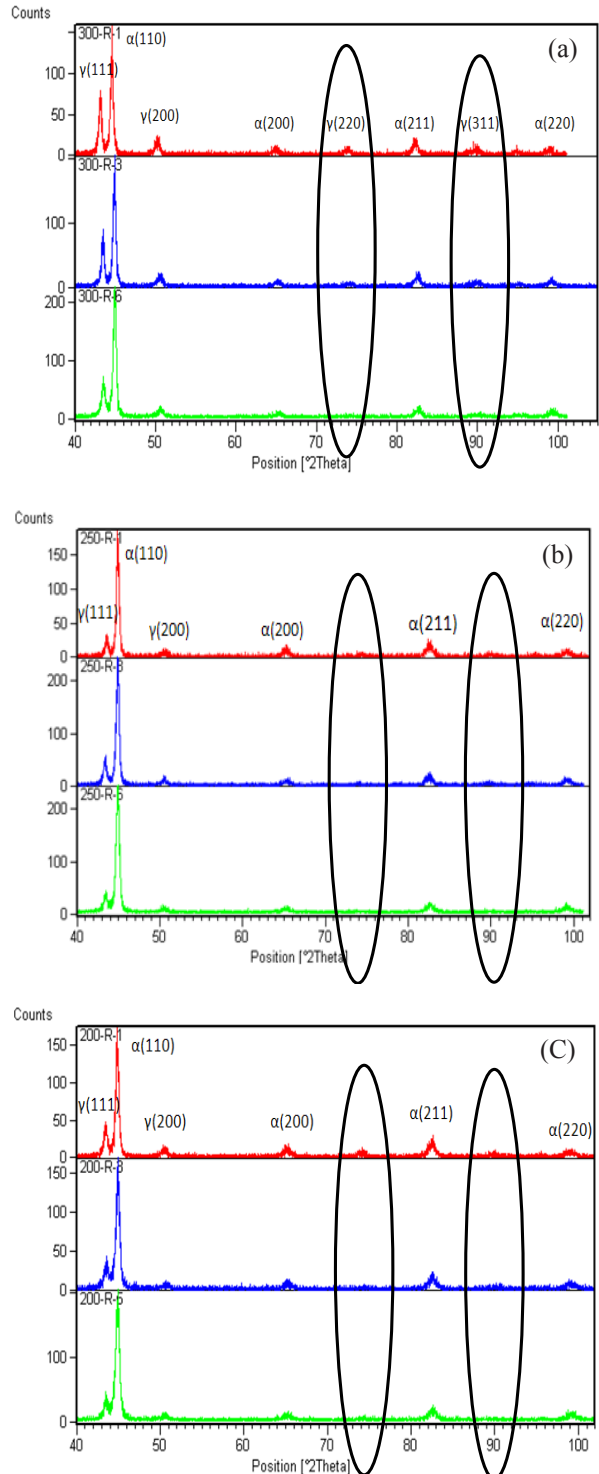


Fig. 7. The XRD patterns of the specimens transformed at a) 300, b) 250 and c) 200 °C for three true pressure strain levels of 1, 3 and 5%.

steels (about 1% C).

Variation of volume fraction of retained austenite (VRA) during deformation has been studied by XRD. The XRD patterns of the specimens transformed at 200, 250 and 300 °C have been shown in Fig. 7a-c respectively for three true pressure strain levels of 1, 3 and 5 %. It is evident that, the intensity of the austenite peaks has been decreased gradually for three different of specimens. From the gradual removal of the austenite picks by plastic strain, it can be understood that, austenite consumption for the samples transformed at 300 °C is slow, so that the peaks of austenite is detectable after 15% engineering strain. As mentioned previously, samples transformed at 300 °C have shown higher mechanical stability, according to the XRD results.

#### 4. Conclusion

The same mechanical stability was observed during tensile testing and cold rolling process. The higher rate of austenite consumption at the early stages of deformation is due to the lower mechanical stability of blocky shape austenite. But, filmy shape austenite, due to its higher carbon content, remains intact until the late stages of deformation.

Higher mechanical stability of retained austenite of the specimens transformed at 300 °C is due to the higher carbon content of retained austenite and its higher volume fraction in the microstructure. In the case of specimens transformed at 200 and 250 °C, there is a rapid transformation of austenite into martensite that makes it impossible to get full advantages of the ductility increase which is the consequence of the complete and gentle transformation of the austenite into the martensite.

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