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Evolution of austenite static recrystallization and grain size during hot rolling of a V-microalloyed steel

Manuel Gómez, Lucía Rancel, Bernardo J. Fernández, Sebastián F. Medina

National Center for Metallurgical Research, (CENIM-CSIC), Av. Gregorio del Amo 8, 28040 Madrid, Spain.

mgomez@cenim.csic.es, smedina@cenim.csic.es

Abstract

Laboratory double-deformation isothermal tests and multipass continuous-cooling hot torsion tests were used to study the static recrystallization of austenite under isothermal and anisothermal conditions as well as to simulate the hot rolling of a 0.13% V-microalloyed steel. Characterization of the evolution of austenite microstructure was carried out. It has been verified that no-recrystallization temperature (T_{nr}) approximately corresponds to the temperature where recrystallization starts to be incomplete during rolling. However, incomplete recrystallization is visually evident at temperatures 25-50 ºC below T_{nr}, where grain elongation and increase in aspect ratio with temperature drop start to be significant. An accurate method to estimate the recrystallized fraction during hot rolling from stress-strain data and with no need of metallographic studies has been designed. The results of this method have been compared to metallographic measurements, the values of anisothermal fractional

softening and the accumulated stress measured in the MFS plots at T< T_{nr}. A pronounced austenite grain refinement has been detected in the first hot rolling passes after reheating, as grain size decreases from 155 μm to 27 μm in six passes. If the effect of grain size on recrystallization and precipitation is taken into account, the correlation of isothermal and continuous cooling tests as well as the relationship between SRCT and T_{nr} or RLT temperatures can be better understood.

Keywords

Microalloyed steels; Hot rolling; Static recrystallization; Austenite grain size; Anisothermal Softening.

1. Introduction

The static recrystallization of microalloyed steels after hot deformation can be obstructed by the pinning effect exerted by strain induced precipitates. In the curves that represent the recrystallized fraction against post-deformation isothermal holding time ($X_a(t)$), the inhibition of recrystallization by precipitates is manifested by the formation of plateaus that temporarily interrupt the typical sigmoidal shape of an Avrami´s law [1-3]:

$$X_a = 1 - \exp\left(-\ln 2\left(\frac{t}{t_{0.5}}\right)^n\right) \quad (1)$$

The exponent n usually takes values between 1 and 2 [1,4,5] and according to some authors it depends on temperature [6-8]. $t_{0.5}$ is the time corresponding to a recrystallized volume fraction of 50%, which depends on temperature T, equivalent pass strain (ε), strain rate ($\dot{\varepsilon}$), austenite grain size D and chemical composition of steel according to [9]:

$$t_{0.5} = A\varepsilon^{p}\dot{\varepsilon}^{q}D^{s}\exp\left(\frac{Q}{RT}\right) \quad (2)$$

where Q is the apparent activation energy, R is the universal gas constant (8.3145 $Jmol^{-1}K^{-1}$) and p, q, s and A are constants.

The temperature limit between the stages of full recrystallization and beginning of recrystallization inhibition as a consequence of precipitation comes at a point known as Static Recrystallization Critical Temperature (SRCT) [9,10]. On the other hand, in a continuous-cooling multipass thermomechanical test, the temperature below which the recrystallization of austenite during the interpass time between successive passes starts to be incomplete is known as temperature of no-recrystallization (T_{nr}) [2].

The microstructure of austenite just before cooling to room temperature has a major influence on final ferrite microstructure [11,12]. Therefore, one of the most important aspects to be studied is the accurate assessment of the strengthening state of austenite and the quantification of the volume fraction of recrystallization during rolling at temperatures below T_{nr} and especially at the end of rolling, near A_{r3}. To carry out this

characterization, metallographic studies can be used [13,14], but metallography is a time-consuming and not always successful technique. Several mathematical models have been developed for the study of recrystallization or softening kinetics during hot rolling of steels, i.e. under non-isothermal conditions [15,16]. The “anisothermal fractional softening” method [17,18] gives an approximation of the recrystallized fraction, but this technique includes the contribution of recovery to softening, which can be very high under certain deformation conditions and/or material characteristics [19]. Besides, it is sometimes necessary to know other empirical constitutive parameters or to make preliminary tests to approximate the yield stress of a fully recrystallized material [18,20].

Niobium is known to be the microalloying element that most delays static recrystallization kinetics, even when it is in solution in the austenite [1,21]. The influence of vanadium (dissolved or precipitated) on static recrystallization is weaker but still significant, as can be seen in the increase in the values of $t_{0.5}$, activation energy and SRCT for higher V additions [7,22-24]. A strong solute drag effect of V on the kinetics of metadynamic recrystallization has also been described [24]. In this work the evolution of static recrystallization of austenite is characterized in a vanadium microalloyed steel. Thermomechanical tests are carried out under isothermal and continuous cooling conditions and conclusions about the relationship between the results coming from both thermal paths are extracted. An empirical method to estimate the recrystallized fraction during rolling from the stress-strain data obtained in the thermomechanical tests is presented and comparison with the results of metallography and the technique of anisothermal fractional softening is done.

2. Experimental Procedure

The steel studied was manufactured by Electroslag Remelting (ESR) in a laboratory unit capable of producing 30 kg ingots. This technique avoids macrosegregation, both in alloying elements and impurities, and there is considerably less microsegregation. These effects are often present in conventional ingots and continuous casting billets. As Table 1 shows, the steel has 0.48% carbon content and 0.13% vanadium and contains a residual Ti content.

Double-deformation isothermal tests as well as multipass hot rolling simulation tests were carried out in a computer-controlled hot torsion machine, on specimens with a gauge length of 50 mm and a diameter of 6 mm. Prior to the torsion tests the specimens were austenitized during 10 min at 1200 ºC, a temperature high enough to completely dissolve the V precipitates [25]. However, it should be taken into account that, despite the very low Ti content, a certain amount of undissolved TiN can remain after reheating, as this type of precipitate is very stable [26]. The calculated temperature for its complete solution in austenite is 1368 ºC. The calculated solubility temperatures and the reheating conditions are shown in **Table 2**.

After reheating, the temperature was rapidly lowered to that corresponding to the first deformation pass in each test. In the isothermal double-deformation experiments, this temperature was maintained after deformation during a certain holding time, after which a longer second deformation was applied in order to calculate recrystallized fraction (X_a). The value of X_a was measured using the “back extrapolation” technique [27]. This

method presents the advantage of determining the recrystallized fraction instead of the softened fraction, i.e. the effect of recovery is excluded from the double deformation data [13], so more accurate comparison with observed microstructures can be carried out.

Temperatures of 950 ºC, 1000 ºC, 1050 ºC and 1100 ºC and several holding times between 1 s and 500 s were used in order to build the curves of recrystallized volume fraction against time (X_a(t)) as well as the Recrystallization-Precipitation-Time-Temperature (RPTT) diagram [7,28-30], where the interaction between these two processes is clearly observed. The equivalent strain (ε) applied in the first pass was 0.20, lower than the critical strain which leads to the start of dynamic recrystallization. The strain rate ($\dot{\varepsilon}$) was kept at 3.63 s^{-1}.

In some samples the second deformation after holding time was replaced by a water-quench and subsequent metallographic preparation. Microstructure was observed in optical microscope to verify the accuracy of back extrapolation method in determining the recrystallized fraction and to compare with the results of multipass simulations. All the microstructural studies carried out in this study were done observing more than 20 fields on a longitudinal surface of the specimens at 2.65 mm from the axis. To reveal the prior austenite grain boundaries, the samples were etched with an aqueous solution of saturated picric acid mixed with a wetting agent. Some droplets of hydrochloric acid were added just before etching to activate the solution.

In the hot rolling simulation tests, the temperature of the first pass was 1150 ºC. The simulations consisted of 20 passes made under continuous cooling conditions, with a

temperature step of 25 ºC between passes, the last pass being carried out at 675 ºC. The time between successive deformations, known as "interpass time" (Δt) was equal to 20 s. The strain and strain rate applied in each pass were the same as those used in isothermal tests ($\varepsilon = 0.20$, $\dot{\varepsilon} = 3.63\ s^{-1}$). After determining the critical hot rolling temperatures (T_{nr}, A_{r3}), supplementary quench-interrupted tests were carried out in order to evaluate the evolution of microstructure (grain size, recrystallization) during rolling. In these tests, samples were deformed following the same schedule until they were water-quenched from certain temperatures along rolling schedule. In most cases the sample suffered a last deformation step and then the temperature was lowered 25 ºC during the corresponding Δt to reach the quenching temperature, so microstructure just before the application of the rolling pass was assessed. In some case, the sample was quenched immediately after deformation to observe the microstructure prior to a hypothetical cooling from a certain last hot rolling-pass temperature. The austenite grain size ($D\gamma$) and aspect ratio were determined by means of the linear intersection technique. Recrystallized fraction was measured using standard point-counting metallographic technique. Distinction of recrystallized and deformed grains was based on the shape and size of the grains as well as the appearance of grain boundaries [19,31].

3. Results

3.1. Double deformation isothermal tests and multipass hot rolling simulations

First of all, the initial austenite grain size after reheating was determined. As shown in **Table 2**, the steel studied had a grain size of 155 μm after 10 min at 1200 ºC. The

torsion test gives the values of torque applied versus the number of turns made on the specimen, which are transformed respectively into equivalent stress and strain using Von Mises criterion [32]. **Figure 1a** shows the relationship between recrystallized fraction (X_a) and time. At high temperatures (1100 ºC) the recrystallization progresses continuously following an Avrami´s law until completion. However, at a lower temperature (1050 ºC) a plateau is formed, indicating a period of time where the recrystallization is inhibited by the pinning forces exerted by precipitates. After the plateau, the recrystallization progresses again until it is complete, following a similar trend to that recorded before the plateau. At the lowest temperatures (1000 ºC, 950 ºC) a double plateau can be observed. Previous studies have verified by means of differential thermal analysis (DTA) [3] and lattice parameter comparison [22,33] that this phenomenon results from the formation of two different types of precipitates having different stoichiometries and similar solubility temperatures. In the particular case of V-Ti steels, the precipitate is initially a N rich (Ti, V) carbonitride that produces the first plateau of the curves. Stoichiometry changes with holding time after deformation to a C-richer carbonitride with higher V content, which would cause the second plateau [22].

Curves of **Fig. 1a** have been used to deduce the temperatures and times corresponding to different recrystallized fractions, as well as the induced precipitation start (P_s) and finish (P_f) times, given by the times of the beginning and the end of the plateaus of inhibition of recrystallization [3]. In this way, the RPTT diagram of **Figure 1b** has been drawn. The horizontal asymptote of the P_s and P_f curves is the Static Recrystallization Critical Temperature (SRCT) [9,10]. As there are two plateaus, two P_s and P_f curves and two values of SRCT corresponding to two separated stages of precipitation can be distinguished.

Figure 2a shows the stress-strain curves of the 20 hot rolling passes simulated. At first deformations, stress raises as temperature decreases, after which there is a change in the slope with a growth in the stress that means a greater tendency to strengthening. The meaning of this figure is better explained by observing **Figure 2b,** which shows the graphic representation of mean flow stress (MFS) versus the inverse of the absolute temperature. MFS is determined in each step by dividing the area below the stress-strain curve by the strain applied. In **Fig. 2b** it is possible to distinguish three different zones corresponding to complete recrystallization between passes at high temperatures (I), accumulation of strengthening due to partial inhibition of recrystallization (II) and transformation of austenite to ferrite and pearlite (III) [2]. The intersection of the straight regression lines of phases I and II defines the value of T_{nr} and the intersection of the regression lines of phases II and III determines the value of A_{r3} [2]. The magnitude known as "accumulated stress" ($\Delta\sigma$) [34] will be given by the length of the vertical segment drawn between the regression lines of phases I and II, as illustrated in **Fig. 2b**. The value of $\Delta\sigma$ informs about the progressive strengthening of austenite during rolling at temperatures below T_{nr}. **Figure 3** shows the evolution of accumulated stress during hot rolling. $\Delta\sigma$ is zero at T = T_{nr} and grows during phase II to reach its maximum value at A_{r3} [34].

3.2. Microstructural evolution of austenite during hot rolling

Figure 4 shows the microstructural evolution of austenite during a wide range of deformation temperatures along the rolling schedule represented in **Fig. 2**. Starting from an initial grain size of 155 μm (**Fig. 4a**), grain is significantly refined by successive

recrystallizations in the first rolling passes above T_{nr}. For example, the grain is refined to 53 μm after two passes and it reaches 27 μm before applying deformation at 1000 ºC, i.e. the value of mean grain size is divided by six in six passes. Elongation of grains that do not recrystallize during interpass time is observed at lower temperatures. Although T_{nr} is equal to 897 ºC, grain elongation starts to be remarkable at temperatures near 850 ºC. In the pictures corresponding to the lowest temperatures the progressive strengthening of the austenite can be easily seen. For example, it can be observed how the grains elongate as a result of a deformation of ε = 0.20 at 800 (**Figs. 4d-4e**). On the other hand, the interpass time of 20 s is very short to permit a high degree of recrystallization at those low temperatures, so austenite accumulates a progressive strengthening and approaches to the typical "pancake" microstructure. This can be verified observing the weak effect of interpass time from 775 ºC to 750 ºC (**Figs. 4f-4g**).

The curve of **Figure 5** has been obtained from the analysis of **Fig. 4**. In this figure, the evolution of average austenite grain size and aspect ratio during rolling is graphically represented against quenching temperature. The last deformation temperature of each sample was 25 ºC higher than quenching temperature. It can be seen that austenite grain refinement occurs mainly in the first rolling passes, before reaching no-recrystallization temperature, as a result of successive recrystallizations from a relatively coarse initial grain size of 155 μm. This is in agreement with several equations proposed for austenite recrystallized grain size and interpass grain growth for CMn and microalloyed steels [1,4,5,31]. **Fig. 5** shows that the experimental values of grain size would converge at temperatures near T_{nr} with the line corresponding to recrystallized grain size proposed by Sellars [1] (assuming that prior grain size d_0 is equal to recrystallized grain size from previous pass d_{rex}, i.e. grain growth is not considered). The austenite grain size at T =

T_{nr} is close to 20 μm, which does not greatly differ from the values found by other authors [8] for Ti steels with two different Ti contents and initial grain sizes and similar deformation conditions. The substantial grain refinement occurred in the first rolling passes makes the grain size at T = T_{nr} to be practically independent on the initial microstructure after reheating [8,35]. However, the grain size depends on pass strain and finer final austenite grain sizes close to 10 μm would be desired for high strength steels. Below T_{nr}, and especially in the last passes, interpass time is not long enough for the complete recrystallization of all austenite grains. Pinning forces exerted by precipitates take comparable or even higher values than the driving forces for recrystallization [28,36,37] and some grains elongate progressively. This can be verified by evaluating the austenite grain aspect ratio [36], which represents the ratio between the number of grains detected per unit length in deformation direction and its perpendicular. In this case, the directions perpendicular and parallel to the axis of torsion sample have been respectively taken. **Fig. 5** shows that the growth of aspect ratio with the temperature drop is remarkably faster when the last pass is carried about 50 ºC below T_{nr}. The average value of aspect ratio at T = T_{nr} is about 1.1.

3.3. Approximation of recrystallized fraction during hot rolling

A method to estimate the volume fraction of static recrystallization during hot rolling without carrying out metallographic studies of quenched samples was designed. This method could be useful in cases where experimental measurement of X_a by observation of quenched samples is difficult. The first step of this method consists of the application of back extrapolation technique on pairs of stress-strain curves corresponding to

consecutive hot rolling passes (**Fig. 2a**). An "apparent" value of recrystallized fraction (X_a) is obtained. This value is affected by the temperature drop between passes, i.e. it reflects the hardening of the steel derived from cooling that can be seen in phase I of **Fig. 2b**. The value of X_a is always lower than 1 and it is practically constant and close to 0.9 (90%) at T>T_{nr}. In line with the classical definition of T_{nr} and also according to the results of the microstructural studies carried out on quenched samples (**Fig. 4** and aspect ratio of **Fig. 5**), the static recrystallization of austenite can be assumed to be complete at temperatures above T_{nr}. To subtract the effect of temperature on austenite strength this hypothesis was adopted. Therefore, the value of X_a in each pass was "normalized", i.e. it was divided by the average of the values of X_a at temperatures above T_{nr} ($\overline{X}_a$). In this way, the value of the new X_a (X_a') will be approximately equal to 1 between the first pass and T = T_{nr}. Finally, it should be taken into account that the values of X_a and X_a' inform about the austenite recrystallization during a single interpass (i.e. between the *i*th and (*i*+1)th passes) but they do not reflect the progressive accumulation of strengthening in an average mixed microstructure compared to the fully recrystallized material existing at temperatures above T_{nr}. For example, if the calculated recrystallized fraction after three consecutive passes is respectively equal to 100, 90% and 80%, the "accumulated" recrystallized fraction after the third interpass time (X_a*) would not be 80% but it should be the result of multiplying the successive recrystallized fractions: 100%·90%·80% = 72%. Summarizing, the mathematical expressions would be:

$$X_a^i = \text{Result of } \textit{Back Extrapolation} \text{ method applied on curves of passes } (i\text{-}1) \text{ and } i; \quad (3)$$

$$\overline{X_a} = \frac{\sum_{i=2}^{n} X_a^i}{n-1}; \quad n = \text{number of pass where } (T \approx T_{nr}); \quad (4)$$

$$(X_a^i)' = \frac{X_a^i}{\overline{X_a}}; \tag{5}$$

$$X_a^i * = (X_a^i)' X_a^{i-1} * \tag{6}$$

It should be taken into account that the value of $\overline{X_a}$ should be really the number that makes the average of X_a* between the first pass and T_{nr} (and not the average of X_a') to be close to 1. Then, equation (4) should be replaced by the condition:

$$\overline{X_a} \Rightarrow \frac{\sum_{i=2}^{n} X_a^i *}{n-1} = \frac{\sum_{i=2}^{n} \frac{X_a^i}{\overline{X_a}} X_a^{i-1} *}{n-1} = 1; \quad n = \text{number of pass where } (T \approx T_{nr}); \tag{7}$$

The values of X_a* depend on $\overline{X_a}$, so a numeric resolution process is necessary to make this minor correction in order to calculate $\overline{X_a}$, X_a' and X_a* more accurately. **Figure 6** shows the application of this method to calculate X_a* using the σ-ε plots of **Fig. 2a**. In agreement with the initial hypotheses, recrystallized fraction X_a* is almost constant and close to 1 until T_{nr} is attained. Below this temperature, X_a* starts to decrease and austenite accumulates a strengthening. The calculated values of X_a* have been compared to those measured by quantitative metallography (X_{am}) on microstructures of quenched samples like those shown in **Fig. 4**. **Figure 7** shows an optical micrograph of same sample as **Fig. 4g** taken at higher magnification. Deformed (unrecrystallized) grains are characterized by larger size, elongated shape and usually serrated boundaries. On the other hand, recrystallized grains are finer and more equiaxed [19,31]. The amounts of both kinds of grains were measured by point counting in several fields and the standard area fraction was calculated to quantify the recrystallized volume fraction

(X_{am}). Calculated and observed recrystallized fractions are similar so it can be concluded that the method proposed is adequate and reflects with reasonable accuracy the evolution of recrystallization during hot rolling.

From the stress-strain curves obtained in multipass torsion tests (**Fig. 2a**) it was also possible to determine the values of the "fractional softening" that takes place between passes. The anisothermal fractional softening (*FS*) was calculated by means of the following expression [8,17,35,38,39]:

$$FS(\%) = \frac{\sigma_m^i - \sigma_y^{i+1}\dfrac{\sigma_0^i}{\sigma_0^{i+1}}}{\sigma_m^i - \sigma_0^i} \times 100 \qquad (8)$$

where σ_m^i and σ_y^{i+1} are the maximum and the yield stresses for both, the *i*th (at temperature T_i) and (*i*+1)th (at temperature T_{i+1}) passes, respectively. σ_0^i and σ_0^{i+1} are the yield stresses of a fully recrystallized material for the *i*th and (*i*+1)th passes. The yield stresses were determined by the 2% offset method. σ_m^i and σ_y^{i+1} are based on the pass-to-pass flow curves of **Fig. 2a.** σ_0^i and σ_0^{i+1} are determined from the linear relationship derived from the values of yield stresses (σ_y) measured in the stress-strain curves of **Fig. 2a** corresponding to the temperature range of complete recrystallization ($T>T_{nr}$).

The values of *FS* are also represented against temperature in **Fig. 6**. It can be seen that the value of anisothermal fractional softening at temperatures above or close to T_{nr} properly eliminates the influence of temperature on stress, so the values of *FS* are close

to the calculated recrystallized fraction X_a* and the value of X_{am} measured by optical microscopy (i.e. near 100%). However, at lower temperatures, mixed (i.e. partially recrystallized) microstructures appear. As mentioned above, the "back extrapolation" method suppresses the recovered fraction from the total softened fraction, so it yields lower values than the double-deformation method used to calculate *FS*. The fraction of recovery can be remarkably high for short interpass times or low temperatures, i.e. when the degree of recrystallization is lower [19,31]. Besides, the "anisothermal fractional softening" method does not seem to consider accurately the accumulation of successive partial recrystallizations at temperatures lower than T_{nr} that is expressed in equation (6). As a result, the values of *FS* start to deviate from X_a* at lower temperatures and at the end of hot rolling, near A_{r3}, the values of *FS* are more than 20% higher than calculated X_a* and X_{am} measured by metallography.

It is also interesting to note that the percentage of recrystallization when T_{nr} is reached is not exactly 100%, but about 90%. It is obvious that the accuracy in the determination of T_{nr} is limited by the temperature step of 25 ºC in multipass hot torsion tests. Besides, the method to determine T_{nr} through the increase in MFS will always need an appreciable degree of hardening that can easily correspond to 5-10% of unrecrystallized fraction. Furthermore, the presence of particles at temperatures below the precipitation start will affect to some extent the values of yield, maximum and mean flow stress and consequently the results of these techniques. On the other hand, several authors have previously shown that T_{nr} does not necessarily correspond to the exact transition from complete to partial recrystallization and, depending on interpass time and pass strain, there can be a certain amount of unrecrystallized austenite after the pass immediately before T_{nr} [8,10,35,39]. Some authors define the lowest temperature above which

recrystallization between passes is complete (i.e. higher than 85-95%) as the "recrystallization limit temperature" (RLT) [40]. The "recrystallization stop temperature" (RST) would be the highest temperature at which recrystallization is completely absent (i.e. less than 5%) [40]. Radovic et al. [39] approximated the *FS* versus (1/T) plots by three linear segments, which intersect at two temperatures that can be assimilated to RLT and RST. These authors found that T_{nr} correlates well with RLT, but Abad et al [35] affirm that T_{nr} values are located between RLT and RST. **Figure 8** shows that RLT is slightly higher than T_{nr}. The second change in slope that determines RST is not easy to locate, probably because X_a* is higher than 10% even at low temperatures close to A_{r3}. For example, **Fig. 7** shows that very small, equiaxed recrystallized grains can be found in the sample deformed at 775 ºC and quenched from 750 ºC. This means that although the austenite has suffered several deformations in the "no-recrystallization" temperature range, a certain fraction of recrystallization during interpass time is still possible at those low temperatures.

On the other hand, the comparison of **Fig. 3** to **Figs. 4-8** lets to conclude that the increasing value of $\Delta\sigma$ as long as samples are cooled is an accurate qualitative indication of the progressive strengthening of austenite below T_{nr} [34].

3.4. Effect of successive hot rolling passes. Comparison between SRCT and T_{nr}

A comparative study of the values of T_{nr} and SRCT was attempted with the assistance of plots and pictures of **Figs. 1a, 2b**, **4** and **5**. Comparison of these parameters is not easy, as both represent the start of inhibition of recrystallization but they do not have the

same definition [10]. SRCT is determined by isothermal tests and is independent of the time, whereas T_{nr} is determined by means of continuous cooling rolling simulations and depends on Δt. T_{nr} presents a value lower than SRCT in the steel studied. According to theory, the value of T_{nr} in microalloyed steels would be higher than SRCT for shorter interpass times and would pass through a minimum, which occurs at Δt = 10-15 s. This interval reflects the solute drag effect of alloying elements [41]. For medium times, T_{nr} raises with increasing interpass times because of the higher volume fraction of fine precipitates that exert a strong pinning effect. Finally, for interpass times longer than 50 s, T_{nr} decreases again as a consequence of particle coarsening [39,41].

Fig. 1a indicates that the value of static recrystallization at 950 ºC when isothermal time after deformation is 20 s is approximately equal to 40 %. At that point, the curve of X_a is in the end of first the plateau of inhibition of recrystallization by precipitation. However, **Fig. 2b** shows that austenite presents complete recrystallization at 950ºC ($T>T_{nr}$) with an interpass time of 20 s. As seen in **Fig. 5**, the application of successive rolling passes (especially at temperatures above T_{nr}) provokes an austenite grain refinement. As equation (2) shows, this refinement accelerates recrystallization kinetics. When grain is refined, the area of austenite grain boundaries (potential nucleation sites) increases. Besides, a reduction in grain size can contribute together with strain applied to augment the stored energy due to deformation [42]. The curves of recrystallized fraction against time are then shifted to shorter times. Thus, a steel that is being deformed at temperatures below SRCT can pass the stage of inhibition of recrystallization by induced precipitation delimited by the plateaus in **Fig. 1a** and reach complete recrystallization in a much shorter time than the indicated in $X_a(t)$ curves. In this case, the strain induced precipitation could be expected to happen after rolling

passes applied at temperatures above T_{nr}, as grain refinement also accelerates precipitation [43].

Acceleration of recrystallization (displacement of $X_a(t)$ curves towards shorter times and drop in SRCT value) derived from grain refinement is verified observing the microstructures shown in **Figures 4b** and **9.** Both specimens have been quenched from 950 ºC after waiting 20 s following a deformation of $\varepsilon = 0.20$. However, the first one corresponds to a simulation of 8 passes with a last deformation at 975 ºC and the second is taken from a double deformation isothermal test. In other words, **Fig. 4b** corresponds to the ninth point in the MFS curve of **Fig. 2b** ($T>T_{nr}$) and the eight point in **Fig. 6** ($X_a^* \approx 100\%$) whereas **Fig. 9** coincides with the end of the first plateau in **Fig. 1a** ($X_a = 40\%$). As a result, the sample from the rolling simulation test presents fine and rather equiaxed grains, while the sample taken from double deformation test shows a mixed microstructure with many coarser, elongated grains. The accuracy of the back extrapolation method in determining the recrystallized fraction in isothermal tests can be observed comparing the calculated value of X_a with the measured value by optical microscopy (about 35%).

To make a more appropriate comparison between T_{nr} and SRCT, experimental value of SRCT obtained in isothermal double-deformation tests should be replaced by a calculated value where the influence of actual austenite grain size before deformation in the multipass tests is considered. According to the model of Medina et al. for SRCT, this temperature depends on several variables affecting microstructural evolution during hot deformation (strain ε, strain rate $\dot{\varepsilon}$, initial grain size D and chemical composition) [9]:

$$SRCT\ (K) = T_s - 708.2[X \cdot 1000]^{0.38} \varepsilon^{0.3} \dot{\varepsilon}^{0.16} D^{-0.34} \quad (9)$$

where T_s is the solubility temperature of VN and X represents (%V)(%N) for V-microalloyed steels. It can be seen that SRCT is equal to T_s when strain or strain rate are zero.

In **Fig 5**, it can be seen that austenite grain size is equal to 19 μm when temperature is close to T_{nr} (897 ºC). If this value is introduced in equation (9), then the calculated SRCT would be equal to 857 ºC, i.e. much closer and even lower than T_{nr}. **Figure 10** shows the result of making this calculation in several hot rolling simulation passes. It can be seen that, as long as austenite grain is refined, SRCT value would converge to T_{nr}. There is however a certain error that can be partially attributed to the discrepancy between the model of Eq. (9) and the experimental value of SRCT (1076 ºC). It can be expected that the residual amount of Ti (0.003%) would cause TiN particles to precipitate at high temperatures. These particles would serve as nucleation sites for new strain-induced vanadium precipitates [22] and experimental SRCT would be slightly higher than the value obtained with the model.

4. Conclusions

- By means of thermomechanical tests and metallographic studies carried out in a V-microalloyed steel, it has been verified that, for the deformation conditions used, no-recrystallization temperature (T_{nr}) approximately corresponds to the temperature where

recrystallization starts to be incomplete, as observable by the increase of MFS in a multipass torsion test.

- However, incomplete recrystallization is visually evident at temperatures 50 ºC below T_{nr}, where grain elongation and increase in aspect ratio with temperature drop start to be significant.
- The fraction of austenite static recrystallization during hot rolling (X_a*) can be estimated by means of a calculation method that only needs the stress-strain data of thermomechanical tests. Comparison with metallographic studies verifies the accuracy of this method.
- The value of anisothermal fractional softening (*FS*) is close to the recrystallized fraction at temperatures above or close to T_{nr}. However, at lower temperatures, this parameter overestimates the value of austenite recrystallization during hot rolling because it includes the percentage of recovery and does not properly consider the accumulation of strengthening between successive passes. Consequently, the value of *FS* at temperatures near A_{r3} could be more than 20% higher than X_a*.
- The value of T_{nr} obtained in mean flow stress (MFS) *vs*. (1/T) plots is close to the value of recrystallization limit temperature (RLT) measured in *FS*-(1/T) plots.
- Accumulated stress measured in the MFS plots represents a suitable indication of the progressive strengthening of austenite below T_{nr}.
- A large proportion of austenite grain refinement is achieved in the first hot deformation passes after reheating, i.e. at temperatures much higher than T_{nr}. For example, the value of grain size measured after interpass time is divided by six in the first six passes.
- The effect of grain size on recrystallization and precipitation kinetics and their mutual interaction helps to explain the discrepancy in the microstructure observed at the same

quenching temperature on isothermal double-deformation tests and continuous-cooling multipass hot rolling simulations. Besides, if this effect is considered, SRCT temperature determined in isothermal double deformation tests approaches to T_{nr} obtained in multipass continuous cooling tests.

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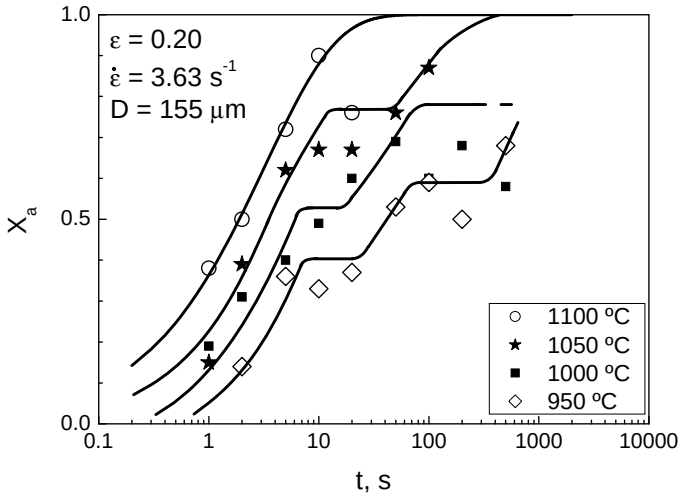
Table 1. Chemical composition of the steel studied (weight %).

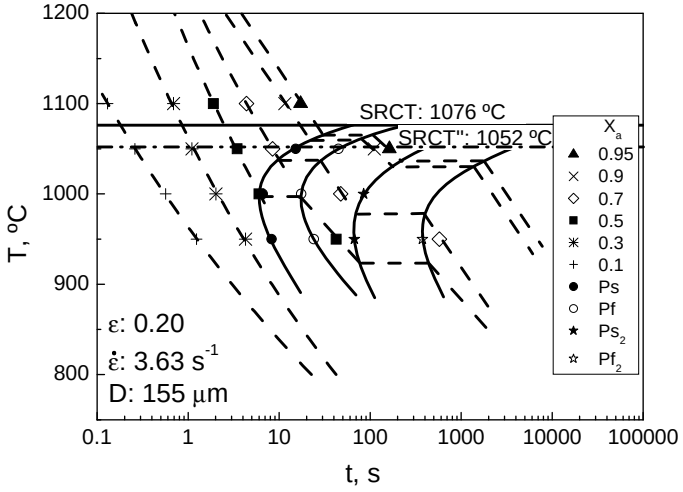
C	Si	Mn	P	S	Al	Cr	V	Ti	N	O
0.48	0.28	1.45	0.024	0.018	0.009	0.22	0.13	0.003	0.0200	0.0044

Table 2. Calculated solubility temperatures of the steel studied [25]. Reheating temperature (reheating time = 10 min) and measured austenite grain size (D_γ) at reheating temperature.

Solubility temperatures (ºC)				Reheating temperature (ºC)	D_γ (μm)
VN	$VC_{0.75}$	TiN	TiC		
1141	904	1368	979	1200	155

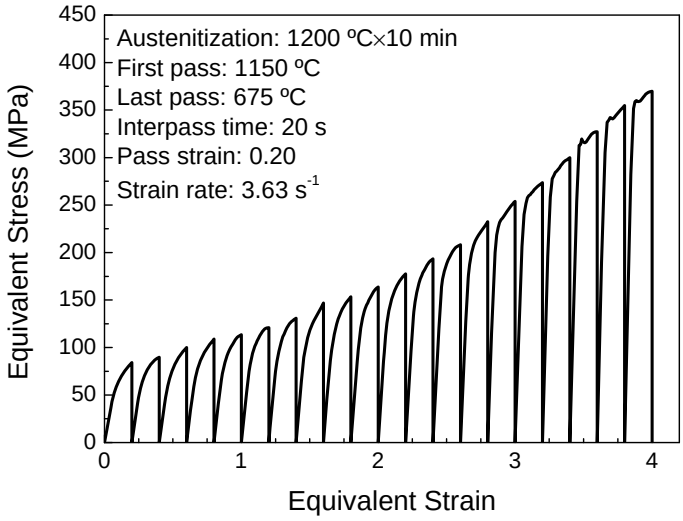


a)

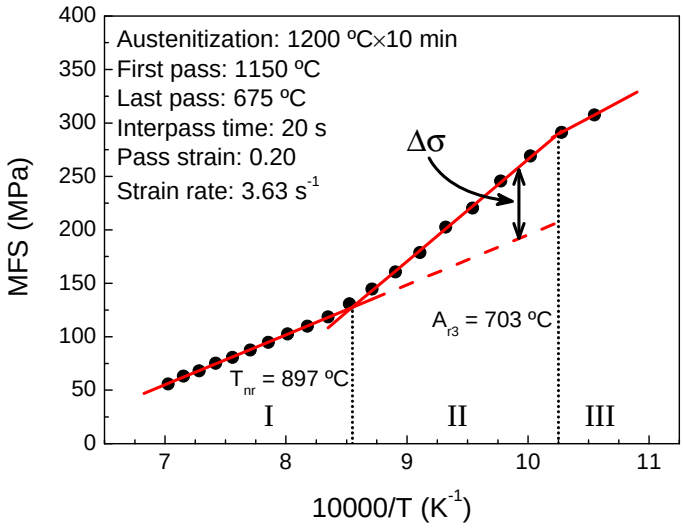


b)

Fig. 1. Double deformation isothermal test. **a)** Volume fraction of static recrystallization (X_a) versus holding time after deformation; **b)** RPTT diagram.



a)



b)

Fig. 2. Multipass hot rolling simulation. **a)** Stress-strain curves corresponding to a 20-pass hot torsion deformation schedule; **b)** Mean flow stress versus the inverse of absolute temperature.

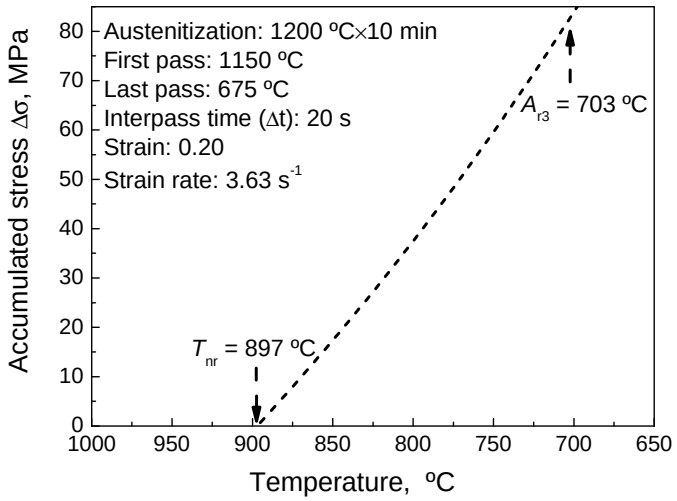
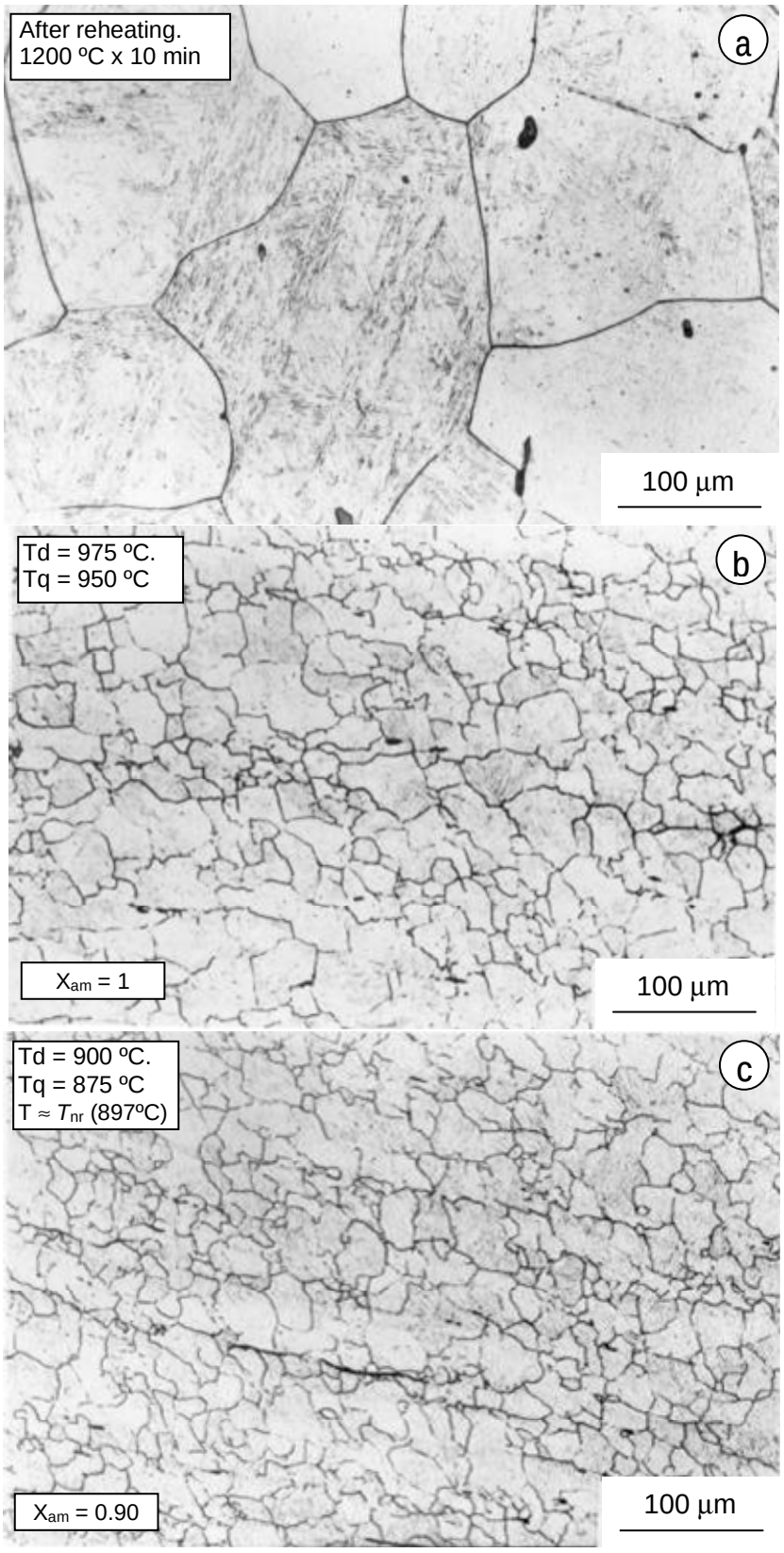
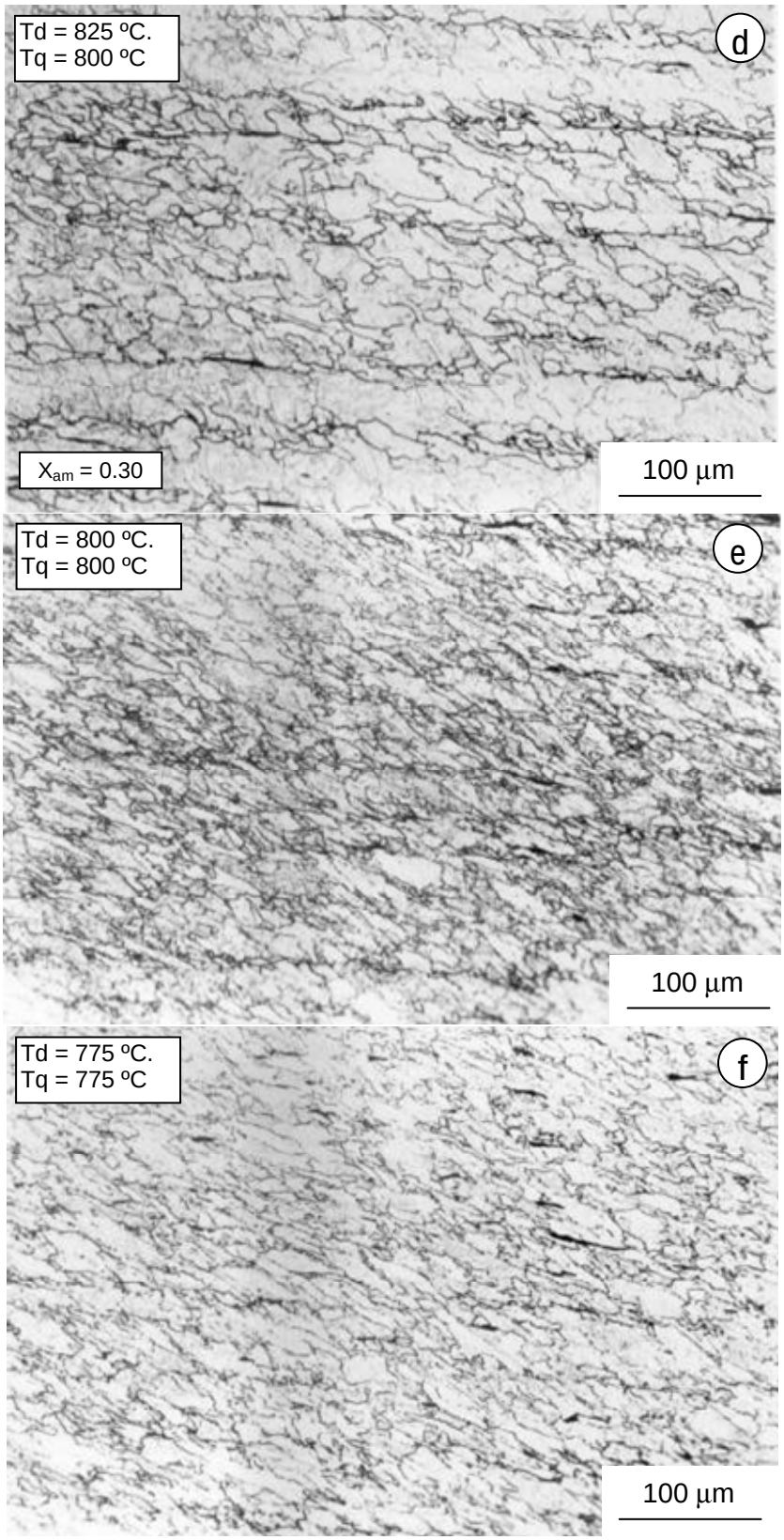


Fig. 3. Evolution of austenite accumulated stress ($\Delta\sigma$) during hot rolling.


After reheating.
1200 ºC x 10 min
a
100 µm
Td = 975 ºC.
Tq = 950 ºC
b
Xam = 1
100 µm
Td = 900 ºC.
Tq = 875 ºC
T ≈ Tnr (897ºC)
c
Xam = 0.90
100 µm


Td = 825 ºC.
Tq = 800 ºC
d
X_{am} = 0.30
100 µm
Td = 800 ºC.
Tq = 800 ºC
e
100 µm
Td = 775 ºC.
Tq = 775 ºC
f
100 µm

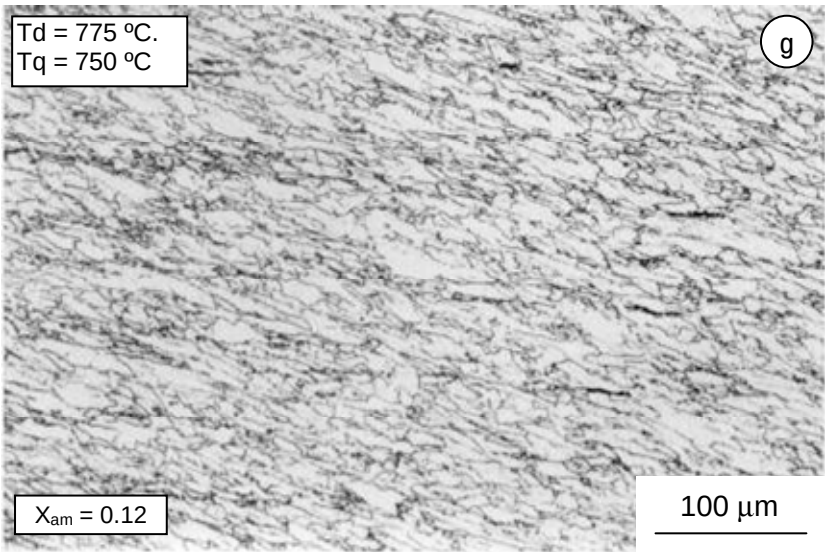


Fig. 4. Microstructures obtained at different stages of hot rolling simulation. ε = 0.20; $\dot{\varepsilon}$ = 3.63 s^{-1}; Δt = 20 s; "Td" means deformation temperature and "Tq" quenching temperature. X_{am} is the recrystallized fraction measured by metallography after observation of several fields.

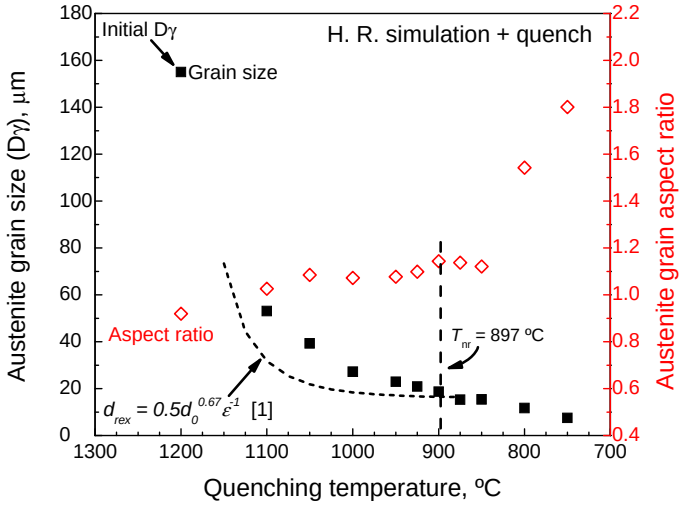


Fig. 5. Evolution of austenite microstructure as a function of deformation temperature during hot rolling simulation. Average austenite grain size and average austenite grain aspect ratio.

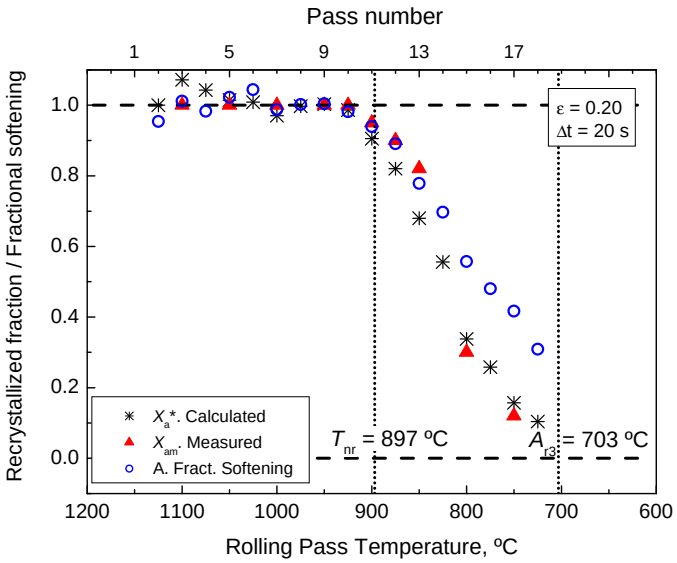


Fig. 6. Estimation of statically recrystallized fraction (X_a*) during a hot rolling simulation using Eqs. (3)-(7). Comparison with experimental values measured by optical microscopy (X_{am}) and anisothermal fractional softening measured with Eq. (8).

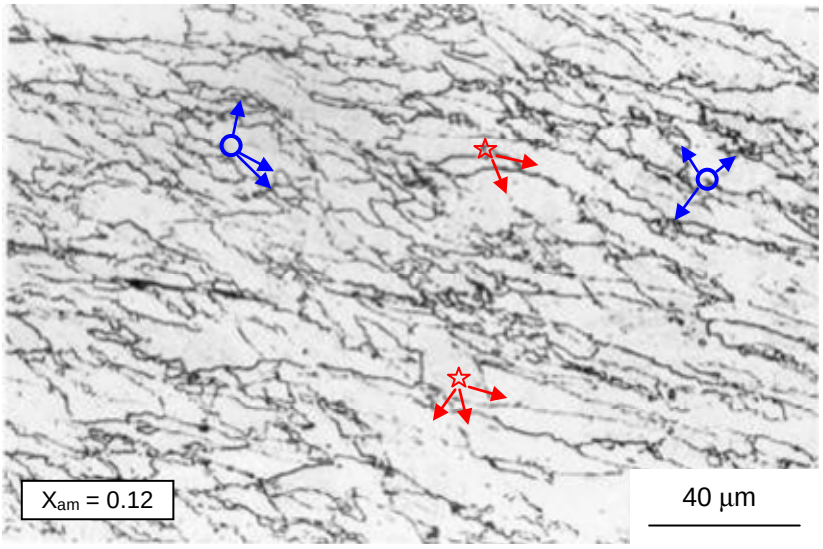


Fig. 7. Microstructure obtained after a hot rolling simulation of 16 passes. ε = 0.20; $\dot{\varepsilon}$ = 3.63 s^{-1}; Δt = 20 s; last pass temperature = 775 ºC; quenching from 750 ºC (same sample as **Fig. 4g**). Calculated recrystallized fraction (X_a*) = 0.16; Measured recrystallized fraction (X_{am}) = 0.12; Fractional softening (*FS*) = 0.42. Examples of recrystallized grains are indicated by circles and unrecrystallized grains are denoted by stars. X_{am} is always obtained after observation of several fields.

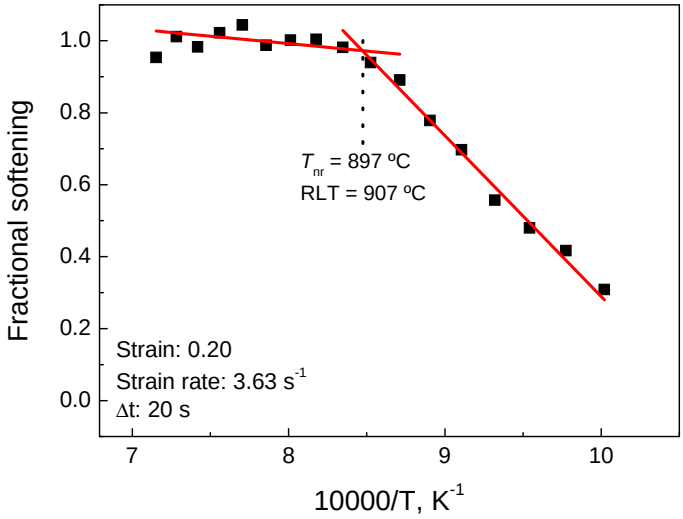


Fig. 8. Anisothermal fractional softening versus the inverse of pass absolute temperature.

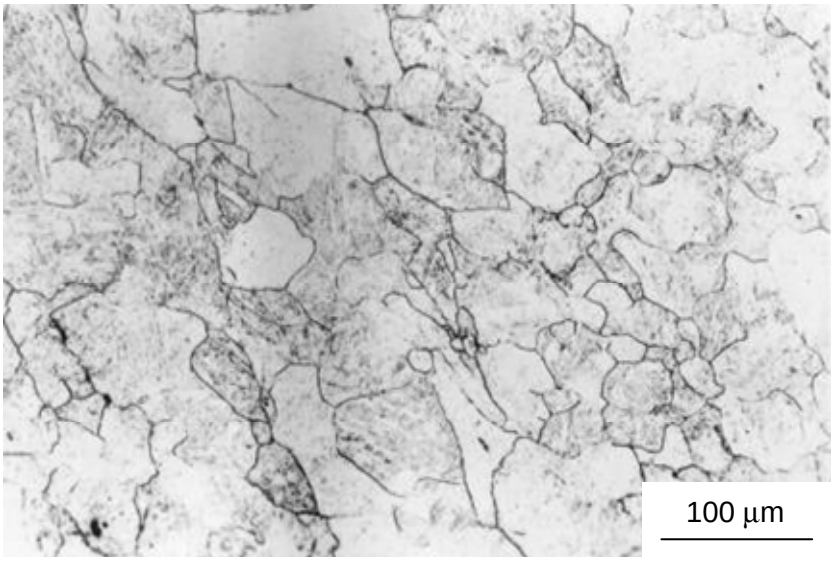


Fig. 9. Microstructure of austenite deformed at 950ºC. ε = 0.20; $\dot{\varepsilon}$ = 3.63 s^{-1}; Td = Tq = 950 ºC; Isothermal holding time = 20 s; X_a = 0.40, X_{am} = 0.35.

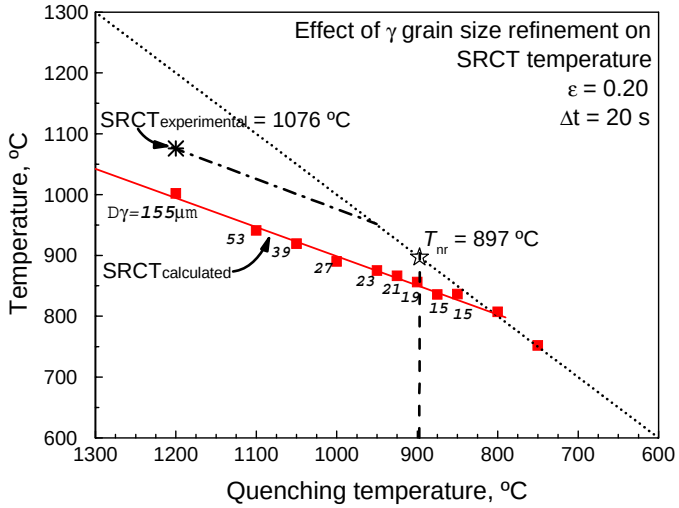


Fig. 10. Effect of austenite grain refinement on the temperature below which static recrystallization is incomplete. The values of grain size shown in **Fig. 5** (numbers in *italics*) are introduced in Equation (9) to estimate SRCT temperature in each rolling pass.