



# Hot work modelling of two equivalent low carbon strip steels produced, respectively, by the cold charge route and by the hot charge route

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## Synopsis

Mean flow stress analyses from plant mill logs for the conventionally produced hot rolled low carbon steel from the Ispat Iscor Vanderbijlpark strip producing plant, indicated a higher temperature sensitivity than that found for the equivalent SAE 1006 strip steel produced in the compact strip plant of Saldanha Steel. This was confirmed quantitatively through detailed hot deformation studies on an as-cast structure as well as on a finer grained condition on the same steel. In the conventionally produced steel, activation energies for hot deformation of 309 kJ.mol<sup>-1</sup> and 388 kJ.mol<sup>-1</sup> for the two respective starting structures were found, both values extending over the entire normal hot deformation temperature range of 1140 to 900°C. These and other experimentally determined constitutive hot deformation constants were applied to the hot rolling austenite grain size development model used earlier. From this, it is predicted that dynamic recrystallization occurs only in the first two break-down passes at the Vanderbijlpark plant and thereafter static recrystallization occurs after exiting from all of the remaining ten or eleven roughing and finishing passes. This is significantly different from what was predicted earlier for the Saldanha Steel plant with its lesser number of passes where static recrystallization is predicted to occur only in the last three passes. The effects of this process difference and other observed product differences between the two types of product are discussed.

Keywords: Low carbon steel, hot rolling of strip steel, rolling mill log analyses, austenite grain size, grain size modelling.

## Introduction

In South Africa Ispat Iscor produces hot rolled low carbon strip steel at both its Vanderbijlpark (VDB) as well as at its Saldanha Steel (SS) works. The former is a conventional strip producing plant using the cold charged route (CCR) in which the continuously cast slab is allowed to cool down to below the  $\gamma$  to  $\alpha$  transformation temperature before entering the reheating furnace for hot rolling. The Saldanha Steel plant uses the hot charged route (HCR) in a compact strip plant where the cast slab still in the austenitic condition, enters the roller hearth furnace for reheating directly before hot rolling commences. Although both products are nominally of very similar composition, relatively small but nevertheless significant

differences existed in their hot rolled microstructures, mechanical properties,  $\gamma$  to  $\alpha$  transformation behaviour and also in their static recrystallization behaviour after cold rolling<sup>1</sup>.

Apart from the differences between the CCR and the HCR operations at the start of hot rolling, other process differences also exist between the two plants. These include the number of hot rolling passes, with the Saldanha Steel plant with its compact design having far fewer hot rolling passes than the conventional VDB plant but also starting with a thinner slab thickness of 75 or 90 mm versus the 240 mm of VDB. Furthermore, Saldanha Steel generally achieves a lower sulphur content than the equivalent VDB product in its steelmaking. Greater or lesser use of electric arc furnace (EAF) melting in any of the two types of plant may also result in differences in the nitrogen content.

In view of the many differences between the two processes, it is, therefore, unlikely that hot rolled HCR strip from a compact strip mill would necessarily have the same properties as the nominally equivalent CCR material. Very few systematic studies of differences between the low carbon hot rolled strip steels of the two types of plant have been published. The most extensive comparative study is probably the quantitative hot deformation work of Muojekwu *et al.*<sup>2</sup> while Frawley *et al.*<sup>3</sup> simulated both the HCR and CCR process through hot rolling of laboratory cast ingots of low carbon manganese steels with varying sulphur contents. Their findings have been reviewed in the earlier work<sup>1</sup> on the hot deformation study of the Saldanha Steel SAE 1006 where both of these sets of authors have found some significant differences in the expected microstructures as well as in the hot deformation behaviour between the two types of products.

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## Hot work modelling of two equivalent low carbon strip steels

In earlier work<sup>1</sup> quantitative hot deformation studies on as-cast as well as on laboratory hot rolled SAE 1006 from Saldanha Steel (henceforth called HCR steel) showed a two-stage hot deformation process with a relatively low activation energy of 238 kJ.mol<sup>-1</sup> for the as-cast steel with an austenite starting grain size of 277 µm at testing temperatures above 1000°C and a higher activation energy of 330 kJ.mol<sup>-1</sup> for the finer grained laboratory hot rolled steel with a grain size of 26 µm and tested below 1000°C. Introducing these and other determined constitutive hot working constants into modelling equations for the austenite grain size, predicted that under actual plant conditions dynamic recrystallization (DRX) occurred within the rolls during the initial four passes R1 to F2 in the Saldanha Steel plant, whereas static recrystallization (SRX) occurred during the interpass times or after exiting in the final finishing passes of F3 to F5. Predictions of the modelled final ferrite grain size in the end product agreed reasonably well with the actual grain size.

In this work, the comparative investigation between the two types of product is taken one step further through a comprehensive study of the hot working behaviour of an as-cast and also of a laboratory hot rolled low carbon strip steel from VDB and compared to the earlier results on the equivalent SAE 1006 strip steel from Saldanha Steel. The two products will henceforth be distinguished as CCR and HCR products, respectively. A current study of the austenite to ferrite transformation behaviour of the two types of product has also revealed considerable thermodynamic and kinetic differences and will be the subject of another publication.

### Overview of the hot rolling processes for strip steel at Ispat Iscor Vanderbijlpark and Saldanha Steel

Apart from the different charging routes, other differences between the two respective hot rolling processes include:

- ▶ the starting slab thicknesses of 240 mm at VDB versus 75 or 90 mm at SS;
- ▶ up to seven roughing passes (R1 to R7) in a double reversing mill at VDB versus the two (R1 and R2) in a tandem mill at SS
- ▶ seven finishing passes (F1 to F7) at VDB versus the five (F1 to F5) at SS, in both cases in tandem finishing mills
- ▶ differences in compressive strain per pass with at VDB: starting low at about 0.2 to 0.3 at R1 and R2 then rising to about 0.4 to 0.5 at R6, increasing to 0.8 to 0.9 at F1, then gradually dropping down to 0.4 at F5 and finally 0.2 to 0.1 at F6 and F7, whereas at SS, between 0.7 to 0.8 per pass for R1 to F2 and between 0.3 and 0.45 for F3 to F5
- ▶ differences in compressive strain rate per pass with at VDB: starting at about 3 s<sup>-1</sup> at R1 and increasing up to 500 s<sup>-1</sup> at F7, whereas at SS, varying from 7 s<sup>-1</sup> for R1 to about 150 s<sup>-1</sup> for F5.

### Mean flow stress of the CCR steel calculated from mill logs

The mean flow stress was calculated as described earlier<sup>1</sup> for the hot rolling of CCR slabs in the finishing line from a number of mill logs kindly supplied by Ispat Iscor. These slabs had all been melted via the BOF route and had nitrogen contents typically in the range 40 to 60 ppm. In general, the

last finishing pass at F7 at the VDB strip mill applies a very small reduction and this MFS value was discarded. The same calculations for the roughing line at VDB were somewhat inconclusive, probably due to the intermittent rolling and waiting operation of the slab through the double reversing mill.

The calculated MFS values were corrected to a constant strain of 0.4 and a strain rate of 5 s<sup>-1</sup> per pass through the Misaka<sup>4</sup> equation and are compared in Figure 1 with the earlier equivalent trend lines from the HCR steel. Although there is some significant scatter in the calculated points (this is common with mill log analyses), the trend appears to show a larger temperature sensitivity of the MFS for the CCR steel than was found earlier for the DRX regime of the HCR steel. These two sets of data points were compared statistically<sup>5</sup> through an H<sub>0</sub> hypothesis and the two sided t-test and it was confirmed that the slopes of the two trend lines differ significantly at a confidence level of greater than 99.99%. This is already a strong indication that the activation energy  $Q$  for hot deformation of the CCR steel may be significantly higher than that found earlier for the DRX regime in the HCR steel.

Some published MFS values from other low carbon strip steel producers<sup>6</sup>, also corrected to the same strain and strain rate per pass as above, are compared to the values for the two Ispat Iscor strip steels in Figure 2. It is significant to note that the calculated MFS values from mill load data of these three strip producers also indicate a larger temperature sensitivity than was found in the HCR steel. In fact, the temperature sensitivities of the hot rolling process/products of these producers appear to be even higher than that of the CCR steel of VDB. The absolute MFS values of both the CCR and HCR steels of Ispat Iscor are, however, notably lower than those of the other low carbon strip steel producers.

In view of the strong indication of a possibly higher activation energy  $Q$  for the hot deformation of the VDB CCR steel than its equivalent SS HCR steel, a series of hot deformation studies were planned for the former. To enable a comparison, the tests were equivalent to those done formerly on the HCR steel.

### Austenitization behaviour

For the purpose of the hot deformation studies, Ispat Iscor kindly supplied some as-cast low carbon steel intended for strip production and melted via an EAF route with the composition shown in Table 1. The composition of the earlier tested HCR steel from Saldanha Steel is also repeated for comparison purposes.

The higher nitrogen content in this CCR slab was due to an EAF melting route for this particular cast rather than the more usual BOF melting route. BOF melted casts with lower nitrogen contents, which is the usual melting route for the CCR slabs from VDB, were, however, used for the mill log analyses in Figure 1. The significantly higher sulphur content of 140 ppm in the CCR steel against the less than 2 ppm sulphur in this particular HCR steel should also be noted. The CCR steel was similarly tested in two initial conditions, i.e. as-cast with a pronounced columnar ferrite grain structure to represent the early roughing stages in the strip mill and secondly, with a finer grain size of 34 µm to represent later hot rolling stages. This was obtained through first hot rolling the same as-cast steel in a laboratory hot rolling mill, as described earlier<sup>1</sup>.

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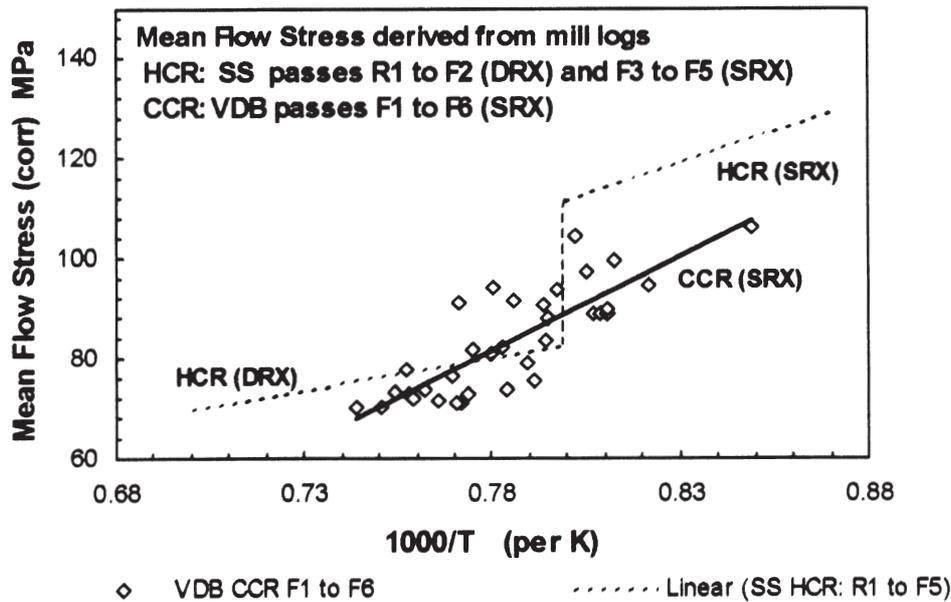


Figure 1—Corrected mean flow stress (in MPa) derived from mill logs for finishing passes F1 to F6 at VDB as a function of the reciprocal absolute temperature  $1/T$  (in  $1000\text{ K}^{-1}$ ) of the CCR low carbon steel (solid line and open symbols) and compared to the earlier trend lines<sup>1</sup> of the equivalent HCR low carbon steel of Saldanha Steel

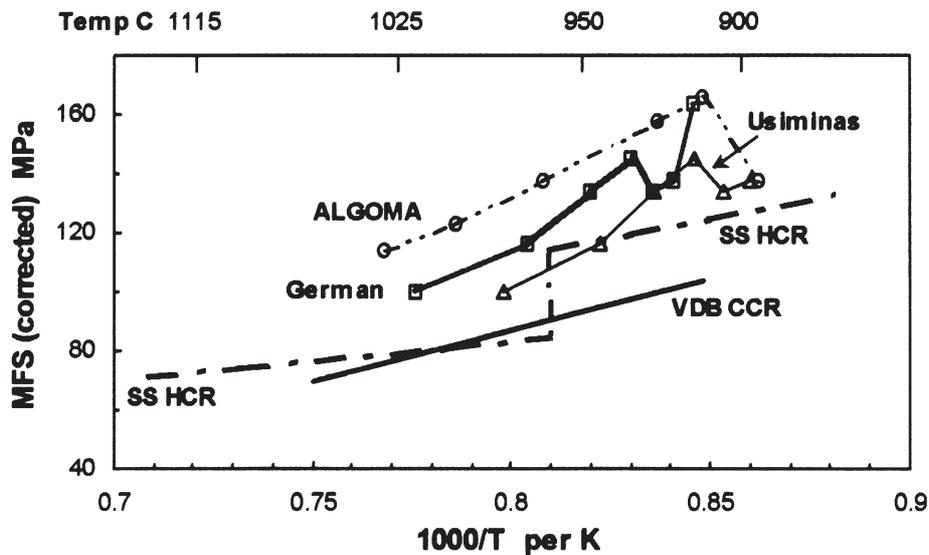


Figure 2—Corrected mean flow stress (in MPa) trend lines from mill log analyses for the CCR and the HCR steels of Ispat Iscor and three other producers of low carbon strip steels<sup>6</sup> as a function of the reciprocal absolute temperature  $1/T$  (in  $1000\text{ K}^{-1}$ )

Table 1

Composition of the as-cast CCR steel from VDB and compared to the earlier HCR strip steel SAE 1006 from Saldanha Steel

Steel	Cast No.	%C	%Mn	%Si	%Al (tot)	%Cr	%Ni	%V	%Ti
VDB CCR	Y206727	0.051	0.224	0.014	0.05	0.027	0.011	0.002	0.002
SS HCR	7037	0.038	0.25	0.02	0.045	0.008	0.001	0.007	0.001
Steel	Cast No.	%Nb	%Cu	%Mo	%P	%Ca	S ppm	N ppm	B ppm
VDB CCR	Y206727	0.001	0.026	0.001	0.008	NA	140	104	4
SS HCR	7037	0.002	0.009	0.001	0.005	0.002	1.8	65	NA

NA = Not analysed

## Hot work modelling of two equivalent low carbon strip steels

The starting austenite grain size  $D_0$  after the soaking time of 3 minutes at 1200°C to be used in the hot deformation tests, was determined by the *in situ* austenite carburization technique described earlier<sup>1</sup>. The ferrite and the decorated austenite grain structures are shown in Figure 3 for the two respective starting conditions of the CCR steel.

For comparison purposes, the equivalent optical micrographs for the HCR steel are shown in Figure 4, and already some significant differences in the austenitization behaviour may be noted as summarized in Table II.

In the as-cast starting condition the ferrite phase in the CCR steel has a more pronounced and sharply defined columnar structure than its counterpart in the HCR steel. This former grain structure in the CCR steel transforms to a visibly irregular austenite grain boundary structure with a smaller grain size of 124  $\mu\text{m}$  at 1200°C than is the case with the HCR steel. The latter presents a much larger austenite grain size of 277  $\mu\text{m}$  with a structure that is noticeably more regular and uniform, with relatively straight grain boundaries.

In the finer grain sized starting conditions, the CCR steel once again presents a highly irregular austenite grain structure but now with a larger austenite grain size of 50  $\mu\text{m}$  versus the more regular and uniform austenite grain size of 26  $\mu\text{m}$  of the HCR steel. Significant differences in the austenite to ferrite transformation behaviour upon cooling have also been found in a parallel investigation to this work and will be published later.

### Hot work tests on the steel processed by the CCR

The hot compression deformation tests were performed on a modified Gleeble 1500™ machine as described earlier<sup>1</sup> and the constitutive constants were derived from the Figures 5(a)

to (c) for the as-cast starting condition and Figures 6(a) to (c) for the fine grained starting condition. These figures plot the two main hot deformation equations of, firstly, an expression that relates the steady state flow stress to the absolute temperature and the strain rate through the so-called universal creep and hot deformation equation<sup>7,8</sup> and, secondly, an expression that predicts the minimum or critical strain  $\epsilon_c$  required to initiate DRX through the peak strain  $\epsilon_p$  with  $\epsilon_c = 0.8 \epsilon_p$  for low carbon steels<sup>9</sup>:

$$\exp(Q/RT) = Z = A_3 \left\{ \sinh(\alpha \sigma_{ss}) \right\}^n \quad [1]$$

$$\epsilon_c = A_1 (D_0)^m Z^q \quad [2]$$

where  $\dot{\epsilon}$  is the strain rate in  $\text{s}^{-1}$ ,  $Q$  is the activation energy of hot deformation in  $\text{J}\cdot\text{mol}^{-1}$ ,  $T$  is the absolute temperature in K,  $R$  is the universal gas constant in  $\text{J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$ ,  $A_3$  and  $A_1$  are structure factors which are assumed to be constant if one applies steady state flow stress conditions,  $\alpha$  is a material constant,  $\sigma_{ss}$  is the steady state flow stress in MPa,  $D_0$  is the initial austenite grain size (in  $\mu\text{m}$ ) at the start of the hot deformation test and  $Z$  is the Zener-Holloman parameter<sup>10</sup> with units  $\text{s}^{-1}$ . The dimensionless exponents  $n$  (also called the stress sensitivity),  $m$  and  $q$  are normally considered to be constant with temperature.

The experimentally determined hot deformation constitutive constants for the two starting conditions of the CCR steel are compared in Table III with those found earlier for the HCR steel.

In the as-cast condition the CCR steel has an activation energy  $Q$  for hot deformation of  $309 \pm 10 \text{ kJ}\cdot\text{mol}^{-1}$  which is very close to the accepted value of about  $312 \text{ kJ}\cdot\text{mol}^{-1}$  often quoted<sup>1</sup> in the literature for low carbon steels, mostly processed by the CCR process. In the fine grained condition,

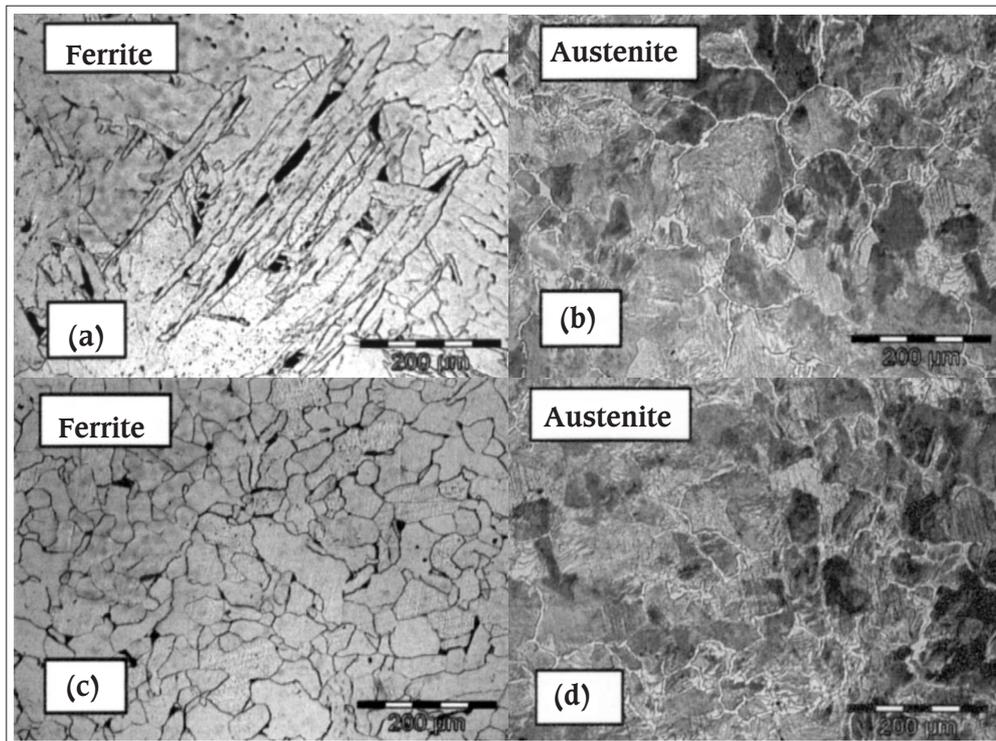


Figure 3—Optical micrographs of the CCR low carbon steel with (a) the as-cast ferrite structure, (b) the austenite as-cast structure after solution treatment, (c) the laboratory hot rolled ferrite grain structure and (d) the laboratory hot rolled austenite grain structure after solution treatment. The solution treatment consisted of 3 minutes at 1200°C in both cases. The etchant was 5% Nital

## Hot work modelling of two equivalent low carbon strip steels

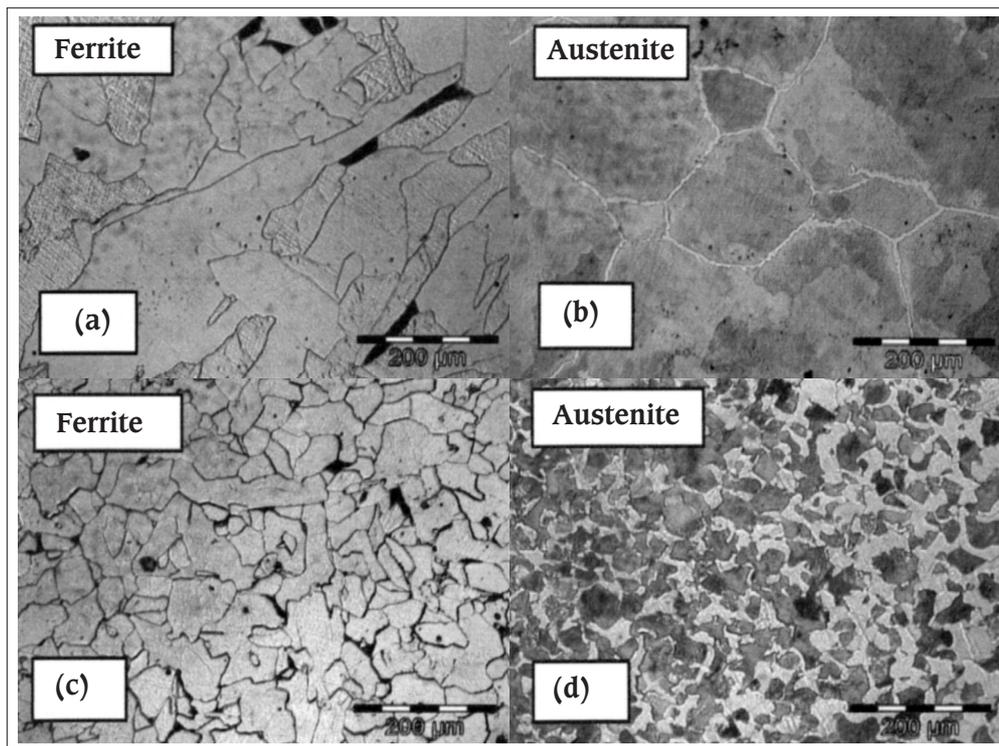


Figure 4—Optical micrographs of the HCR low carbon steel with (a) the as-cast ferrite structure, (b) the austenite as-cast structure after solution treatment, (c) the laboratory hot rolled ferrite grain structure and (d) the laboratory hot rolled austenite grain structure after solution treatment. The solution treatment consisted of 3 minutes at 1200°C in both cases. The etchant was 5% Nital

Table II

Effect of the initial ferrite grain size structure and steel origin on the austenite grain size  $D_0$  after solution treating at 1200°C for 3 minutes

Steel	Initial ferrite grain structure	Austenite grain size $D_0$	Comments
VDB CCR	As-cast	124 $\mu\text{m}$	Finer and more pronounced ferrite columnar grains in the CCR steel. Non-uniform austenite structure in the CCR and relatively uniform austenite structure in the HCR steel.
SS HCR	As-cast	277 $\mu\text{m}$	
VDB CCR	Hot rolled with 34 $\mu\text{m}$ ferrite grain size	50 $\mu\text{m}$	No observable difference in starting ferrite structure between the two steels. Non-uniform austenite structure in the CCR and relatively uniform austenite structure in the HCR steel.
SS HCR	Hot rolled with 33.7 $\mu\text{m}$ ferrite grain size	26 $\mu\text{m}$	

however, the measured activation energy rises to  $388 \pm 10$   $\text{kJ}\cdot\text{mol}^{-1}$  which now quantitatively confirms the higher temperature sensitivity already suggested by the MFS mill log analyses from Figure 1 for the finishing line of this steel. Secondly, the previous increase in activation energy from a lower value to a higher value as one moves from a coarse as-cast grain structure to a finer one in the case of the HCR steel, is echoed by this CCR steel, although the absolute values of  $Q$  differ somewhat between the two steels. This raises the distinct possibility that the activation energy and, therefore, the fundamental mechanism for softening, undergoes some change between an as-cast grain structure and a finer hot rolled grain structure. This has definite implications for modelling studies of the austenite grain size during hot rolling because an as-cast structure would probably be only applicable during the first one or two roughing stages, and with a finer grain size and hence a different activation energy during all subsequent roughing and finishing passes in strip rolling.

### Austenite grain size modelling during hot rolling at the VDB strip mill

The constitutive hot working constants for the CCR steel from Table III were introduced into the comprehensive austenite grain size development model used earlier on the HCR steel, whereas all other constants for SRX and grain growth that were not covered by Table III, were considered to be identical for the two steels and taken from the literature<sup>1</sup>. This soon revealed an interesting observation on the CCR steel, i.e. softening by DRX is likely to take place only in the first two roughing stages R1 and R2, with softening by SRX taking place in all remaining roughing and finishing stages. This is quite different from what was found with the HCR steel at the Saldanha Steel plant where softening by DRX was predicted to take place over the first four stages of R1 to F2 and softening by SRX only over the last three stages of F3 to F5. The modelled austenite grain size during hot rolling for the two steels in their respective strip mills is shown in Figure 7.

# Hot work modelling of two equivalent low carbon strip steels

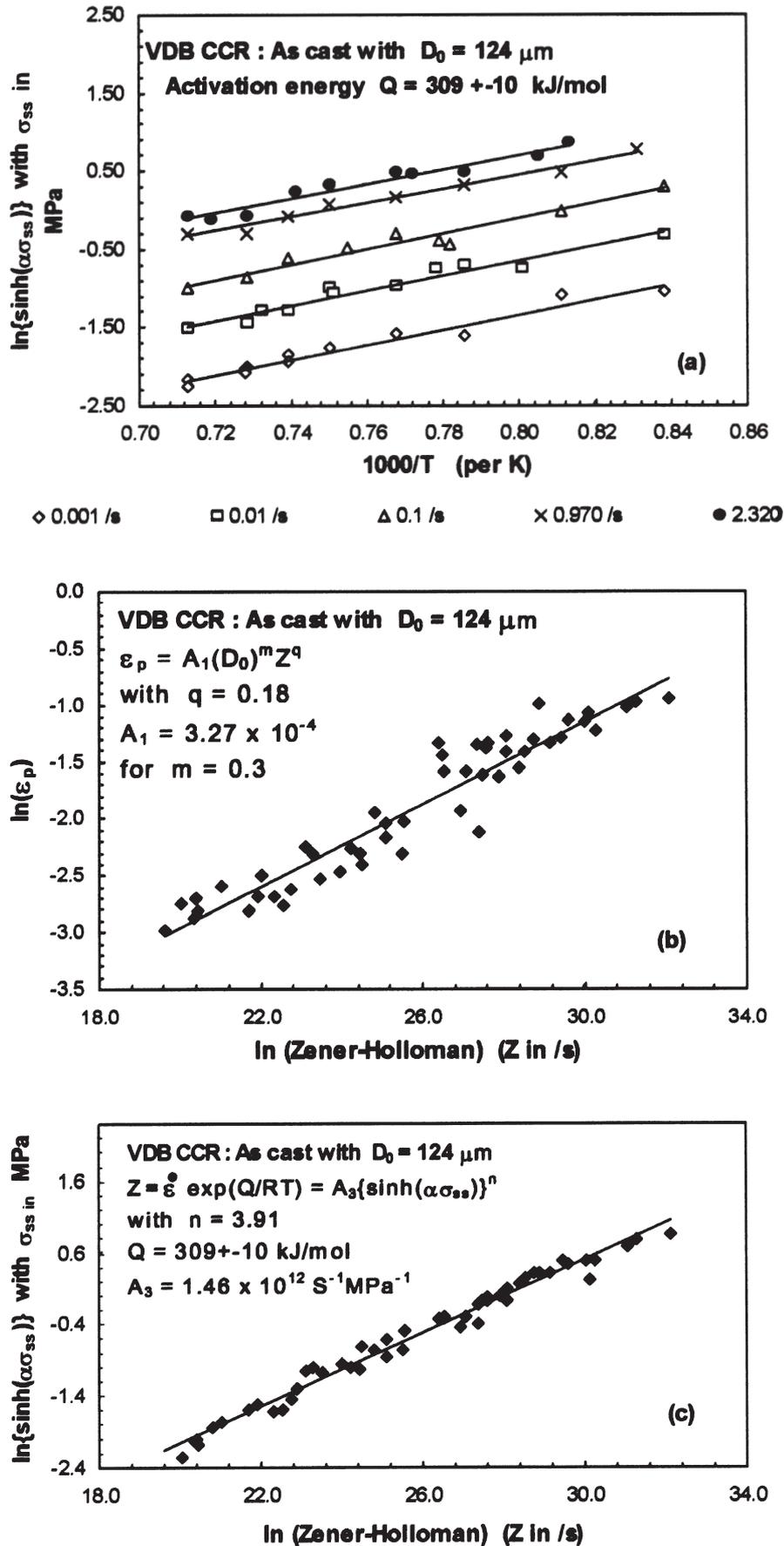


Figure 5—Curves for the hot deformation of the CCR steel in the coarse grained as-cast condition with (a) used to find the activation energy, (b) used to find the values for  $q$  and  $A_1$  and (c) the overall hot deformation curve combining all of the data. Line fitting: regression coefficients  $R^2$  were larger than 0.95 in all cases

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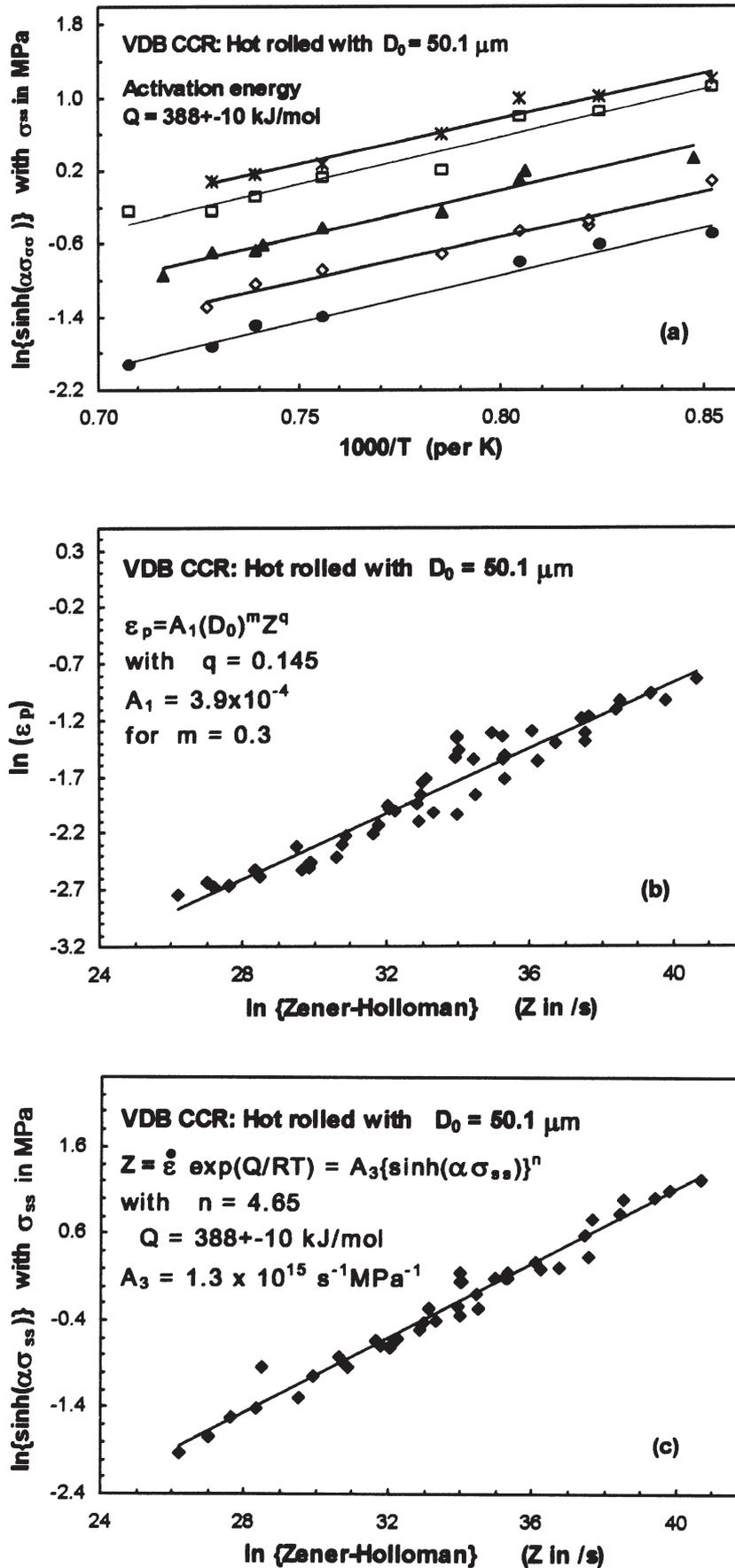


Figure 6—Curves for the hot deformation of the CCR steel in the laboratory hot rolled fine grained condition with (a) used to find the activation energy, (b) used to find the values for  $q$  and  $A_1$  and (c) the overall hot deformation curve combining all of the data. Line fitting: regression coefficients  $R^2$  were larger than 0.95 in all cases

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Table III  
Hot work constitutive constants for the DRX regime for the CCR and HCR low carbon strip steels

Steel	Starting condition and applicable temperature range	$\dot{\epsilon} \exp(Q/RT) = Z = A_3 \{\sinh(\alpha\epsilon_{ss})\}^n$	$\epsilon_p = 0.8 \epsilon_c = A_1(D_0)^m Z^q$
CCR	As-cast with $D_0 = 124 \mu\text{m}$ and $T = 1140$ to $900^\circ\text{C}$	$Q = 309 \pm 10 \text{ kJ.mol}^{-1}$ $n = 3.91$ $A_3 = 1.46 \times 10^{12} \text{ s}^{-1} \text{ MPa}^{-1}$	$q = 0.18$ $A_1 = 3.27 \times 10^{-4}$ for $D_0 = 124 \mu\text{m}$
	Hot rolled with $D_0 = 50 \mu\text{m}$ and $T = 1140$ to $900^\circ\text{C}$	$Q = 388 \pm 10 \text{ kJ.mol}^{-1}$ $n = 4.65$ $A_3 = 1.3 \times 10^{15} \text{ s}^{-1} \text{ MPa}^{-1}$	$q = 0.145$ $A_1 = 3.9 \times 10^{-4}$ for $D_0 = 50 \mu\text{m}$
HCR <sup>1</sup>	As-cast with $D_0 = 277 \mu\text{m}$ and $T = 1140$ to $1000^\circ\text{C}$	$Q = 238 \text{ kJ.mol}^{-1}$ $n = 5.3$ $A_3 = 3.8 \times 10^8 \text{ s}^{-1} \text{ MPa}^{-1}$	$q = 0.22$ $A_1 = 6.2 \times 10^{-4}$ for $D_0 = 277 \mu\text{m}$
	Hot rolled with $D_0 = 26 \mu\text{m}$ and $T = 1000$ to $900^\circ\text{C}$	$Q = 333 \pm 20 \text{ kJ.mol}^{-1}$ $n = 5.1$ $A_3 = 1.25 \times 10^{13} \text{ s}^{-1} \text{ MPa}^{-1}$	$q = 0.20$ $A_1 = 1.3 \times 10^{-4}$ for $D_0 = 26 \mu\text{m}$
Typical values from the literature for low carbon steels	CCR processed strip steels <sup>1</sup>	$Q = 202$ to $314 \text{ kJ.mol}^{-1}$ with 312 accepted widely $n = 4.34$ to $5.3$ $A_3 = 9.69 \times 10^{11} \text{ s}^{-1} \text{ MPa}^{-1}$ $\alpha = 0.0095$ to $0.0143 \text{ MPa}^{-1}$ with $0.0143$ accepted widely	$q = 0.17$ to $0.23$ $m = 0.1$ to $0.5$ with $0.3$ accepted widely $A_1 = 4.6 \times 10^{-4}$ to $1.3 \times 10^{-2}$
	HCR processed strip steel <sup>2</sup>	$Q = 314 \text{ kJ.mol}^{-1}$ $n = 4.34$ $A_3 = 9.69 \times 10^{11} \text{ s}^{-1} \text{ MPa}^{-1}$ with $\alpha = 0.013 \cdot f(T, \epsilon)$	$Z_1 = \dot{\epsilon} \exp(Q_1/RT)$ with $Q_1 = 147.3 \text{ kJ.mol}^{-1}$ $q = 0.165$ and $\epsilon_c = 0.83 \epsilon_p$ $\dot{\epsilon}_{ss} = f\{D_0, \exp(Q_2/RT)\}$ with $Q_2 = 355 \text{ kJ.mol}^{-1}$ for $D_0 < 244 \mu\text{m}$ and $Q_2 = 21.9 \text{ kJ.mol}^{-1}$ for $D_0 \geq 244 \mu\text{m}$ .

Note: In the calculations of the Ispat Iscor CCR and HCR steels  $\alpha = 0.0143 \text{ MPa}^{-1}$  and  $m = 0.3$  were used throughout.

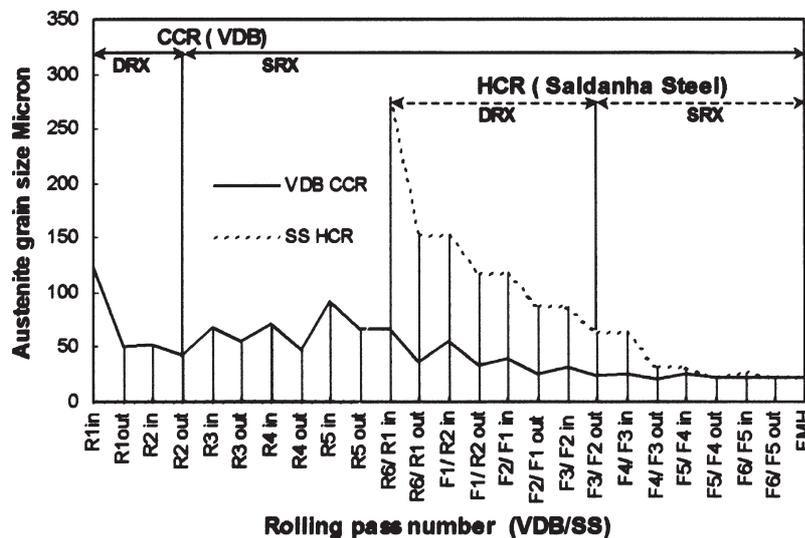


Figure 7—Comparison of the modelled austenite grain size during hot rolling for the CCR and the HCR processed steels in the respective VDB and Saldanha Steel strip mills. FMH = finishing mill head

The assumption that the SRX and grain growth constants for the two steels are identical and were, therefore, taken from the literature<sup>4</sup>, requires verification.

Note the differences in starting austenite grain size for the two cases, with the CCR steel predicted to reach relatively small austenite grain sizes very soon within the first and second roughing passes. Considerable grain growth is

predicted to take place during the interpass times in this steel, particularly after the even numbered roughing passes where the slab/plate requires reversing in the reversing mill. The interpass times for these particular passes were taken as the worst case for the intermediate product, i.e. the leading edge that becomes the trailing edge upon reversing experiences the longest interpass time. Secondly, the already

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smaller austenite grain sizes at relatively high temperatures in the CCR steel introduce greater driving forces and hence greater rates for grain growth. Both of these contributing factors are, respectively, either fully or largely absent in the HCR process at Saldanha Steel, with a smaller austenite grain size only appearing when the temperature is already relatively low; hence very little grain growth is predicted during the hot rolling of the HCR steel if compared to the CCR steel at VDB.

### Discussion

Using the assumption that both steels and their respective hot rolling processes possess identical constants for SRX and grain growth (an assumption that may be questioned), the modelling does not predict a significant difference in the final austenite grain size at the finishing mill head between the two steels in their respective strip hot rolling plants. If this assumption proves to be correct, the observed plant differences in actual ferrite grain size after cooling on the run-out table would be solely due to the austenite to ferrite transformation behaviour. In practice it has been found that the HCR steel generally has a smaller ferrite grain size and a somewhat higher yield and tensile strength than the equivalent CCR steel. Should this assumption on the constants for SRX, however, be found to be invalid, a revisit of the modelling studies with experimentally determined constants for SRX and grain growth for the two steels, may solve some of the questions as to why the nominally equivalent two steels and their respective processes behave metallurgically so different in many respects after hot rolling.

Calculations from the model of the predicted retained strain  $\epsilon_{\text{ret}}$  in the austenite due to incomplete recrystallization after the final finishing pass, showed a higher retained strain in the process/product combination of Saldanha Steel than may be the case for VDB, typically about 0.25 for the former against about 0.17 for the CCR steel. This may partly cause the observed smaller ferrite grain size in the HCR steel compared to the CCR steel although, due to the above uncertainty of the constants used in the SRX part of the model, this conclusion needs to be verified.

In this comparison of the hot working characteristics, a number of similarities but also some significant differences have been encountered between the two nominally equivalent steels. Both steels are similar in having a distinct difference in hot working constitutive constants between a coarse grained as-cast structure and a finer grained hot rolled structure, with the activation energy for austenitic hot deformation generally lower for the as-cast structure than for the hot rolled structure. This is in spite of a somewhat large difference in the as-cast starting austenite grain size  $D_0$  of 124  $\mu\text{m}$  for the CCR versus 277  $\mu\text{m}$  for the HCR steel. This is, therefore, a principle phenomenon in these two steels which appears to be unaffected by any differences in chemistry and processing and appears to be related to the fundamental differences in austenite grain structure between the two starting conditions. Unfortunately no published results exist on hot working studies of low carbon strip steels with an as-cast starting grain structure and how this compares to a finer grained structure. It is, therefore, uncertain whether this could also in other steels be a universal phenomenon between the two types of starting structure.

Significant differences exist, however, in the absolute values of the two sets of activation energies and this has a decided effect on the softening behaviour during hot rolling in the two plants. The higher  $Q$  value of 388  $\text{kJ}\cdot\text{mol}^{-1}$  of the fine grained CCR steel brings about that relatively high  $Z$  values are attained very early in the hot rolling process at VDB, even where the strain rate is still typically as low as 5  $\text{s}^{-1}$  at the roughing pass R3. This causes the actual strain per pass to fall below the critical value  $\epsilon_c$  required for DRX, and hence SRX softening is predicted to be the main softening mechanism almost throughout the entire hot rolling schedule. With the lower  $Q$  value of 333  $\text{kJ}\cdot\text{mol}^{-1}$  for the HCR steel, higher  $Z$  values are only attained much later in the entire rolling process, typically at the finishing pass F3 from which onwards SRX becomes the softening process. As indicated in Table III, Muojekwu *et al.*<sup>2</sup> also found differences in the activation energies for the hot deformation of an HCR processed strip steel between small (<244  $\mu\text{m}$ ) and large (>244  $\mu\text{m}$ ) grain sizes. They introduced these activation energies into an equation for the critical steady state strain rate  $\dot{\epsilon}_{\text{ss}}$  for DRX (and not the usual critical strain as in Equation [2]), which was found to be a function of both the temperature and the austenite grain size. Because of this unusual approach and also because of differences in the composition of their steel to these studied here, a direct comparison of activation energies between this work and theirs is not meaningful.

Fundamentally, a difference in activation energies between any two thermally activated processes usually signals a difference in rate controlling mechanism. Earlier<sup>1</sup> it was pointed out that the formation of a recrystallizing nucleus on a pinned grain boundary by the Bailey-Hirsch method<sup>11,12</sup> is critically dependent on both the grain size and the pinning frequency of the grain boundaries. It is, therefore, understandable that the starting austenite grain size  $D_0$  appears in the expression for the critical strain to initiate DRX in Equation [2]. Secondly, the differences in austenite grain structures in Figures 3 and 4 also indicate that the wavy appearance of the austenite grain boundaries in the CCR steel may be due to a higher pinning frequency of its austenite grain boundaries than is the case of the HCR steel that has generally straight austenite grain boundaries. This will certainly introduce differences in softening kinetics but also possibly in the softening mechanism between the two steels, leading to the observed differences in the activation energies for hot working.

In general, activation energies obtained through hot deformation tests are defined as 'apparent activation energies' and authors do not attempt to assign any specific deformation and softening mechanism to the activation energy. This prudent approach is also reflected in most hot deformation studies on creep resistant steels where it has been found that the activation energy for hot deformation is up to 50 per cent greater than that for self-diffusion or low strain rate creep<sup>13</sup>. This general discrepancy has defied any satisfactory explanations based on existing models for deformation at elevated temperatures<sup>14</sup> but is generally ascribed to the existence of other (undefined) temperature sensitive parameters that have not been captured by the existing hot deformation models.

Finally, a few words need to be said on the possible effects of the differences in chemical composition between

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the CCR and HCR steels in Table II. The CCR steel contains measurably greater quantities of the residual elements of chromium, nickel, copper and nitrogen and, of course, sulphur. Except for the higher sulphur content in this steel (which is due to no calcium treatment at VDB) and also the higher nitrogen content from the EAF melting (which is not usual for CCR steels from VDB that are mostly processed via the BOF route), the other three residuals may reflect the larger scrap recycling quantities exercised at VDB than at the relatively new (at the time of cast 7037 for this study) Saldanha Steel plant. In the austenitic condition these three residuals should all be in solution and, at these levels, are not expected to significantly affect the hot working characteristics of this steel. Even at room temperature after cooling, they do not appear to reflect in a strengthening as the VDB strip steels were invariably found to have lower yield and tensile strengths than the equivalent strip steels of Saldanha Steel. Could the higher activation energy for hot working of the CCR steel then possibly be ascribed to the unusually high nitrogen content in this slab? This is not thought to be so as the calculated MFS values in Figure 1 from a large number of mill logs of all BOF-melted steels with lower nitrogen contents, also show a strong indication of a higher temperature sensitivity of the MFS than was found in the HCR steel. This really leaves only the sulphur content difference between the two types of steel that may have some undefined effect on the hot working characteristics. This needs to be considered in any future work.

## Conclusions

The following may be concluded from this study:

- ▶ Measurable differences have been observed in the ferrite to austenite transformation behaviour between the nominally equivalent HCR processed steel of Saldanha Steel and the CCR processed low carbon strip steel of VDB. In general, the CCR steel produced an irregular austenite grain structure with a smaller austenite grain size starting from an as-cast structure and a larger austenite grain size starting from a hot rolled structure if compared to the HCR steel with its highly regular austenite grain structure.
- ▶ The CCR processed low carbon strip steel shows a significantly different set of hot working constitutive constants between testing it in the as-cast coarse-grained microstructure as opposed to a hot rolled fine-grained structure. This is similar to what was found in the HCR processed steel of nominally the same composition, although the absolute values of the constitutive constants differed significantly between the two steels
- ▶ The higher activation energy of 388 kJ.mol<sup>-1</sup> for the fine-grained CCR steel as opposed to the lower value of 333 kJ.mol<sup>-1</sup> for the equivalent HCR steel, results in significant differences in the expected softening mechanisms during hot rolling between the two plants. At VDB with the CCR steel, DRX takes place only during the first two roughing passes with SRX taking place for the last ten to eleven passes from the roughing pass R3 onwards. At Saldanha Steel with the HCR steel, DRX takes place during the two roughing

and the first two finishing passes and SRX is only expected to occur for the last three passes from F3 onwards

- ▶ The current model for austenite grain size development uses the respective DRX experimentally determined constants for each of the two steels but also assumed published and identical values for the SRX constants. Based on this assumption, differences in the austenite grain size development during the earlier stages of both hot rolling operations are predicted to be smoothed out towards the end, with no differences predicted in the final austenite grain size after exiting from the finishing line. In view of the significant differences in hot working constants for DRX already found between the two steels, this assumption needs to be examined before a trustworthy grain size model can be presented that covers both the DRX and the SRX portions of hot rolling for these two steels.

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