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EFFECT OF THE HOT ROLLING PROCESS ON THE MICROSTRUCTURE AND MECHANICAL PROPERTIES OF DUAL-PHASE STEELS

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ABSTRACT

In this paper the effects of the finishing temperature and cooling rate on the development of microstructure and mechanical properties of an as-hot-rolled Mn-Si-Cr-Mo dual-phase steel were described. It was verifyed that for samples directly cooled from the rolling heat the cooling rate was the most important process parameter for the determination of steel properties. However, for the samples submitted to a cooling pattern that includes an isothermal hold period between 550 and 650°C, the temperature of the quench interruption was the most influencing parameter for the development of the dual-phase microstructure and its characteristic mechanical properties.

- INTRODUCTION

The HSLA steels developed since the seventies allowed a great reduction in the thickness and weight of the auto-parts originally made with traditional mild steels owing to its increased mechanical strength. However, these new steels do not show the same degree of cold formability that is characteristic to mild steels. So, its application to parts with complex shapes

is limited.

The conciliation between high strength and good cold formability was obtained through the development of steels with a microstructure composed of a polygonal ferritic matrix with 15 to 20% of martensite evenly distributed. This microstructure originates a peculiar mechanical behavior[1]: continuous yielding, yield strength at 0,2% between 300 and 380 MPa, a high strain hardening coefficient $\underline{n}$ - 0,2 to 0,3 -, tensile strength between 620 and 655 MPa, a low yield ratio - 0,5 to 0,6 - and total elongation equal to or higher than 27%. This kind of steel was denominated as dual-phase owing to its characteristic microstructure.

Dual-phase steels can be produced directly from the rolling heat or through intercritical annealing. The as-hot-rolled dual-phase steels were first developed in 1976 and some heats were made under industrial scale[2]. This first approach had as its main goal to develop an alloy design in order to obtain a strip with dual-phase microstructure without significant changes in the normal process parameters of hot strip rolling. Besides, the austenite decomposition must have some insensibility to the process oscilations that are unavoidable under industrial conditions. After some trials it was developed an otimized alloy that satisfyed all conditions described above: 0.065% C, 1.20% Mn, 0.90% Si, 0.38% Mo and 0.61% Cr.

The characteristic transformation behavior of this alloy results from two fundamental mechanisms. The first of them is a balance between the acceleration of ferritic transformation

promoted by the low C content, high Si content and the hot deformation, and the delay of the pearlitic transformation owing to the presence of Mo, Mn and Cr. The other is the stabilization of the C enriched remaining austenite promoted by Mo, Si, Cr and Mo, in order to avoid the immediate transformation of austenite to bainite in the range of coiling temperatures (510-620°C), allowing the formation of martensite during the subsequent very slow cooling of the coil[2,3].

During the processing of this steel in the Hot Strip Mill the finishing temperature must be immediately above the Ar_3 temperature, in order to maximize the strain hardening of the austenite that is ready to transform. However, the alloy can not be rolled below this temperature, because the as-formed ferrite will be strain hardened and so its mechanical strength and yield ratio would be increased; this fact can jeopardize the cold formability of the product[4,5]. So, the finishing temperature must be criteriously chosen, as the Ar_3 temperature of the stock being rolled is a function of the chemical composition of the alloy as well of its thermomechanical "history" and cooling rate[4].

The deformation plays a fundamental role on the development of the dual-phase microstructure. The ferritic reaction of a dual-phase Mn-Si-Cr-Mo steel shows an incubation time of 200 s at 700°C. The industrial processing of dual-phase steels at the Hot Strip Mill would not be possible without the influence of deformation on the austenite transformation, as the cooling time of the strip between the finishing stand and the

coiler is approximately 6 to 15 s. However, both the results obtained at laboratory and at plant showed that the deformation accelerated the beginning of the ferrite and pearlite transformation from 30 to 100 times, and concentrated the transformation lines in the superior section of the CCT diagram. This corresponds to the formation of 80% ferrite during the time which the strip took to pass through the cooling table[2,4,6-8].

According to several authors[1,5,9,10] the principal process parameter of the hot strip rolling of dual-phase steels is the coiling temperature. This parameter must be kept below 600°C in order to avoid the formation of pearlite, this assuring the continuous yielding of the product.

The objective of this work was to study the effect of the finishing temperature and cooling rate on the development of microstructure and mechanical properties of an as-hot-rolled Mn-Si-Cr-Mo dual-phase steel, in order to determine otimized process parameters for the hot strip rolling of this kind of product.

- EXPERIMENTAL PROCEDURE

It was selected for this study an alloy which promotes the largest "coiling window" in the CCT diagram, i.e., the largest range of coiling temperatures that can be applied at the Hot Strip Mill for the production of dual-phase steels. Its

chemical composition was 0.063% C, 0.87% Mn, 1.46% Si, 0.41% Cr and 0.38% Mo. The alloy was melted in a vacuum induction furnace, and a 100-kg ingot was obtained. This ingot was subsequently forged and rolled in order to homogeinize its as-cast structure. The specimens for the rolling tests were machined from the rolled bars.

The samples were heated to 1200°C during 45 minutes and rolled according to a five pass schedule from 25 to 5 mm. The first two passes constituted the roughing stage and the remainder the finishing stage, with a holding period between the two stages. The duration of this period was a function of the selected finishing temperature: 950, 900 or 850°C. After rolling the specimens were quenched in one of the several cooling media available: water, oil, aqueous solution of 0.55% polyacrilamide, air or diatomite. In another series of tests the specimens were cooled in the aqueous solution of polyacrilamide down to a "coiling" temperature of 650 or 550°C; the quench was interrupted and the specimen was introducted in a furnace previously heated to this temperature. The specimen was hold there during one hour; then the furnace was powered off and the specimen cooled inside it down to room temperature. This special cooling cycle roughly simulates the cooling pattern applied in the Hot Strip Mill.

The temperature evolution during the rolling tests were monitored through a chromel-alumel thermocouple sheated by a 3,0 mm diameter inoxydable steel tube inserted in the specimen being rolled. The signal generated by the thermocouple was registered in a graphical recorder.

The samples thus obtained were submitted to quantitative optical metallographic analysis. It was determined the volumetric fraction of the non-ferritic constituents, grain size and Vickers hardness of the constituents present in the microstructure[11,12]. It was used Le Pera[13] or Picral 5% etch for the revelation of the non-ferritic constituents and Nital 5% for the determination of grain size and constituent identification for the Vickers hardness tests.

In addition tensile tests were made, using subsize specimens machined from the rolled samples, according to the ASTM A-370 Standard. From these tests the yield and tensile strength, yield ratio, uniform and total elongation were determined.

- RESULTS AND DISCUSSION

All the samples showed a prevailing ferritic microstructure, except the specimens quenched in water, which showed bainite as the prevalent constituent. The ferritic samples had a mixture of bainite and martensite as the second constituent dispersed as islands on the matrix. However, in the samples "coiled" at 650°C the second constituent was pearlite. In this case the remaining austenite continued to transform in ferrite, and, subsequently, to pearlite, during the holding period in the furnace after the quench. This indicates that the metastable

austenite bay in the CCT diagram of this steel is below 650°C. The figure 1 shows the microstructures of the rolled samples with finishing temperature of 950°C and cooled in (a) water and (b) aqueous solution of 0.55% polyacrilamide with interruption at 550°C and subsequent furnace cooling.

Figure 2a shows that the bainite-martensite constituent volumetric fraction decreased with a declining cooling rate, as related previously[4,6]. The finishing temperature only affected the bainite-martensite fraction of water quenched samples, which decreased with the reduction of this temperature. This can be justified by the longer time available for ferrite formation. As for the samples submitted to interrupted quenching, it was verified that a "coiling" temperature of 650°C led to lower values of volumetric fraction of second constituent than the "coiling" at 550°C, as can be seen in figure 2b. This can be explained by the continuation of the austenite decomposition during the holding period in the furnaces for the samples "coiled" at 650°C. In this case it was not observed any influence of the finishing temperature as well.

Both the increase in cooling rate as the decrease in finishing temperature led to a more refined ferritic grain size, as the figure 3a shows. The decrease in finishing temperature promotes a smaller prior austenitic grain size, and the increase in the cooling rate caused a greater elevation in the nucleation rate of ferrite in relation to its growth[14]. These two factors promoted a refine of the ferritic grain size. The samples submitted to interrupted quenching showed similar ferritic grain

size, irrespective of their coiling temperatures, as the fig. 3b shows. However, it can be perceived a subtle effect of the finishing temperature: its reduction promoted a small degree of refining in the ferritic grain size.

The evolution of the bainite-martensite constituent grain size showed a peculiar trend, irrespective of the finishing temperature, as can be seen from figure 4a: it was obtained a minimum value for this parameter under intermediate cooling rates, i.e., when the quench was performed in the aqueous solution of 0.55% polyacrilamide. It was observed no effect of the finishing temperature on this parameter.

This fact can be explained when one notes that, as the cooling rate decreased, the ferritic grain size increased and the volumetric fraction of second constituent decreased only very slightly for quenching medium other than water. As the second constituent grain size is proportional to its volumetric fraction and the ferritic grain size, the reduction of the second constituent volumetric fraction will cancel the increase on its grain size promoted by the greater ferritic grain size down to intermediate cooling rates. For slower cooling rates the influence of the increased ferritic grain size is greater, and the second constituent grain size will increase. This was also observed by other author[4].

In the case of samples submitted to interrupted quenching the second constituent grain size decreased discretly as the finishing temperature lowered, as can be seen from the figure 4b. This grain size is minimum for the "coiling"

temperature of 650°C; this fact can be attributed to the continuation of the austenite transformation during the holding period inside the furnace after the quench.

The ferrite hardness evolution of the samples directly quenched showed a decline with decreased cooling rates, as figure 5a shows. The finishing temperature affected only the ferrite hardness of the samples quenched in water: in this case, the lowering of this temperature shows a trend to reduce the hardness values. This fact can be attributed to a increase in the ferrite dislocation density as its formation temperature is decreased[15], or the second constituent fraction increased[16]. Is was further suggested a "support effect" caused by the hard second constituent that could affect the Vickers Hardness tests[17]. As for the samples submitted to interrupted quenching, both the finishing as the "coiling" temperature exerted no significant effect on the ferrite hardness - figure 5b.

One can notice that the bainite/martensite constituent hardness showed a maximum in the samples cooled in air, irrespective of the finishing temperature applied, as shown in figure 6a.. This can be attributed to a balance between two conflictant factors: as the time available for the carbon diffusion increases - i.e., the cooling rate decreases - the diffusivity of this element declined due to a smaller grain boundary area resultant from the increased ferritic grain size[4]. In addition, it must be considered the effect of self-tempering that could occur in the samples slowly cooled[11].

The finishing temperature did not affect the bainite-martensite constituent hardness, except for the case of the water quenched samples, which showed an increase in this hardness as the finishing temperature was lowered. This can be due to the formation of a greater fraction of polygonal ferrite, which promotes an enrichment of C in the remaining austenite that will subsequently transform to the bainite-martensite constituent[4].

The samples submitted to interrupted quenching and "coiled" at 650 or 550°C presented pearlite or bainite-martensite as the second constituent, respectively. This affected decisively the hardness of the second constituent, as can be seen at figure 6b. As expected, pearlite was softer than bainite. On the other hand, it was not verifyed any significant influence of the finishing temperature on the hardness of the second constituent.

The yield strength of the directly quenched samples fall with decreasing cooling rates and increasing finish temperatures, as shown in figure 7a. This can be justifyed by the greater fraction of softer ferrite formed under these conditions as well by its greater grain size[4,6]. The excepcionally high value for the water quenched samples can be explained by the high volumetric fraction of bainite present on its microstructure.

The yield strength of the samples submitted to interrupted quenching were affected only by the "coiling" temperature; the effect of the finishing temperature was not significant, as can be seen in figure 7b. As a matter of fact, the yielding of the samples "coiled" at 650°C was discontinuous, while the samples "coiled" at 550°C showed continuous yielding.

The type of yielding was determined by the nature of the second constituent present in microstructure, that is, pearlite in the first case and bainite/martensite in the other. The formation of bainite/martensite induced to a higher level of residual stress and free mobile dislocations in the ferritic matrix than pearlite. These stresses and free dislocations ease the yielding processes of the material, making it continuous[19].

All the samples showed continuous yielding, except those cooled in diatomite from the finishing temperatures of 900 and 850°C and the samples submitted to interrupted quenching with "coiling" temperature of 650°C. In this latter case the elongation during yielding was greater. The reason for this behavior can be the partial or total formation of pearlite as the second constituent of the microstructure.

The yield strength can be quantitatively described from the microstructural parameters by the following equations:

$$Y.S.[MPa] = -41 + 1128\ d_{\alpha}^{-1/2}\ [\mu] \qquad r^2 = 0,65 \qquad (1)$$

$$Y.S.[MPa] = 203 + 855\ L_{\alpha\alpha}^{-1/2}\ [\mu] \qquad r^2 = 0,88 \qquad (2)$$

where d_{α} is the mean ferritic grain size, and $L_{\alpha\alpha}$ is the mean ferritic free path, or the mean spacing between the bainite-martensite islands. The figures 8a and 8b show the plot of data and the regression equations.

Equation (1) is the traditional Hall-Petch relation. However, it did not generate a good fit, as can be seen by the

low r^2 value; besides, the constant term in the equation is negative: this evidenced that the equation is physically incorrect. However, the substitution of the ferritic grain size by the ferritic free path in the same equation led to a new relation with a better correlation index and that is physically correct. This suggests that phase boundaries are more effective obstacles to dislocation migration than the ferritic grain boundaries in the specific case of the dual-phase steels[20].

The tensile strength also decreased with declining cooling rates for the samples directly quenched, but in a more sudden way, as can be seen at figure 9a. This can be attributed to the greater volumetric fraction and grain size of ferrite, as well its lower hardness. However, the finishing temperature did not affect the yield strength, except for the water quenched samples, because of the greater alteration in the microstructure caused by the change of this temperature. The finishing temperature also did not affect the tensile strength of the samples submitted to the interrupted quench, as the figure 9b shows. In this latter case the "coiling" temperature had a remarkable influence: the tensile strength of the samples "coiled" at 550°C was greater than the samples "coiled" at 650°C; an inverse situation was observed for yield strength. This can be due to the higher strain hardening coefficient n of the microstructures with bainite-martensite as second constituent.

The tensile strength can be expressed by the following equation:

$$T.S.[MPa] = 266 + 548\, L_{\alpha\alpha}^{-1/2} + 1741 \sqrt{f_\beta / d_\beta} \qquad r^2=0,87 \qquad (3)$$

where f_β is the volumetric fraction of the bainite-martensite constituent and d_β its grain size.

It can be seen from equation (3) that there are two contributions of the microstructure for the description of the tensile strength. One term is represented by the Hall-Petch relation using the mean ferritic free path instead of grain size, as in equation (2): this indicates again the effect of the phase boundaries as a more effective obstacle to the dislocation migration. The other is a term that ressembles the description of the strain hardening behavior according to Ashby theory, i.e., the re-arranging of dislocations in the ferritic matrix by the islands of hard second constituent during the deformation[21].

The yield ratio of the directly quenched samples reached a minimum value for intermediate rates of cooling, except for the samples submitted to the finishing temperature of 850°C, when the lowest yield ratio was obtained for the water quenched sample, as can be seen at figure 10a.

The effect of the finishing temperature on yield strength and yield ratio was similar: for the water quenched samples the yield ratio was reduced as the finising temperature decreased, while for the other samples it was observed a contrary effect.

As for the samples submitted to interrupted quenching, it was verifyed that only the "coiling" temperature affected the yield ratio, as the fig. 10b shows. This fact could

be expected from the observed effects of the finishing and "coiling" temperatures on the yield and tensile strengths, respectively showed in the figures 7b and 9b.

A decrease in the cooling rate promoted an increase both in uniform and total elongation, as can be seen at fig. 11a. This can be attributed to the greater volumetric fraction and grain size of ferrite, as well its lower hardness. The finishing temperature did not exert any evident effect, except for the water quenched samples.

In the case of the samples submitted to interrupted quenching it was observed that the samples "coiled" at 550°C showed a slightly greater uniform elongation than those "coiled" at 650°C. The situation was inversed for the total elongation - figure 11b. The decrease in the finishing temperature led to a small reduction in the elongation values.

Finally, one can observe that only the samples directly cooled in air and the samples submitted to interrupted quenching with "coiling" temperature of 550°C fulfil all the microstructural and mechanical requirements to be considered as "standard" dual-phase steels. This suggests the necessity to use high finishing temperatures - 950 to 900°C - and coiling temperatures below 550°C in order to guarantee the supression of pearlite in microstructure, thus promoting continuous yielding, low yield ratio and high strain hardening coefficient, which provides the good cold formability and high mechanical strength to the dual-phase steels.

- CONCLUSIONS

- The cooling rate was the most important process parameter for the determination of microstructure and mechanical properties of the as-hot-rolled dual-phase steel. The air cooled samples showed microstructures and mechanical properties typical of the "standard" dual-phase steels;

- The finishing temperature had a less remarkable effect, except for the water quenched samples. Higher finishing temperatures - 950 to 900°C - led to lower yield strength and increased elongation values;

- The "coiling" temperature was the most significant parameter that affected the development of the second constituent in the microstructure and the mechanical properties of the samples submitted to interrupted cooling. It must be kept at or below 550°C in order to warrant the microstructure and mechanical properties typical of the "standard" dual-phase steels;

- The yield strength can be quantitatively described by a Hall-Petch-type equation using the ferritic mean free path instead of its mean grain size. The tensile strength requires an additional term to include the effect of the hard second constituent islands of the microstructure on the strain hardening, as determined by the Ashby theory.

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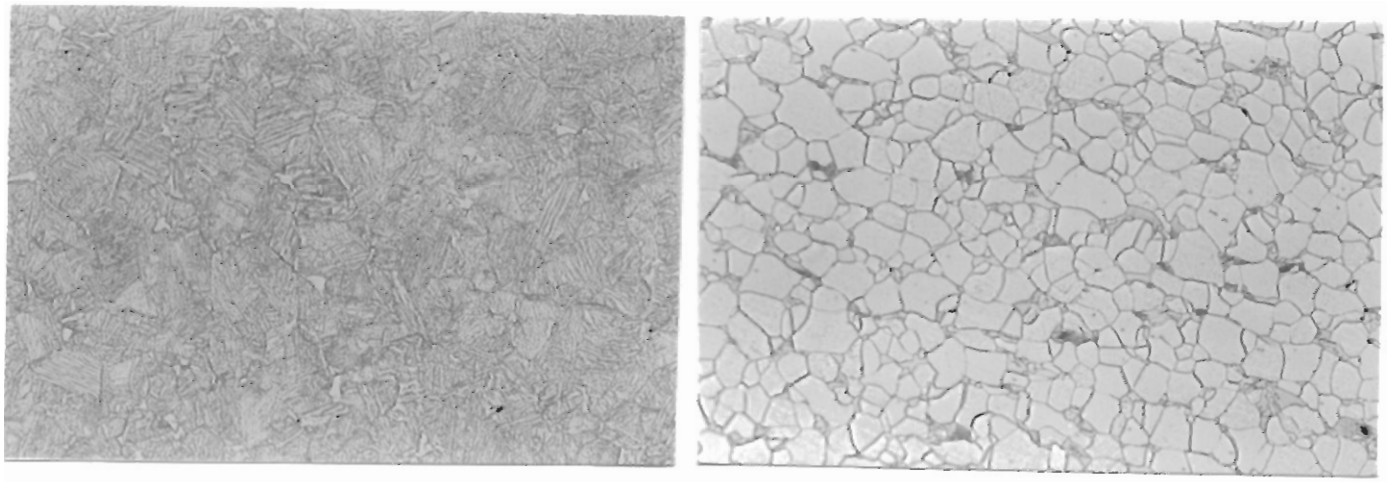
- FIGURES



(a) (b)

Fig. 1: Microestruture of the samples finished at 950°C and cooled in (a) water or b) aqueous solution of 0.55% polyacrilamide interrupted at 550°C, held one hour and furnace cooled. Nital 5% etch, 400 x.

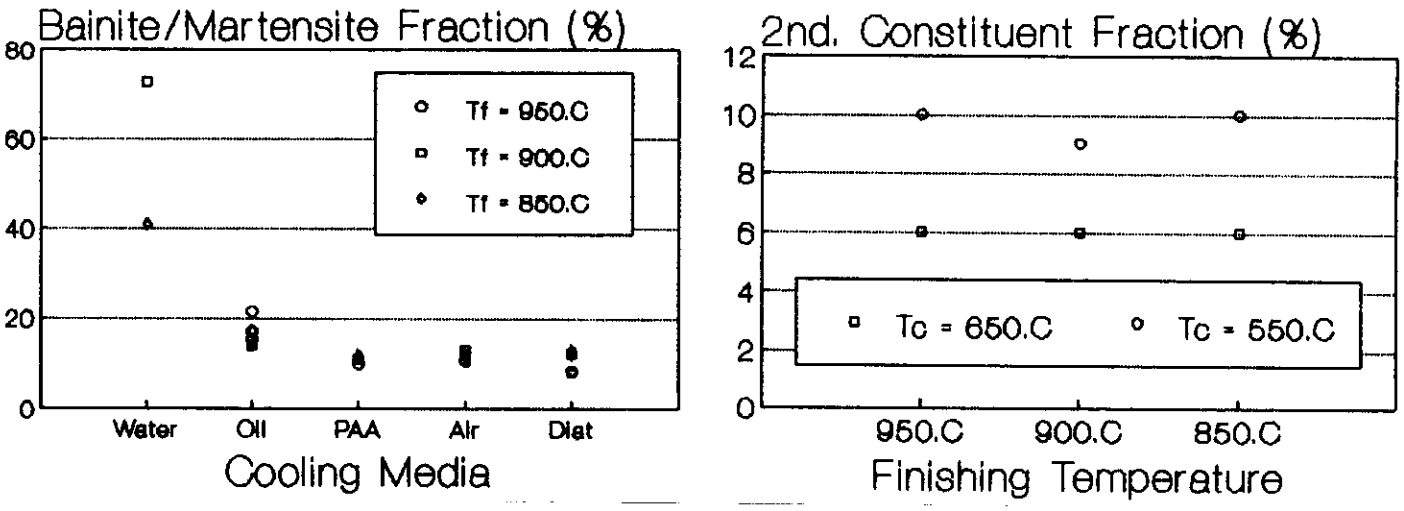


Fig. 2: Volumetric fraction of second constituent according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.

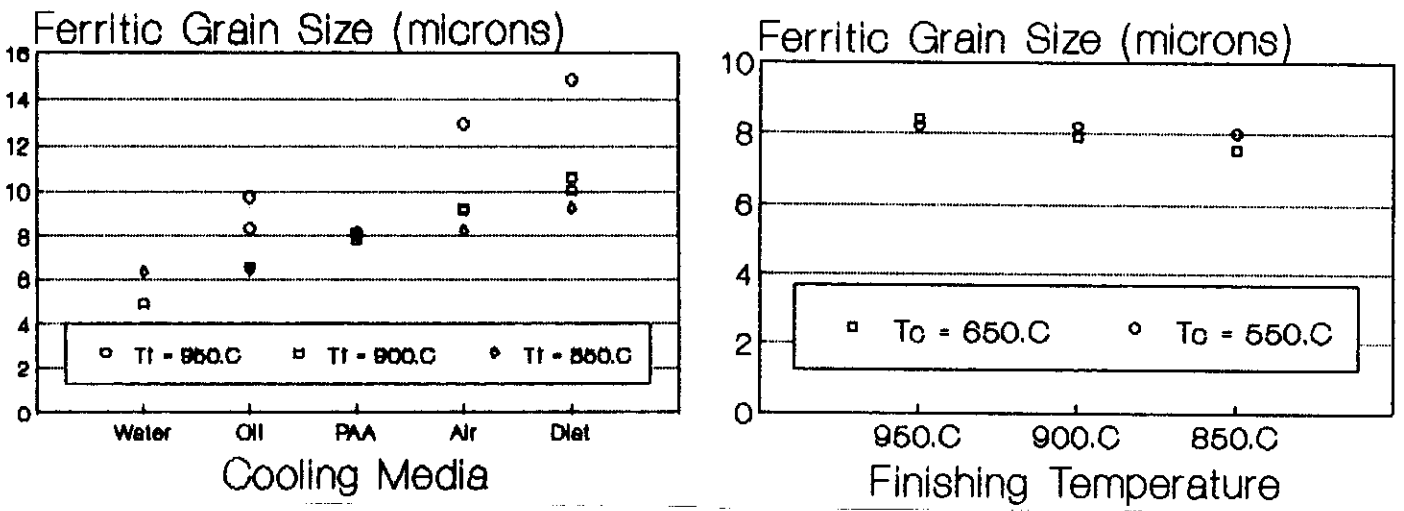
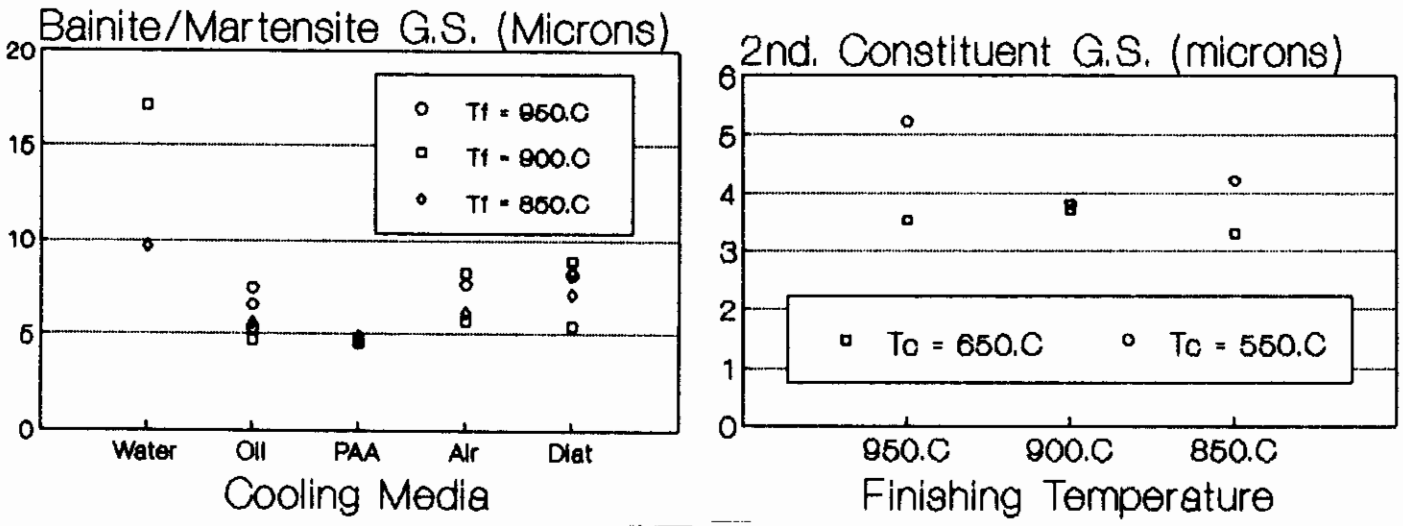
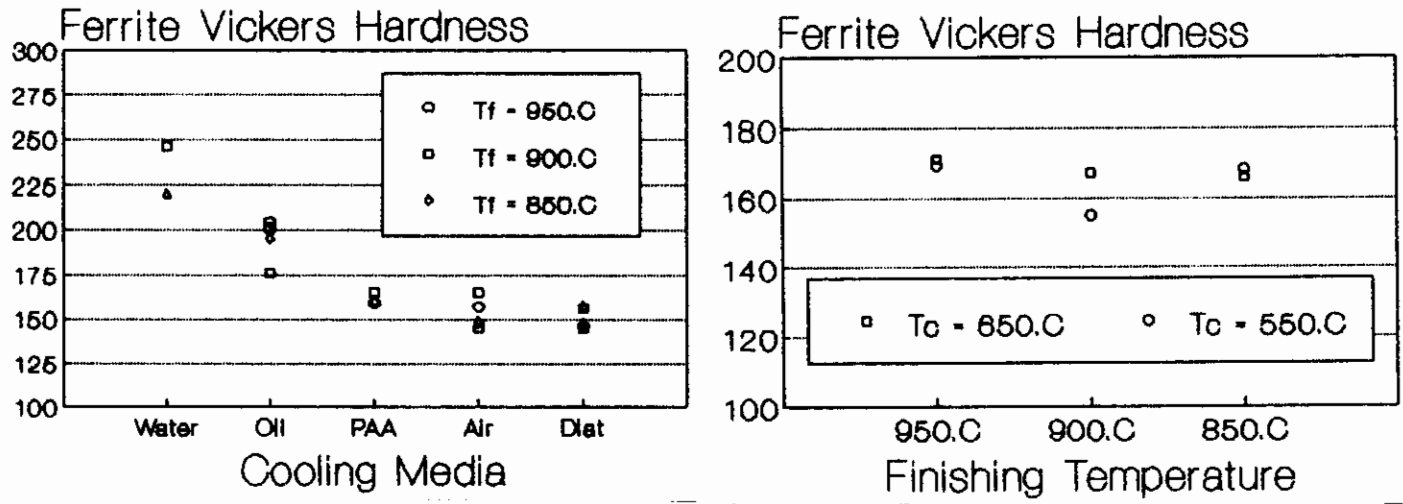


Fig. 3: Ferritic Grain Size according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



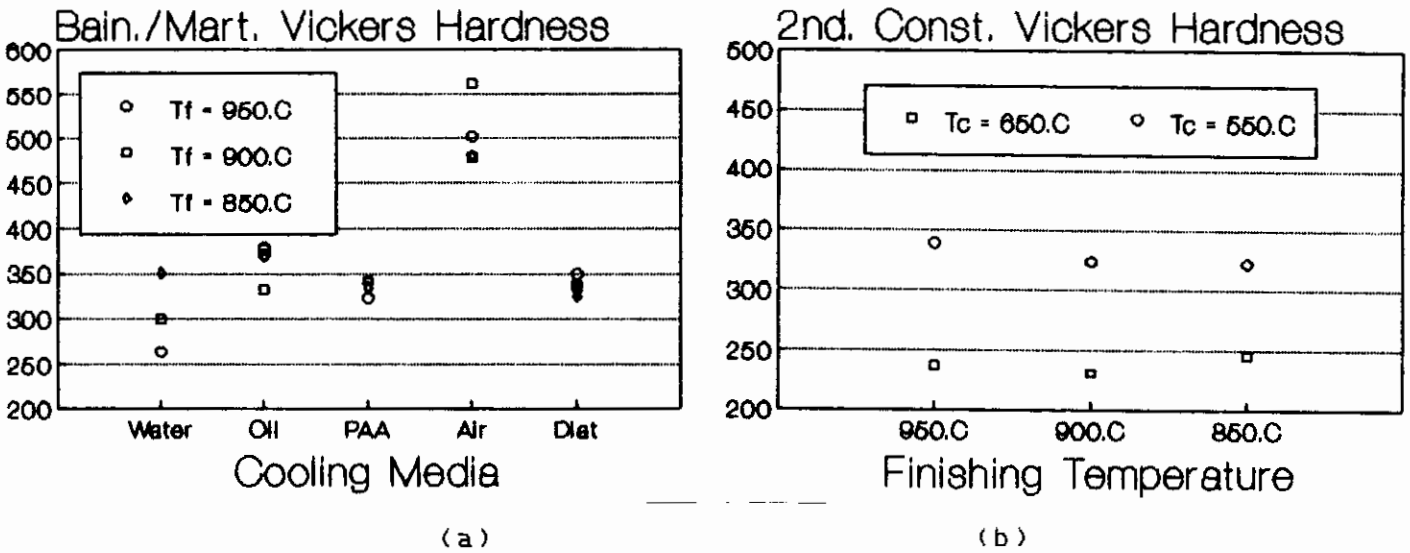
(a) (b)

Fig. 4: Second constituent grain size according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



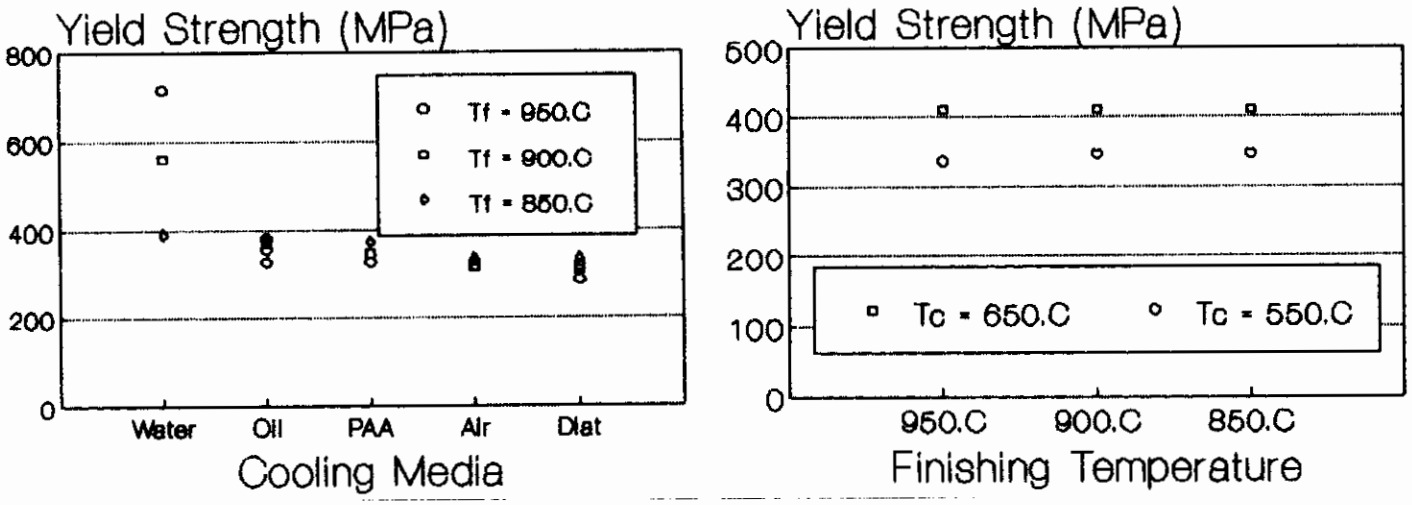
(a) (b)

Fig. 5: Ferrite hardness according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



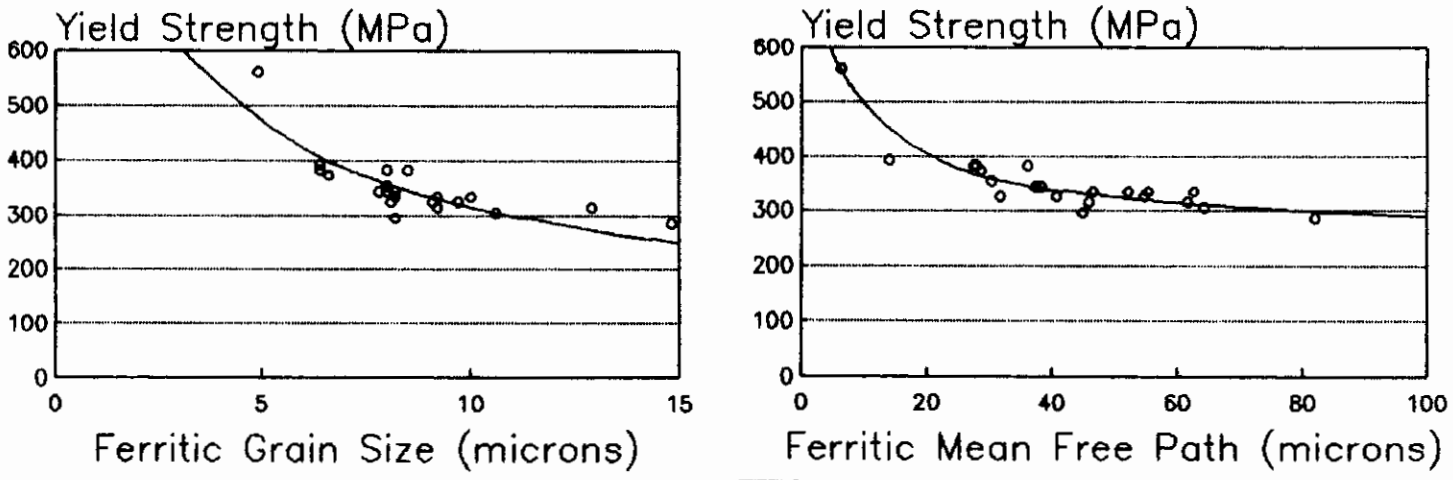
(a) (b)

Fig. 6: Second constituent hardness according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



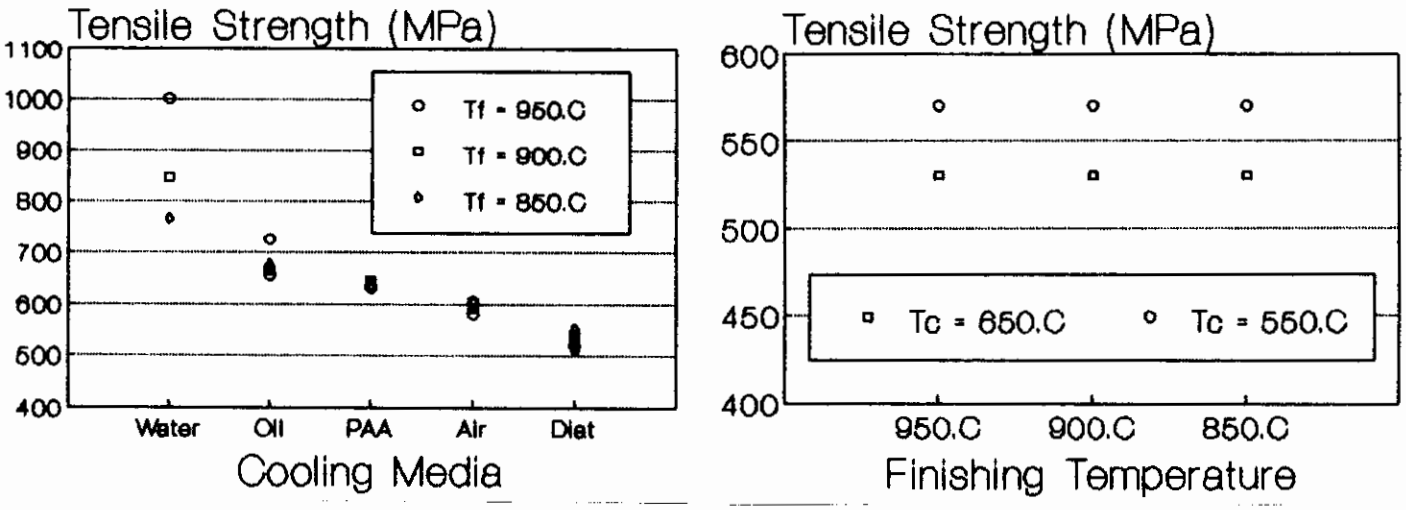
(a) (b)

Fig. 7: Yield Strength according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



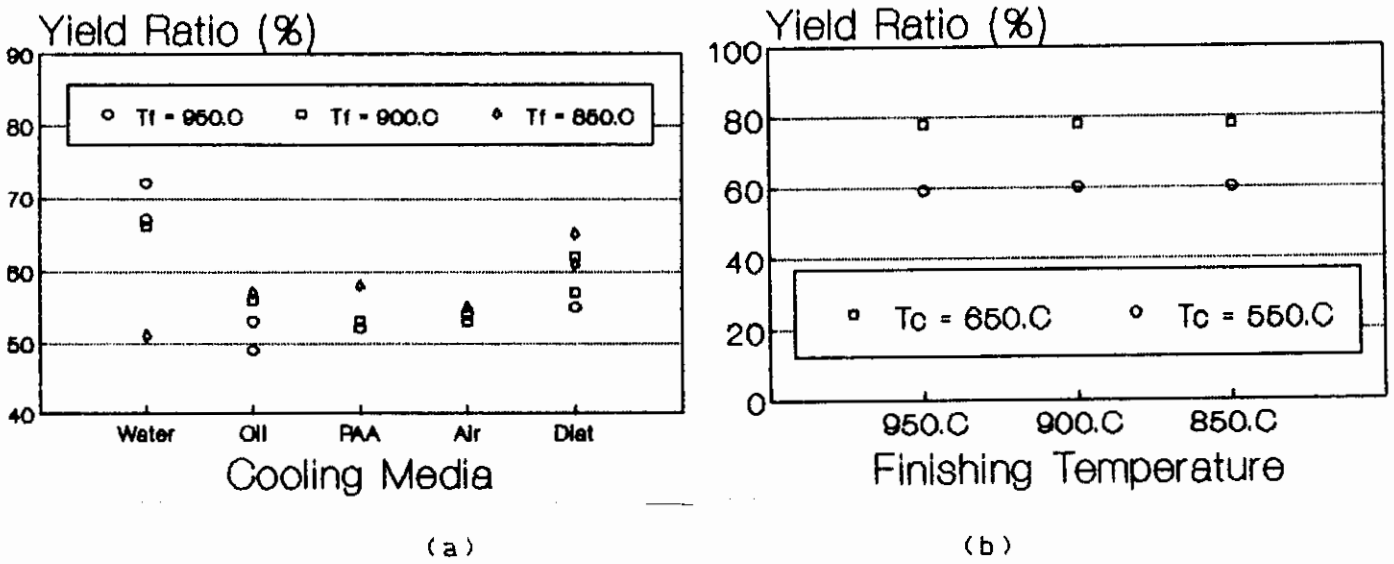
(a) (b)

Fig. 8: Relationship between the yield strength and a) mean ferritic grain size; b) mean ferritic free path.



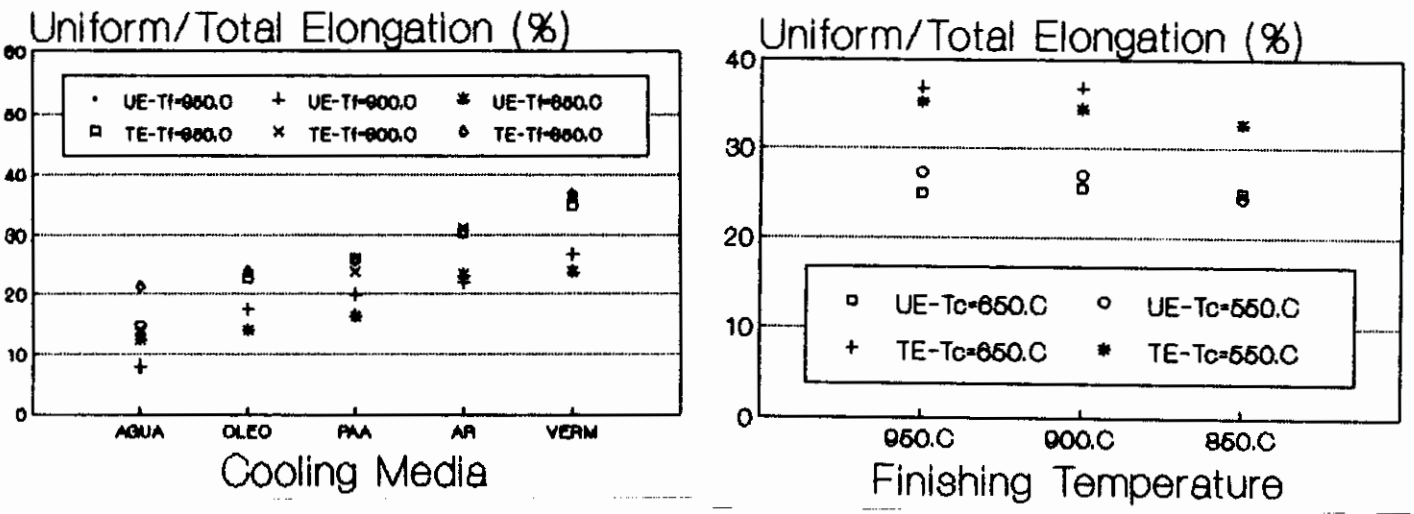
(a) (b)

Fig. 9: Tensile strength according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.



(a) (b)

Fig. 10: Yield ratio according the rolling parameters. a) Direct cooling; b) Interrupted cooling.



(a) (b)

Fig. 11: Uniform and total elongation according to the rolling parameters. a) Direct cooling; b) Interrupted cooling.