Hot Deformation Characteristics of Si–Mn TRIP Steels with and without Nb Microalloy Additions

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It has been long recognized that steels with high formability and strength can be produced with the aid of the deformation induced transformation of retained austenite to martensite. Selecting the steel composition and microstructure to obtain significant amounts of retained austenite has been the first step in designing TRIP steels. Thus, it is now well established that a significant amount of retained austenite can be obtained in steels containing high concentrations ($> 1\%$) of Si, where bainite is one of the main microconstituents.

The effects of different intercritical annealing conditions on various aspects of Si–Mn TRIP steels have been extensively studied in other investigations. The present work was, however, conducted to investigate the influence of thermomechanical processing on the general behaviour of two Si–Mn TRIP steels, including a microalloyed Nb-bearing grade. The effects of microstructural features produced by dynamic recrystallization and pancaking of austenite, along with that of finishing temperature below the T_{nr} (the austenite no-recrystallization temperature) on the subsequent transformations of the parent austenite, and the state of retained austenite at room temperature, were examined. It was found that, for microstructures comprised of polygonal ferrite, bainite and retained austenite, the variation of the retained austenite volume fraction with strain in the no-recrystallization region exhibits a maximum. Furthermore, the dynamically recrystallized austenite substructure, when retained to transformation temperatures, tends to retain more austenite.

KEY WORDS: hot deformation; TRIP steels; retained austenite; thermomechanical processing (TMP); dynamic recrystallization and pancaking.

1. Introduction

In general, optimization of mechanical properties in metals and alloys can be achieved by alloying, in conjunction with heat treatment or thermomechanical processing. In steels, various types of compositions and heat treatments have been utilized to improve strength. However, alloying and heat treatments to improve strength are usually accompanied by lower ductility, *i.e.* there is frequently a compromise between strength and ductility. Therefore alternative methods which can result in the simultaneous improvement of both properties have been investigated. The corner stone of one of these approaches was laid by Zackay *et al.*¹⁾ They proposed that the strain or stress induced transformation of any retained austenite present in the microstructure to martensite can significantly strengthen the deforming alloy, and therefore result in higher ductility. This is the basic principle of Transformation Induced Plasticity (TRIP) effects.

Employing the TRIP phenomenon to produce steels with higher strength without loss of formability has been the subject of many recent investigations.^{2–5)} The majority of the studies, concentrating on Si(1–2%)–

Mn(1–2%) steels, have capitalised on the benefits of the presence of bainite in the microstructure. In this way, a new class of high formable steels has been developed, which can be significantly strengthened by deformation induced transformation of retained austenite to martensite. Since these grades of Si–Mn steels are designed to contain retained austenite in order to achieve TRIP properties, they have been classified as TRIP steels.⁶⁾

Si–Mn bainitic TRIP steels in the form of sheet can be used mostly in the automotive industries, such as automatic structural press forming parts, bumpers, reinforcement parts, doorguard bars, *etc.*⁷⁾ The production of Si–Mn TRIP steels in the form of bar and rod to make, for example, high strength fasteners, (bolts, nuts, studs, steering links, tie rods, *etc.*), is also promising.

The steels for these applications must possess high strengths and good formability. For some forming operations, *e.g.* press forming, formability in TRIP steels can be evaluated by the product of tensile strength and total elongation, (TS $\times$ T.EL). The higher the value of this parameter, the better the formability. If the volume fraction of microstructural components (*e.g.*, bainite, ferrite and retained austenite) is kept the same, the highest

value of $TS \times T.EL$ is obtained when the retained austenite stability is optimised.

The present study concerns the control of the microstructure by thermomechanical processing (TMP) in order to attain the optimum conditions (size, distribution, quantity, composition and morphology) of retained austenite in Si-Mn TRIP steels. Although TMP is a powerful method to control microstructure, its application to TRIP steel processing has yet to be investigated systematically. This paper details the effects of microstructural features produced by dynamic recrystallization and pancaking of austenite on the subsequent transformations of the parent austenite, and hence the state of retained austenite at room temperature.

2. Experimental Procedure

Table 1 shows the composition of the steels which were used in this study. These steels were received in the as hot rolled condition, from which specimens were machined. The only difference between these compositions is Nb in steel A. The silicon in these steels exerts a retarding effect on the carbide precipitation from austenite. The resulting carbon enrichment of the austenite adjacent to ferrite plates in this way leads to more retained austenite in these grades of TRIP steels. Nb was originally added to steel A (Nb-steel) to raise the austenite no-recrystallization temperature and enable the steel to be processed by a variety of TMP routes, leading to both recrystallized and pancaked austenite prior to the austenite transformation.

All the mechanical experiments were carried out on a computerized materials testing system, adapted for hot compression tests, which is described in Ref. 8). The retained austenite measurements were performed by X-ray diffraction, as detailed in Ref. 9). The techniques used led to an accuracy of $\pm 2\%$ in the amount of retained austenite. The X-ray data were also used to determine the carbon concentration in the retained austenite. The austenite lattice parameter extrapolation method, together with the following empirical expression,

was employed to calculate the carbon content.⁹⁾

$$a_0 = 3.578 + 0.044 \% C$$

The mechanical properties at room temperature were determined using a small specimen test technique,¹⁰⁾ which was based on punching a 3 mm diameter disk out of a thin (300–350 μm thick specimen) while measuring the load/displacement behaviour on an MTS machine. Test calibration was accomplished by comparing the results with those of conventional tension tests on the same steels.⁶⁾

For the purpose of designing TMP schedules, knowledge of the T_m , Ar_3 and Ar_1 , which are the no-recrystallization temperature, and the start and finish temperatures of the austenite-to-ferrite transformation, respectively, is required. These temperatures were measured by continuous cooling compression (CCC) testing.¹¹⁾ The corresponding critical temperatures of steels A and B (Nb- and Reference steels) are given in **Table 2**. As can be seen, the transformation characteristics of the Nb-steel were significantly affected by the presence of Nb.

3. Parameters Affecting th Austenite Stability

One of the measures of stability of austenite, which has been commonly accepted, is the martensite-start (M_s) temperature. A low M_s temperature is indicative of high austenite stability, *i.e.*, high resistance to martensitic transformation. Generally, austenite stability has been shown to depend on austenite composition,¹²⁾ the size of austenite at the time of transformation (particle size effect),¹³⁾ austenitizing parameters,¹⁴⁾ cooling rates,¹⁵⁾ specimen thickness,¹²⁾ applied stress,¹⁶⁾ pressure,¹⁷⁾ austenite-to-martensite-to-austenite reverse cycling,¹⁸⁾ and thermal arrest (thermal stabilization) whilst cooling.¹⁹⁾

Austenite stability, in fact, can be considered to influence the kinetics of the martensitic transformation, in that it can effectively retard the nucleation and propagation events. Thus, the fundamentals of the mechanisms of austenite stabilization are based on how the aforementioned parameters may influence the nucleation and growth of martensite. Since the martensitic transformation progresses by shear, the rate of martensite formation is lower when the austenite resistance to shear is higher. Microscopic features, such as grain size and austenite fragmentation, as well as submicroscopic characteristics, such as the nature and distribution of defects, have been invoked to qualitatively explain austenite stability. The mechanisms appropriate to the present findings will be covered more deeply at the appropriate points in the discussion.

4. Results and Discussion

4.1. The Effect of Dynamic Recrystallization

As is well known, dynamic recrystallization (*i.e.*, recrystallization under concurrent straining) is observed in FCC metals and alloys (except aluminum) containing moderate to low stacking-fault energy, and is associated with relatively low rates of dynamic recovery and higher

Table 1. Chemical compositions of experimental steels.

Composition	C%	Si%	Mn%	Nb%	Al%	N(ppm)
Steel A (Nb-steel)	0.22	1.55	1.55	0.035	0.028	20–40
Steel B (Reference-steel)	0.19	0.150	1.54	—	0.024	20–40

Table 2. Critical temperatures of experimental steels.

Grade	Nb-steel	Reference-steel
No-recrystallization temp. (T_m , °C)	930 $\pm$ 5	910 $\pm$ 5
Austenite-to-ferrite transformation temp. (Ar_e , °C)	755 $\pm$ 5	750 $\pm$ 5
Start of acicular ferrite formation temp. (°C)	625 $\pm$ 5	650 $\pm$ 5
Austenite-to-pearlite transformation temp. (Ar_1 , °C)	590 $\pm$ 5	650 $\pm$ 5

than average rates of work hardening. Dynamic recrystallization commences only when a critical strain (also called the incubation strain)²⁰⁾ ϵ_c is reached during hot deformation. This is somewhat less than the strain at the maximum stress, *i.e.* the maximum (or peak) strain ϵ_p .²¹⁾ It is important to note that dynamic recrystallization produces both a texture and grain size distribution which are different from that of static recrystallization.²²⁾ Another noteworthy point of the dynamically recrystallized structure concerns the “dislocation substructure”. Here, a rather poorly recovered dislocation substructure develops in the initial stage of deformation; *i.e.* the cells have a relatively small size and more tangled walls compared to structures which have undergone a high degree of static recrystallization. This substructure, when retained to room temperature, leads to yield

Fig. 1. Schematic diagram of heat and deformation schedule to generate different recrystallization conditions.

Fig. 2. Typical double-hit compression true stress-true strain curves of Reference-steel.

strengths higher than those of annealed materials.²¹⁾

4.1.1. TMP Schedule

To follow the effect of the dynamically recrystallized microstructure on the state of the retained austenite, the specimens were first solutionized (Fig. 1) to dissolve the Nb(CN) and AlN precipitates present in the initial microstructure and then processed by double hit compression at two different temperatures. In order to dynamically recrystallize the austenite, the second pass was carried out with a strain rate one to two orders of magnitude (depending on the temperature) lower than that of the first pass (Fig. 2). The specimen was then cooled down to a temperature below the A_{r3} and held isothermally to form a known volume fraction of polygonal ferrite. This was followed by a salt bath quench, where the specimen was held isothermally for a predetermined duration at 400°C to transform the austenite to bainite (this temperature was shown to be the optimum temperature for maximizing the retained austenite volume fraction.^{4,6)} The specimen was then air cooled to room temperature.

For the Nb-steel, the typical microstructure after the two compressive strains and a 3 min hold at 650°C is shown in Fig. 3. In this micrograph, the white areas are

Fig. 3. Typical microstructure of Nb-steel after double-hit compression, and 3 min holding at 650°C, followed by quenching into the salt bath (400°C).

Fig. 4.

Variation of the V_{RA} of Nb-steel in the specimens undergone different recrystallization procedure.

polygonal ferrite, which have mostly nucleated at the prior austenite grain boundaries. The black regions are a combination of bainite, retained austenite and martensite. The recrystallized austenite grain sizes for steels A and B were found to be 40 and 45 μm , respectively.

4.1.2. Microstructural and Mechanical Property Results

The results of microstructural examination to determine the distribution of retained austenite, using a modified color etching method, revealed that a wide range of retained austenite distribution and morphologies can be produced by the aforementioned TMP schedule. It was demonstrated^{6,23)} that retained austenite can be enclosed by (i) polygonal ferrite, (ii) bainite packets and (iii) upper and lower bainite platelets. In this case, they were mostly encompassed in upper bainite (*i.e.*, pools (ii) and (iii)).

Figure 4 illustrates the variation of retained austenite volume fraction (V_{RA}) for different conditions of static and dynamic recrystallization of Nb-steel. The decrease of the temperature from 1050 to 1000°C in both cases results in a decrease in the V_{RA} . The level of V_{RA} reduction in the case of dynamic recrystallization ($\sim 12\%$), however, is less than that of the static recrystallization case ($\sim 22\%$). Note that, the differences in prior austenite grain sizes (resulting from temperature and deformation variations in these cases), are in the range of $\pm 5\mu\text{m}$. Their effects on the subsequent transformation characteristics and on the retained austenite volume fraction are therefore negligible.

Fig. 5. Variation of the V_{RA} of Nb-steel with different conditions of dynamic recrystallization.
Condition 1: Slow Cooling (3°C/s)
Condition 2: Rapid Cooling (10°C/s)

Fig. 6. Mechanical properties of steel A for different conditions of dynamic recrystallization. (1 and 2 refer to Conditions 1 and 2, respectively)

In order to observe the effect of dynamic recrystallization substructure, the dynamically recrystallized austenite was cooled down rapidly to the next step of processing. The result of this treatment is presented in Fig. 5. In the case of Condition 1 the specimen was cooled at 3°C/s to a temperature below A_{r3} , and this was increased to 10°C/s in the Condition 2. As can be seen, the V_{RA} was increased to a higher level. A comparison between the results of these two conditions indicates that the dynamically recrystallized austenite substructure, when retained to transformation temperatures, is more prone to retaining austenite. Microstructural examinations did not show any differences in the retained austenite size and distribution.

The mechanical properties of the specimens dynamically recrystallized at 1000°C, followed by rapid and slow cooling, are shown in Fig. 6. As can be seen, the variation of total elongation follows the trend of the retained austenite volume fraction change. The UTS of the rapidly cooled material, even with higher amounts of retained austenite, is still higher than that of the slowly cooled structure.

4.1.3. Discussion Concerning the Effects of Dynamic Recrystallization

Since the prior austenite grain size was the same for both dynamically and statically recrystallized cases at 1000°C, the two major differences between these austenite structures, are: (i) the level of solute Nb in Nb-steel, and (ii) the substructures.

During dynamic recrystallization, austenite had undergone continuous straining with longer holding time period compared to the static case. As is well established, this will likely take the Nb out of solution through carbonitride precipitation,²³⁾ *i.e.* the Nb level was lower in the dynamically recrystallized specimen. As is well known, solute Nb increases hardenability and decreases the M_s temperature by solid solution strengthening of parent austenite. Therefore, for a given TMP schedule, austenite with more Nb in solution has a higher propensity to be retained or stabilized.²³⁾ This can be firstly related to the inhibiting effect on the pearlite transformation (higher hardenability) and, secondly, through the presence of dissolved microalloying elements leading to solid solution strengthening in the austenite, therefore reducing the M_s temperature. Accordingly, taking Nb out of solution by changing the processing parameters (*e.g.*, decreasing the deformation temperature or increasing the hold time) decreases the retained

austenite volume fraction at the end of processing. Therefore, having the austenite dynamically recrystallized at lower temperature should result in less retained austenite. On the other hand, the present results revealed that the faster cooled specimen retained a higher volume fraction of austenite. To rationalize these results, the different features of dynamic recrystallization substructure and their effects on the parent austenite stabilization (changes in the M_s temperature) will be discussed below.

It has been shown that the dislocation density and distribution in a dynamically recrystallized microstructure can be retained to room temperature.^{21,24)} Thus, if the specimen is cooled quickly enough after dynamic recrystallization, the higher temperature substructures and defects will be included in the subsequent structure. This would be encouraged in the presence of Nb, *via* dislocation pinning by precipitates and solute drag effects. The effect of substructures on the austenite stabilization (reducing M_s temperature) can be discussed on the basis of the free energy change required for the martensite transformation to occur.⁶⁾ The total free energy change can be considered to be made up of chemical and non-chemical contributions, *i.e.*:

$$\Delta G_{\text{Chem}}^{\gamma \rightarrow \alpha'} + \Delta G_{\text{Non-Chem}}^{\gamma \rightarrow \alpha'} < 0 \quad \dots\dots\dots(1)$$

where $\Delta G_{\text{Chem}}^{\gamma \rightarrow \alpha'}$ ($=G_{\text{Chem}}^{\alpha'} - G_{\text{Chem}}^{\gamma}$) is the chemical free energy change associated with the formation of a unit volume of martensite. The term $\Delta G_{\text{Non-Chem}}^{\gamma \rightarrow \alpha'}$ is a non-chemical free energy, and is a combination of the strain energy brought about by the volume change (ΔG_{Strain}), the increase in interfacial energy per mole of martensite particle of size r (ΔG_r) and the kinetic energy of the martensite plates (K.E., this term includes all the irreversible factors, such as heat, during the formation of martensite plates). But the transformation can only progress after a martensite embryo gains a critical size (the semi-thickness, c) given by Eq. (2):

$$c = \frac{\frac{6\Gamma_s}{d} - 4\sigma}{5\Delta G_{\text{Chem}}^{\gamma \rightarrow \alpha'}} \quad \dots\dots\dots(2)$$

Where d is the spacing between screw dislocations; σ is the specific interfacial free energy of the austenite/martensite interface, and Γ_s is strain energy per unit length of random dislocation, which is given by:

$$\Gamma_s = \frac{Gb^2}{4\pi} \left[\ln \frac{r_1}{r_0} - 1 \right] \quad \dots\dots\dots(3)$$

Here, G is the shear modulus, b is the Burger's vector, r_0 is the radius of dislocation core and r_1 is one half the mean distance between dislocations. An increase in dislocation density should decrease Γ_s , *via* a decrease in r_1 , which would reduce the value of c . The decrease of embryo size in this way leads to a higher interfacial energy per mole of martensite and consequently higher non-chemical free energy (the term $\Delta G_{\text{Non-Chem}}^{\gamma \rightarrow \alpha'}$ in Eq. (1)). This increase in non-chemical free energy can be offset by an increase in chemical free energy, *i.e.* a lower M_s temperature.

The effect of substructure has been explained by a different approach in Kajiwar's recent model,²⁵⁾ in which plastic accommodation in austenite was considered to be a very important step in the nucleation event. Based on this model, in order for a martensite plate to be brought into existence from a potential nucleation site, three kinds of deformation must take place; these are, the "lattice deformation" which generates a martensite lattice from austenite, the "lattice invariant deformation" which produces an interface with no average distortion between the two phases (*i.e.*, an invariant habit plane), and the deformation which accommodates the shape strain of the martensite plate. The shape strain of nucleating martensite (such as the shape strain of martensite plates during growth) must be accommodated by plastic deformation in austenite. Therefore any factors impeding the plastic accommodation, such as work hardening and small austenite particle size, suppress the transformation.

The present results therefore support the idea of an effect of substructure on austenite stabilization. By having some part of the substructure of the dynamically recrystallized specimen retained, the yield strength of the remaining austenite is increased, and the retained austenite volume fraction is higher.

4.2. The Effect of Pancaking

As is well established, the deformation of austenite in the no-recrystallization region results in the pancaking of austenite. The relevant microstructural features of pancaked austenite are an increase in dislocation density, formation of deformation bands, the destruction of the coherency of annealing twins with the matrix and the formation of ledge-like austenite grain boundaries. Both diffusional and non-diffusional transformations can be affected by these characteristics. This, in turn, can ultimately dictate the state of the retained austenite.

4.2.1. TMP Schedule

The effects of pancaked austenite on the retained austenite characteristics were investigated using the hot deformation schedule shown in Fig. 7. The specimens were deformed below the T_{nr} after applying double hit deformation at 1050°C. The rest of the process was the same as before, apart from the isothermal hold temperature in the two phase region. The latter was

Fig. 7. The heat and deformation schedule used to investigate the effects of pancaking.

Fig. 8. Variation of, (a) the V_{RA} of Nb-steel A, (b) the corresponding carbon content of Nb-steel and (c) the V_{RA} of Reference-steel, with strain in the no-recrystallization region.

Fig. 9. Mechanical properties of Nb-steel for different amount of strain in no-recrystallization region.

increased to 700°C in order to avoid pearlite formation during isothermal holding and cooling to bainitic region, since, as will be discussed later, undesired pearlite transformation can be encouraged by pancaking.

4.2.2. Results

Figure 8a) shows the variation of the retained austenite volume fraction with the amount of deformation in the no-recrystallization region for Nb-steel. The amount of ferrite was kept the same for all cases ($30 \pm 3\%$). As can be seen there is a maximum V_{RA} at about 20% strain. At this degree of deformation, the carbon level of the retained austenite is also maximised, (Fig. 8b)). Figure 8c) displays the variation of V_{RA} with the same pancaking conditions for the Reference-steel (the amount of ferrite in these specimens was $40 \pm 4\%$). Again, in this case, the maximum amount of retained austenite is attained at 20% strain.

The corresponding mechanical properties of the Nb-steel microstructures are illustrated in Fig. 9. The total elongation does not change systematically with retained austenite volume fraction. The ultimate strength, however, attains its maximum at about 20% deformation, coinciding with the maximum V_{RA} .

4.2.3. Discussion Concerning the Effect of Pancaking

For both steels, the results revealed a maximum for the amount of retained austenite volume fraction when the strain in the no-recrystallization region was varied. This can be related to two opposing responses of the retained austenite volume fraction to changes in the strain in the no recrystallization region. They, in turn,

can be divided into direct and indirect effects. The former is due to the strengthening of parent austenite and its stabilizing effect; the latter is a consequence of the change in the subsequent transformation characteristics of the prior austenite. These concepts will be clarified in the following paragraphs.

Since the lattice adjacent to the deformation structure is highly strained and severely distorted, these become preferred sites for ferrite and pearlite nucleation. They influence not only the kinetics of formation, but also the morphology and size of these transformation products. For a given composition, the effective nucleation area per unit volume (S_v , which includes both grain boundaries and the boundaries of severely strained regions, mostly deformation bands) increases as the finishing temperature is decreased and total reduction below the no recrystallization temperature is increased. Therefore, for a given undercooling, the nucleation frequency increases with increasing total reduction. Consequently, the chance of undesired pearlite forming increases by increasing the total strain in the no-recrystallization region.

The increase of the retained austenite by straining in the no-recrystallization region is mainly due to the following. Prestraining (beyond a critical small strain) or work hardening of austenite (including the dynamic recrystallization substructure), increases the stability of austenite by lowering the M_s temperature.^{26,27)} Thus, pancaking can be considered to decrease the M_s temperature. Furthermore, recovery during pancaking could lead to subgrains with relatively high angle boundaries. This, in turn, can lead to martensitic transformation suppression, analogous to the austenite small particle effect. Finally, at higher temperatures and concurrent straining, the higher diffusion rate and segregation of alloying elements to the grain and deformation structure boundaries can lead to the poisoning of martensite embryo formation sites via localized strengthening of the austenite.

The decrease in the retained austenite volume fraction beyond a certain level of strain in the no-recrystallization region can be related to the undesired formation of pearlite. The precipitation of Fe_3C during cooperative formation of pearlitic ferrite+carbide aggregates consumes the carbon and therefore decreases the stability of austenite. The latter leads to lower retained austenite

at the end of process. The decrease in the total elongation by strain in the no-recrystallization region can be attributed to the increase in the rate of strain induced Nb(CN) precipitation *via* pancaking.

4.3. The Effect of Finishing Temperature in the No-recrystallization Region

Since there is always some restoration (*i.e.*, dynamic and static recovery and recrystallization, depending on the state of strain and temperature) associated with hot deformation, the amount of strain energy accumulation (*i.e.* the density of dislocations in austenite) changes with the deformation-finish temperature. This, in turn, leads to a change in the kinetics of subsequent transformations. Figure 10 shows the variation of the retained austenite volume fraction with finishing temperature. The ferrite quantities for the specimens of the Nb-steel and the Reference-steel are $16\% \pm 2$ and $32\% \pm 3$, respectively. The Nb-steel does not encounter a significant change in V_{RA} after a 30% reduction at 900 and 800°C. This is not true of the Reference-steel, where the decrease in finishing temperature resulted in a 15% increase in V_{RA} .

The corresponding mechanical properties of the above microstructures of Nb-steel are shown in Fig. 11. Clearly, the total elongation is lower in the case of the specimen finished at 800°C. The ultimate strengths, however, were not changed significantly.

The above findings can be rationalized as follows. The morphology, density and distribution of deformation structures are expected to change according to the deformation conditions in the no-recrystallization region. It is clear that the density of these defects in-

creases with increasing amounts of deformation and decreasing finishing temperature. The latter can be observed in Fig. 10(b), which illustrates the effect of finishing temperature in the no recrystallization region on the retained austenite volume fraction for the Reference-steel. As can be seen, a decrease in the finishing temperature resulted in a higher retained austenite volume fraction. Since every other relevant parameter was kept constant, this effect can be qualitatively linked to the higher density of defects in the deformation structure. In the case of the Nb-steel, as can be seen in Fig. 11, the decrease in the finishing temperature did not change the retained austenite volume fraction. This is related to the presence of Nb (as a solute in austenite or precipitate) and its relatively strong effect on the retardation of the restoration processes, even at higher temperatures.

4.4. Conclusions

The effects of two important TMP parameters, *i.e.* the dynamic recrystallization substructure and the level of strain in the austenite no-recrystallization region, on the state of retained austenite and the corresponding mechanical properties of two grades of Si-Mn TRIP steels were examined. This study resulted in the following conclusions:

(1) The dynamically recrystallized austenite substructure, when retained to transformation temperatures, is more prone to retaining austenite (more V_{RA}). The variation of total elongation correlates with that of the retained austenite volume fraction change. The UTS in the rapidly cooled material, which retained the dynamically recrystallized austenite substructure, is still higher than that of the slowly cooled structure, even though the amount of retained austenite is higher in the former.

(2) TMP conditions which promoted Nb(CN) precipitation resulted in a decrease in the V_{RA} . This was attributed to a reduction of Nb in solid solution and therefore a decrease in parent austenite stability.

(3) The variation of the retained austenite volume fraction with strain in the no-recrystallization region always exhibited a maximum. For both the Nb-steel and Reference-steel, although the ferrite quantity was not the same, the maximum V_{RA} was observed at about 20% strain. The total elongation, in the case of Nb-steel, does not change systematically with the retained austenite volume fraction. The ultimate strength, however, attains its maximum at about 20% deformation, coinciding with the maximum V_{RA} .

(4) The decrease in the finishing temperature resulted in a higher retained austenite volume fraction in the Reference-steel. This effect was attributed to the higher density of defects in the deformation structure. In the case of the Nb-steel, the decrease in the finishing temperature did not change the retained austenite volume fraction. This was correlated to the presence of Nb and its effect on the retarding the restoration processes, even at higher temperatures.

Fig. 10. Variation of the V_{RA} with finishing temperature in the no-recrystallization region; (a) Nb-steel A, (b) Reference-steel.

Fig. 11. Mechanical properties of Nb-steel for different finishing temperature.

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