Loss of hot ductility at the straightening stage of the continuous casting of HSLA steel is attributed to different microalloying elements, in particular Nb. However, such elements are essential for the desired mechanical characteristics of final product. Since the chemistry cannot be altered to alleviate the problem, thermomechanical processing was studied in order to improve the hot ductility. A Nb-microalloyed steel was examined. The thermal history occurring in the continuous casting process was taken into account as well. Firstly, it was noticed that the steel has a low hot ductility after being subjected to in situ melting followed by the thermal schedule. Then, the effect of deformation applied in the vicinity of the $\delta \rightarrow \gamma$ transformation, while the thermal schedule was being executed, was investigated. Such deformation appeared to improve the hot ductility considerably. Finally, the mechanism of such improvement in the hot ductility was discussed.

KEY WORDS: continuous casting; deformation; thermal schedule; $\delta \rightarrow \gamma$ transformation; strain-induced transformation; optical microscopy; recrystallization.

1. Introduction

Loss of ductility at high temperature is a problem occurring during the straightening operation of the continuous casting of steel. This is attributed to a tensile strain of about 1–2% at a strain rate between $10^{-3}$ and $10^{-4}$ s$^{-1}$ that is generated on the top surface of slab/billet. Such loss of hot ductility is associated with transverse surface cracking taking place at surface temperatures between 700 to 1200°C. Despite numerous investigations to resolve the problem, hot cracking still persists and this interferes with productivity and cracks can even lead to scrapping of a coil. On the other hand, the integration of the continuous casting process with ‘direct rolling’ or ‘hot charging’ does not allow for any tolerance of surface cracks, since there is no interruption between casting and subsequent hot rolling processes for inspection and scarfing.

Generally, the mechanisms of the hot ductility loss in steel have been attributed to the grain boundary, or the region adjacent to the grain boundary, which can be weaker than the grain interior. This leads to strain concentrations at or near the grain boundary and, consequently, grain boundary decohesion. Through numerous investigations, it has been found that the thermal history and alloying/residual elements play roles in this respect. Therefore, some solutions based on these findings have been proposed and applied. B has been found to improve the hot ductility in the Nb-containing steel probably by altering segregation patterns and strengthening grain boundaries. However, addition of B may not always be practical since it scavenges N and so reduces the amount of N available for carbonitride precipitation, which is essential in developing the mechanical properties in the microalloyed steels. Also, B encourages ferrite formation inside grains. Moreover, the non-equilibrium co-segregation of B and Nb retards dynamic recrystallization, which can also influence the mechanical characteristics. Then, the problem is rather complicated in the microalloyed steels and to date there has been no specific and applicable solution. This can be attributed to the fact that most investigations have been centered on the alloying elements. For instance, in the Nb-containing steel, Nb has been identified to cause the loss of hot ductility, but removing or scavenging Nb from the lattice would come at the expense of the mechanical properties that are characteristic of High Strength Low Alloy (HSLA) steels. On the other hand, these laboratory findings were obtained on the reheated specimens that are rather homogeneous. Hence, they may not be fully applicable to the continuous casting process, which involves melting, solidification, and segregations of alloying elements. Specimens for the hot ductility test are usually made from steel plates that are hot-worked and therefore homogenized. However, melting specimens in situ, before evaluating the hot ductility, ensures re-solution of precipitates such as TiN and MnS, as well as the generation of the inhomogeneities typical of an as cast structure.

As a result, a solution to transverse cracking without altering the steel chemistry is always preferred. The theme of this work is to consider high temperature deformation as a means to alleviate the problem of hot ductility in the Nb-
containing steel.

2. Experiments

A Nb-microalloyed steel with the chemistry shown in Table 1 was selected. It had been continuously cast and hot-rolled down to 13 mm thickness. Cylindrical specimens with diameter of 9.5 mm and length of 125 mm were machined; because the specimens were subsequently melted, the orientation (with respect to the rolling direction) was not relevant. Specimens, 15 mm threaded in both ends, were vertically screwed into the upper and lower anvils of the deformation machine (MTS). The entire specimen length and part of the anvils were enclosed in a quartz tube. A protective atmosphere of argon containing 1% hydrogen was supplied inside the quartz tube to reduce oxidation of the iron on the specimen. An induction heating power supply was used to heat the specimen. The specimen temperature was measured, from outside the quartz tube, by a dual infrared radiation wavelength detector, which was not significantly affected by undersized objects that do not fill the field of view, bursts of steam, dust, etc. in the sight path. Also, helium was used to achieve a high cooling rate where required by the thermal schedule.

A thermal schedule similar to that experienced by the billet surface in the continuous casting process, as shown in Fig. 1, was employed. The purpose of choosing the billet surface thermal history was to generate the billet surface microstructure since transverse cracking occurs on the surface. Also, specimens were melted in situ in order to dissolve all precipitates and sulfides and to simulate the large grain size and segregation developed during billet solidification. Due to a combination of liquid viscosity, liquid surface tension, and electromagnetic field, it was possible to hold the liquid by levitation for 5–10 s at the maximum heating temperature (see below) before solidification started.

The incipient melting temperature, where melting starts during heating, was determined by performing a continuous heating tensile experiment. In this experiment, the steel was heated slowly while it was being deformed in low strain rate tension. The incipient melting temperature, which is associated with formation of a small volume fraction of liquid, was determined to be 1440°C for this steel. The second temperature to be determined was a minimum temperature at which the molten steel during heating becomes too fluid to be held by levitation and spills readily. Note that the heating rate after the incipient melting temperature was 0.5°C/s in order to minimize the temperature gradient from the surface to the center. This minimum temperature is ~1520°C for the current steel. Since it was necessary to homogenize the temperature throughout the cross section by holding for 5–10 s at the maximum heating temperature, this temperature had to be a few degrees lower than 1520°C. Therefore, temperature 1515°C was selected as the maximum heating temperature before application of the thermal schedule. Examination of some specimens fractured between 1515°C and 1520°C due to the melt spilling, and also observation of 1 mm long columnar grains arranged peripherally together with large equiaxed grains in the middle have proved melting throughout the cross section.

In this way, a melted gauge length of 15 mm was obtained. Solidification began by applying the thermal schedule and the hot ductility was isothermally evaluated at the straightening stage at 1020°C, as specified in Fig. 1, by performing a tensile experiment to fracture at a rate of $5 \times 10^{-3}$ s$^{-1}$. This strain rate is similar to the rate of straightening deformation in the continuous casting process. Reduction in area (RA) after fracture quantified the hot ductility.

Thermal expansions and contractions of specimens during the thermal schedule were continuously and automatically compensated by the MTS anvil displacement so that the load on specimens was maintained around zero, except when there was a deformation.

Firstly, the hot ductility at the straightening stage was studied in the specimen that was solely subjected to the thermal schedule. Then, the effect of deformation on the hot ductility was investigated. The steel was deformed in compression during cooling after solidification, while the thermal schedule was proceeding, i.e. deformation was incorporated with the thermal schedule. Different amounts of deformation, i.e. $\varepsilon=0.02, 0.04, 0.06, 0.08$ and 0.1, incorporated with the thermal schedule, were respectively applied to five specimens. All these deformations were initiated at ~1410°C, except $\varepsilon=0.1$, which began at ~1310°C. Also, one specimen was subjected to a $\varepsilon=0.1$ compression initiated at 1400°C during cooling. In addition, two specimens, which were firstly subjected to a $\varepsilon=0.1$ compression initiated at 1400°C, underwent a cyclic deformation started at two different temperatures as designated by A and B in Fig. 2. The cyclic deformation began with a $\varepsilon=0.1$ tension that was successively followed by a $\varepsilon=0.1$ compression, a $\varepsilon=0.1$ tension, and finally a $\varepsilon=0.1$ compression. The exchange time between the tension and compression modes was about 1 s. The purpose of the cyclic deformation was to

Table 1. Steel chemistry (wt%).

| Elements | C | Mn | S | P | Si | Cu | Ni | Cr | Nb | Mo | Al | Ti | Sn | N |
|----------|---|----|---|---|----|----|----|----|----|----|----|----|----|----|
| Nb-steel | 0.058 | 1.21 | 0.011 | 0.008 | 0.159 | 0.462 | 0.12 | 0.05 | 0.037 | 0.03 | 0.02 | 0.02 | 0.02 | 78 ppm |

![Fig. 1. Thermal schedule used to simulate the thermal condition occurring on the billet surface in the continuous casting process.](image-url)
accumulate strain since the specimen geometry did not allow applying large amounts of monotonic deformation as this would lead to specimen buckling (in compression) or necking (in tension). All the aforementioned deformations were applied at $\varepsilon = 0.0003 \, \text{s}^{-1}$ and will be referred to as pre-deformation (except the cyclic deformation) throughout this paper. In all these experiments, the specimens were unloaded immediately after achieving the desired amount of strain and were then subjected to the remaining portion of the thermal schedule toward the straightening stage. Finally, the hot ductility was evaluated at the straightening stage as described before.

Fractured specimens were studied in optical and the following electron microscopes; a JEOL-840 scanning electron microscope and Hitachi-4700 high-resolution field emission SEM. Microstructures obtained by quenching specimens at the straightening stage of the thermal schedule were etched using either ‘picral’ or a mixture of 80 mL water/28 mL 10% aqueous oxalic acid in water/4 mL 30% hydrogen peroxide. The second etchant is referred to as ‘X’ throughout the paper.

3. Results

3.1. Stress–Strain Behavior

Figure 3 demonstrates that the steel experiences a strength transition during cooling after the solidification. Obviously, the higher strengthening rate after the transition, which occurs at ‘a’ and ‘b’, cannot be solely explained by the decrease in temperature. Such transition is ascribed to the $\delta \rightarrow \gamma$ transformation and temperatures corresponding to ‘a’ and ‘b’ are 1360°C and 1330°C, respectively. The lower $\delta \rightarrow \gamma$ transformation temperatures than those in the equilibrium C–Fe phase diagram is mainly attributed to the high cooling rate, i.e. 10°C/s, associated with the thermal schedule.

The effect of $\varepsilon = 0.1$ precompression going through the $\delta \rightarrow \gamma$ transformation is shown in Fig. 4. As can be seen, such deformation improved the mechanical properties as compared to the effect of the thermal schedule alone, i.e. no predeformation, Fig. 4(b).

Figure 5(a) shows the stress–strain curves owing to different amounts of precompression applied through the transition region. The precompression curves for $\varepsilon = 0.02$ and 0.1, which were respectively applied before and after the transition, are also included. The corresponding tensile stress–strain curves at the straightening stage are shown in Fig. 5(b). The tensile deformation curve of the specimen subjected to the thermal schedule alone (TS) is also included for comparison. It is observed that predeformation before the transition, specimen I, is deleterious to the mechanical characteristics but beneficial if applied during the transition, specimen II, both compared with TS in Fig. 5(b). Even though predeformation after the transition, specimen V, is not as effective as that applied during the transition, it still appears somewhat beneficial, as compared to TS in Fig. 5(b). Note that specimens III and IV, not presented in Fig. 5(b), also exhibited a similar tensile behavior to that of specimen II.

The stress–strain curves, at the straightening stage, of the specimens subjected to the cyclic deformations (Fig. 2) are
plotted in Fig. 6. As can be seen, the specimens display almost similar behaviors to that of the thermal schedule alone. However, it should be indicated that these deformation curves are not real representatives of the hot ductility.

3.2. Hot Ductility Evaluation

The results of the hot ductility assessment at the straightening stage are summarized in Table 2. As can be seen, application of predeformation changes the hot ductility substantially. Such predeformation when applied before the transition is somewhat harmful to the hot ductility, but beneficial when performed through the transition region. Nevertheless, the beneficial effect of predeformation constitutes the subject of this article.

Even though the predeformation after the transition improved the hot ductility, the highest hot ductility was achieved when such predeformation was preceded with a predeformation applied during the transition, comparing specimen V in Fig. 5 with the specimen subjected to \( \varepsilon = 0.1 \) continuous precompression (Fig. 4), in Table 2. In other words, the effect of predeformation after the transition on the hot ductility improvement can be enhanced if it follows a predeformation applied during the transition. This can be clearly perceived by comparing the RA values of 28, 23, and 41% in Table 2. It has been also noticed that the predeformation rate is influential and an optimum combination of strain and strain rate led to 53% RA.17)

3.3. Examination of Microstructures

Fracture surface of the specimen subjected to the thermal schedule alone, shown in Fig. 7(a), displays both ductile and brittle (intergranular) characteristics. Examination of the microstructure quenched immediately after fracture revealed that the failure was associated with grain boundary

| Deformation schedule | RA%  |
|----------------------|------|
| specimen TS in Fig. 4(b) | 15   |
| Fig. 5 (specimen I) | 13   |
| Fig. 5 (specimen II) | 27   |
| Fig. 5 (specimen III) | 28   |
| Fig. 5 (specimen IV) | 23   |
| specimen in Fig. 4(a) | 41   |
| Fig. 6 (specimen A) | 40   |
| Fig. 6 (specimen B) | 40   |
microcracks, Fig. 7(b). The matrix is basically bainite with some martensite islands. The grain size varies from 282 to 891 μm with an average ~500 μm.

Application of predeformation during the δ→γ transformation suppressed grain boundary cracking to some extent. This can be deduced after comparing the fracture surfaces in Fig. 7(a) and Fig. 8(a) and their respective microstructures in Figs. 7(b) and 8(b). Fracture features and grains are also finer in Fig. 8. The important point is that grain boundary cavities are much smaller, isolated, and have oval shapes, instead of large, elongated, and linked-up cavities that resulted in propagation of cracks along grain boundaries, as shown in Fig. 7(b). Oval cavities are also observed in grain interiors, Fig. 8(b).

All specimens I–V presented in Fig. 5 fractured also through grain boundary decohesion. Two typical examples of these specimens are shown in Fig. 9. In these specimens, cracks initiated mainly at grain boundary triple junctions and propagated along the boundaries, leading to complete separation of grains. The microstructures of specimens I and V, quenched after fracture at the straightening stage, displayed the largest average grain size, i.e. ~500 μm, whereas specimens II through IV exhibited fairly identical average grain sizes, i.e. ~350 μm.

The finest grain size, i.e. ~220 μm, was obtained when the steel was subjected to ε=0.1 predeformation during the δ→γ transformation, Fig. 10. Note that the microstructure in Fig. 10 was quenched at the straightening stage just before performing the tensile experiment. As can be seen, the austenite grains are decorated with grain boundary ferrite. On the contrary, the microstructure developed during the thermal schedule alone and quenched at the straightening stage, just before the tensile experiment, exhibited a grain size similar to that shown in Fig. 7(b), i.e. ~500 μm or larger.
4. Discussion

In the previous article, it was explained that grain boundary sliding is the dominant mechanism of embrittlement at high temperatures in the Nb-microalloyed steel. As well, it was discussed that how presence of Nb, either as solute or as carbide precipitate, can lead to such failure at low strains.

As Table II showed, predeformation during the $\delta \rightarrow \gamma$ transformation increased the hot ductility by 26% RA. Comparing the microstructures just before the tensile test at the straightening stage (Fig. 10) and immediately after fracture (Fig. 8(b)), it appears that recrystallization was under way during the tension. With regard to the mechanism of failure which is grain boundary sliding, the dynamic recrystallization could have contributed to the hot ductility improvement in the steel. The microcrystallized microstructure, Fig. 8(b), displays grain sizes varying from 9.5 to 42 $\mu$m with an average of 32 $\mu$m.

Although the specimen that underwent $\varepsilon=0.1$ precompression during the $\delta \rightarrow \gamma$ transformation demonstrated grain refinement after fracture at the straightening stage (Fig. 8(b)), the occurrence of dynamic recrystallization is not conclusive since there is no 'hump' or stress peak, Fig. 4(b), which is the characteristic of dynamic recrystallization. In fact, even though dynamic recrystallization begins before the stress peak occurs, the presence of stress peaks on constant strain rate flow curves is often considered to be the most only indication of the initiation of dynamic recrystallization. Flow curves without well-defined stress peaks, but which display a steady state, are generally believed to pertain to dynamic recovery as the only restoration mechanism. Furthermore, dynamic recrystallization takes place in many materials even though no clearly defined stress peaks are observed in laboratory flow curves. Examples of such materials include Nb-microalloyed low carbon and austenitic stainless steels.

From a consideration of thermodynamic instability, Poliak and Jonas established a critical kinetic condition for the initiation of dynamic recrystallization. This condition, which is based on the dislocation strain energy, corresponds to an inflection point on the $\theta-\sigma$ curve, where $\theta$ is strain hardening rate and $\sigma$ is stress. In other words, dynamic recrystallization starts when $\partial(\partial \theta/\partial \sigma)/\partial \sigma=0$. Gottstein et al. recently studied this criterion with regard to microstructural instability in terms of the substructure. As their predicted results showed a good agreement with the criterion proposed by Poliak and Jonas, i.e. the inflection point model, they suggested that one internal parameter is dominant, namely the one that most strongly affects the flow stress at large strains. In this respect, Gottstein et al.'s model approaches the inflection point model, so that, essentially, the flow stress is represented in terms of the total dislocation density.

The inflection point model was recently employed to study dynamic recrystallization in compression experiments. Even though it should be also applicable in tension experiments, caution should be exercised when grain boundary decohesion and deformation localization are concerned. This is because such phenomena can affect the flow curve, leading to a strain hardening behavior that is not solely dependent on dislocation characteristics.

**Figure 11** shows the strain hardening behavior of the specimen subjected to $\varepsilon=0.1$ precompression during the $\delta \rightarrow \gamma$ transformation (Fig. 4). Note that the deformation curve shown in Fig. 4(b) was first best approximated by a polynomial of appropriate order to filter out the load cell noises. This is necessary since load cells generally do not render exact instantaneous values, but reveal very short range fluctuations around the actual load, leading to iteratively positive and negative strain hardening. As can be seen in Fig. 11, the rate of decrease in $\theta$ decelerates with increase in flow stress until the critical stress ($\sigma_c$) corresponding to the onset of dynamic recrystallization is reached. At this point, $\partial(\partial \theta/\partial \sigma)/\partial \sigma=0$. With increasing strain, $\theta$ again starts to decrease rapidly at an increasing rate. This quantity tends to infinity as the flow stress approaches its peak value, $\sigma_p$, at $\theta=0$.

In addition, since $\sigma=\partial(\partial \theta/\partial \sigma)$, the $\ln \theta-\varepsilon$ plot must also exhibit an inflection at the onset of dynamic recrystallization. Accordingly, the critical strain for dynamic recrystallization can be determined as shown in Fig. 12.

However, in contrast to compression, it should be noticed that $\sigma_p$ in tension does not necessarily correspond to dynamic recrystallization. There are striking qualitative similarities in the $\theta-\sigma$ behavior associated with dynamic recrystallization and that displayed when flow localization occurs in tension. The onset of dynamic recrystallization and flow localization (necking) are both manifested through the appearance of an inflection point in the $\theta-\sigma$ curve. This issue has been considered in more detail in ref 19 and a criterion has been proposed to distinguish the occurrence of dynamic recrystallization from that of flow localization with regard to the inflection point in the $\theta-\sigma$ curve.

Considering the microstructure in Fig. 8(b) and the strain hardening behavior in Figs. 11 and 12, the improvement of hot ductility in the steel is ascribed to the occurrence of dynamic recrystallization at the straightening stage. During dynamic recrystallization, grain boundaries migrate and microvoids initially formed at grain boundaries are isolated from the boundaries. Since the isolated microvoids cannot coalesce easily at grain boundaries, grain boundary decohesion is retarded and fracture occurs at greater strains.
Dynamic recrystallization at the straightening stage is initiated because austenite grains were refined during the precompression in the vicinity of the $\delta \rightarrow \gamma$ transformation, Fig. 10. Such grain refinement can be explained as follows.

The RA values in Table 2 show that there is an improvement in the hot ductility from 15% to 27%, when predeformation goes through the transformation region, comparing specimen II in Fig. 5 with specimen TS in Fig. 4(b). In addition, no further improvement occurred after an increase in predeformation to $\varepsilon = 0.08$, i.e. specimen IV in Fig. 5. Hence, it appears that 13% (the difference between 28% and 15%) (Table 2), is the maximum improvement attainable solely by the accelerated $\delta \rightarrow \gamma$ transformation. This is attributed to a smaller austenite grain size resulted from the strain-induced transformation. On the other hand, predeformation after the $\delta \rightarrow \gamma$ transformation resulted in only 8%, (the difference between 23% and 15%) improvement in the hot ductility, specimen V in Fig. 5. Such improvement can only be attributed to the elimination of solidification shrinkages by precompression since no grain refinement was observed in the microstructure (Fig. 9(b)). Then, the sum of improvements due to the predeformation during and after the $\delta \rightarrow \gamma$ transformation is 21% (8% + 13%), provided these two predeformation schedules have been applied separately to two different specimens. This is less than the minimum 26% (the difference between 41% and 15%, Table 2) improvement in the hot ductility after the application of 0.1 continuous precompression initiated before and continued after the transformation (Fig. 4(a)). Therefore, since the grain size in specimen IV of Fig. 5, i.e. ~350 µm, is larger than the grain size in Fig. 10, i.e. 220 µm, there appears to have been a recrystallization after the transformation in the latter. In other words, the smaller austenite grain size resulted from the accelerated transformation was further refined by the subsequent recrystallization. Even though dynamic recrystallization has played the major role in the second stage of the grain refinement process, the possible contribution of metadynamic/static recrystallization cannot be ruled out, especially since the temperature was high enough to overcome the activation energy in the given time frame. The very small grains in Fig. 10 can be ascribed to this phenomenon.

Since the grain refinement due to the predeformation incorporated with the $\delta \rightarrow \gamma$ transformation improved the hot ductility, it was reasonable to anticipate more improvement with increasing grain refinement. This was the reason for application of the cyclic deformations described in Fig. 2 in order to accumulate strain and stimulate more recrystallization. Nevertheless, no improvement more than that achieved solely by the precompression during the transformation was attained, Table 2. This implies that dynamic recrystallization was most probably not triggered during the cyclic deformations A and B (Fig. 2), even though the total strain was 0.4. The reason for this can partly be attributed to the Bauschinger effect in terms of dislocation generation during plastic flow. As well, it has been found that reversal deformation can delay dynamic recrystallization. Therefore, the amount of strain required to initiate dynamic recrystallization during cyclic deformation is larger than that required during monotonic deformation.

On the other hand, there is a possibility that the cyclic deformation applied around 1200°C (schedule A in Fig. 2) has produced some microstructural defects such as cavities and cracks at grain boundaries. This can be true especially about the tensile half-cycles. It was noticed that such defects contributed to low hot ductility, i.e. 17% RA, at 1200°C. This indicates that tension is deleterious at this stage of the thermal schedule. Therefore, it is reasonable to postulate that dynamic recrystallization would be underway provided the steel was deformed in compression monotonically, rather than cyclically. However, tensile deformation in the vicinity of 1100°C is not expected to produce microstructural defects at low strains since the steel, subjected to the thermal schedule alone, has already shown a high ductility, i.e. 40% RA, at this temperature. Therefore, the question is whether the necessary condition for dynamic recrystallization has been provided by the cyclic deformation B (Fig. 2).

Using equations and $Z = \dot{\varepsilon} \exp(Q_{\text{def}}/RT)$ and $\varepsilon_0 = 2.78 \times 10^{-13} d_0^{0.5} Z^{0.16}$, where $Z$ is Zener–Hollomon parameter, the critical strain of dynamic recrystallization at 1100°C and $\dot{\varepsilon} = 3 \times 10^{-3} \text{s}^{-1}$, in the specimen predeformed during the $\delta \rightarrow \gamma$ transformation, i.e. $d_0 = 220 \mu m$, is calculated to be 0.12 ($Q_{\text{def}}$, activation energy of deformation, is 313 kJ/mol for the intended steel). Since this strain is greater than each half-cycle strain, then, for the same reasons as explained above, dynamic recrystallization was barely triggered by cyclic deformation B. Hence, there was no further...
improvement in the hot ductility. With regard to these findings, the strategy to enhance the hot ductility at the straightening stage, ~1020°C, in the steel could involve a precompression in the proximity of the δ→γ transformation region followed by another compression in the proximity of 1 100°C over the last step of the thermal schedule. The purpose of the first precompression is to reduce the critical strain in order for dynamic recrystallization to commence during the second compression at a lower strain than the critical strain for the as cast structure. Using the aforementioned equation for the second compression, \( \varepsilon = 3 \times 10^{-3} \text{ s}^{-1} \) and \( T = 1 100^\circ \text{C} \), \( Z \) is \( 2 \times 10^9 \text{ s}^{-1} \). Figure 8(b) shows that grains were refined at the straightening stage when \( Z = 2 \times 10^9 \text{ s}^{-1} \) and \( d_0 \approx 220 \mu \text{m} \). As well, the steel was dynamically recrystallized during tension at the straightening stage temperature 1 100°C, i.e. \( Z = 4.14 \times 10^9 \text{ s}^{-1} \) and \( d_0 = 586 \mu \text{m} \).\(^{17} \) Moreover, it was found that the steady state grain size is about 50 \( \mu \text{m} \) for \( Z = 5 \times 10^{10} \text{ s}^{-1} \) in a Nb-containing steel.\(^{27} \) Based on these observations and the typical \( Z - D_s \) (\( D_s \) is the steady state grain size) profile, a \( Z - D_s \) plot for the steel studied here would resemble the graph shown in Fig. 13. Therefore, the steel was most probably in the grain refinement region for the condition of \( Z = 2 \times 10^8 \text{ s}^{-1} \) and \( d_0 = 220 \mu \text{m} \), i.e. cyclic deformation schedule B in Fig. 2. This condition is identified by ‘X’ in Fig. 13. Such grain refinement can be definitely attained if \( Z \) is increased to \( 2 \times 10^9 \text{ s}^{-1} \) or more, i.e. increasing the strain rate. Even though both compression and tension would behave alike in this manner, tension should be avoided since it is likely to produce grain boundary microcracks. This is why compression is advantageous over tension. Figure 13 also suggests that the precompression during the \( \delta \rightarrow \gamma \) transformation can be eliminated if large amounts of strain are allowed, with regard to industrial constraints, to be applied in the vicinity of 1 100°C.

5. Conclusions

Application of deformation in the vicinity of the \( \delta \rightarrow \gamma \) transformation appeared to improve the hot ductility at the straightening stage of the continuous casting process. The beneficial effect of such deformation is enhanced if it is followed by some more deformation after the transformation. This improvement is attributed to a grain refinement which subsequently stimulates dynamic recrystallization at the straightening stage.

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