DETERMINATION OF THE CRITICAL STRAIN FOR GRAIN COARSENING DURING HOT ROLLING

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Abstract

During recrystallisation annealing after cold forming, it has been repeatedly observed that there is a critical range of strains beyond which enormous coarsening of the statically recrystallized grain occurs. It can be assumed that similar behaviour can also be expected during hot rolling with static recrystallisation (SRX) between passes, especially in the first passes at the highest temperatures. This paper aims to experimentally determine the effect of strain on the grain size and to find the critical strain range for grain coarsening. For this purpose, a plain strain compression test (PSCT) was performed on the Hydrawedge II simulation module cooperating with the electrical and hydraulic systems of the HDS-20 simulator (Gleeble 3800). A total of 18 tests were performed for three different deformation temperatures (1000, 1050 and 1100 °C) and 5 different interpass times after deformation (1 to 200 s). Thanks to PC simulation using FEM in the Simufact Forming program, the equivalent strain distribution in the deformed sample was determined. Thanks to this, it was possible to determine the dependence of the size of the original austenite grain after deformation (obtained metallographically) on delay at a given temperature for a strain range of 0.02 to 0.51. Based on the experiments, graphs of the dependence of the size of the original austenite grain on the equivalent strain were compiled. The results showed that in the strain range of 0.05 to 0.15, there was an enormous grain coarsening, especially at the highest temperatures. The achieved results were compared with theoretical values obtained using calculations of SRX kinetics, recrystallized grain size and grain coarsening kinetics.

Keywords: Microalloyed steel, hot rolling, recrystallisation, grain growth, PSCT

1. INTRODUCTION

When distributing gases to consumers or transporting gases in general from production sites to refineries and beyond, the costs associated with their transportation are a crucial aspect. These can be significantly reduced by increasing the operating pressure of the gas flowing through the pipeline. This naturally places greater demands on the pipe itself, where the wall thickness of the pipe (the thickness of the strip/sheet that precedes the production of the welded pipe) or the strength properties of the material used plays a key role. Higher operating pressures will be better handled by a pipe made of a stronger material with a greater wall thickness. If the current level of operating pressure is maintained, then the use of a higher strength material will allow the use of tubes with a lower wall thickness, resulting in a more economical use of raw materials in steel production.

These are the main reasons for the increasing requirements for thermomechanically rolled strips used in the production of welded tubes in grades L485M/X70M, and L555M/X80M. Mining in the Arctic regions, or the transport of gases from these regions, then predetermines more stringent requirements for brittle properties.

These are assessed by the standard Charpy test, which determines the steel's resistance to cracking, and by the drop-weight test, which assesses the material's resistance to crack growth.

The welded tube production process, which logically follows the hot rolled strip production process, modifies to some extent the properties achieved on the thermomechanically rolled strip. In the case of spiral-welded tubes, the Bauschinger effect is often referred to, which results in a reduction in the mechanical strength properties evaluated by tensile testing. Cold plastic deformation or an increase in dislocation density then explains the change in ductile-brittle transition temperature (shift to higher temperatures). It follows that, to obtain satisfactory mechanical properties on a spiral welded tube, the strips used for its manufacturing must meet more stringent requirements. In the case of toughness, for example, as determined by the drop weight test mentioned above, this is ensured by the fact that a minimum percentage of ductile fracture (85 % of the specimen area) is obtained on specimens taken from strips tested at minus 30 °C. However, samples taken from tubes are tested at minus 20 °C. Stricter criteria also apply to the evaluation of mechanical strength properties by tensile testing. The values of yield strength and tensile strength obtained on the strips must be increased so that these properties (considering the Bauschinger effect) also meet the normative requirement on specimens taken from the tubes after bending.

One of the key structural factors that affect the ductile-brittle properties is the homogeneity of the structure [1,2]. An inhomogeneous structure appears in steel when either incomplete recrystallisation occurs during the interpass times (both the original and recrystallized grains are present in the structure) or a coarsening of the structure occurs after complete, usually static, recrystallisation (SRX). This grain coarsening is usually highly inhomogeneous, where some grains grow faster than others. Previous experience, mainly in cold rolling followed by recrystallisation annealing, shows that grain coarsening after SRX can be significantly enhanced by the application of critical strains (typically between 5 and 15 %). It can be assumed that similar behaviour also occurs in hot rolling.

At Liberty Ostrava a.s. (LO), hot strips are rolled on a unique rolling mill. It is a two-stand Stecklel rolling mill without a roughing mill. For this reason, historically smaller deformations are used in the first passes at this mill. This, combined with the rapid progress of the SRX due to the high temperatures and long interpass times, can lead to coarsening of the grains and thus the formation of inhomogeneities that will ultimately have a negative effect on the ductile-brittle properties of the strip.

This work aimed to map the critical strains for austenitic grain coarsening after SRX in connection with high temperature rolling of microalloyed steel in conditions corresponding to rolling at the Steckel mill in LO.

2. EXPERIMENT DESCRIPTION

The 10 x 15 x 20 mm cubic specimens for the physical simulations were prepared from a hot rolled strip of nominal size 1,250x14.5 mm, of complex microalloyed steel grade L485M/X70M. Anvils with a working width of 5 mm were pressed into the test specimens. A plain strain compression test (PSCT) was performed on the Hydrawedge II hot deformation simulator HDS-20, working with the electrical and hydraulic systems (Gleeble 3800). The reduction of friction between the specimen surface and the tungsten carbide anvil working surface was ensured by the application of a special high temperature lubricant and 0.1 mm thick tantalum plates.

Each sample was resistively heated at a rate of 5 °C/s to a deformation temperature T_d (°C). After holding for 180 s at this temperature, 25% strain was applied (strain rate of 4 s^{-1}). The deformation was followed by an isothermal delay at T_d for Δt (s) and quenching of the sample using water jets. The measured parameters were T_d = 1000-1100 °C and Δt = 1-200 s. The temperature was measured by thermocouples welded to the side surface of the formed sample part. An overview of the samples is given in **Table 1**. The appearance of the sample after the test is shown in **Figure 1**.

Table 1 Testing parameters and marking of individual samples

Δ*t* (s)	T (°C)		
	1 000	**1 050**	**1 100**
1	L1	L7	L13
4	L2	L8	L14
10	L3	L9	L15
50	L4	L10	L16
100	L5	L11	L17
200	L6	L12	L18

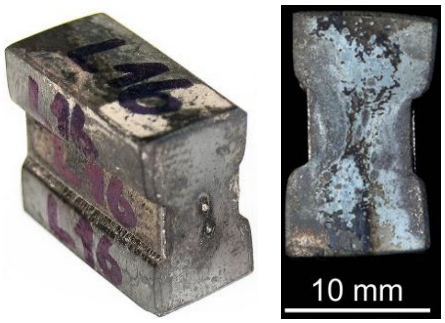

Figure 1 Shape of specimen L16 after deformation, left) overall view, right) detail of the side surface

Simufact Forming software was used for finite element analysis (FEA). Boundary and initial conditions of FEA corresponded to PSCT. The coefficient of friction between the anvil and the sample was set to 0.2. The simulation was performed isothermally. The L485 steel hot flow stress model was taken from the Simufact Forming database. It is clear from the FEA that the equivalent strain is not completely homogeneous over the height or width of the specimen (see the visible "forging" cross (**Figure 2**)). Ignoring the surface regions, the highest equivalent strain is characteristic of the central region of the specimen. As we move horizontally (x-axis) from the specimen centre the equivalent strain decreases with increasing distance from the centre. This phenomenon has been used to reduce the number of specimens. Thus, on a single specimen deformed at a given temperature with a given isothermal delay after pass, it was possible to study regions deformed with different equivalent strains, similar to our earlier work [3]. Micrographs were taken at the location corresponding to the region highlighted in the rectangle in **Figure 3**. The strain embrittlement method was used to visualise the original austenitic grain. A minimum of 12 images per sample were evaluated for grain size, which, for a total of 18 samples, equates to a minimum of 12 x 18 = 216 evaluations.

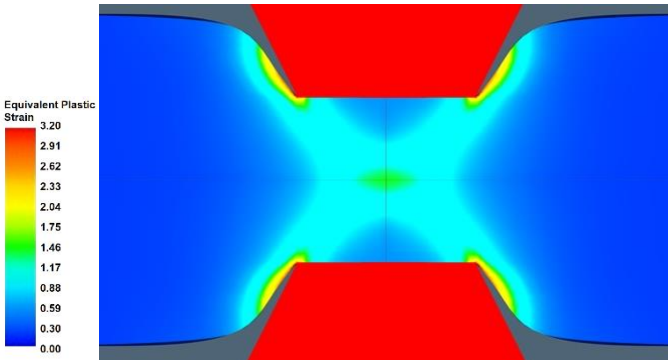

Figure 2 Results of FEA: distribution of equivalent strain

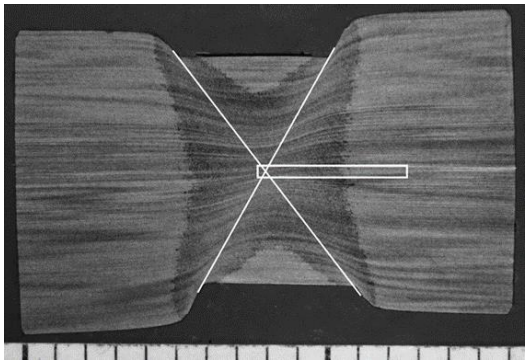

Figure 3 Etched L9 sample

3. RESULTS

An example of the size distribution of the original austenitic grain in the sample after PSCT is shown in **Figure 4**. The austenitic grain size is very inhomogeneous, with a significant coarsening of the grain at 3.6 mm from the centre of the sample. The average grain size here is over 100 μm, with the smallest average grain size in this sample being 29.2 μm. A similar grain size distribution was observed for all samples. At a distance of more than 6.5 mm from the centre, the grain size is not affected by deformation, which allows us to determine the original grain size as follows: 14.3 μm for 1 000°C, 23.4 μm for 1 050°C and 24.3 μm for 1 100°C.

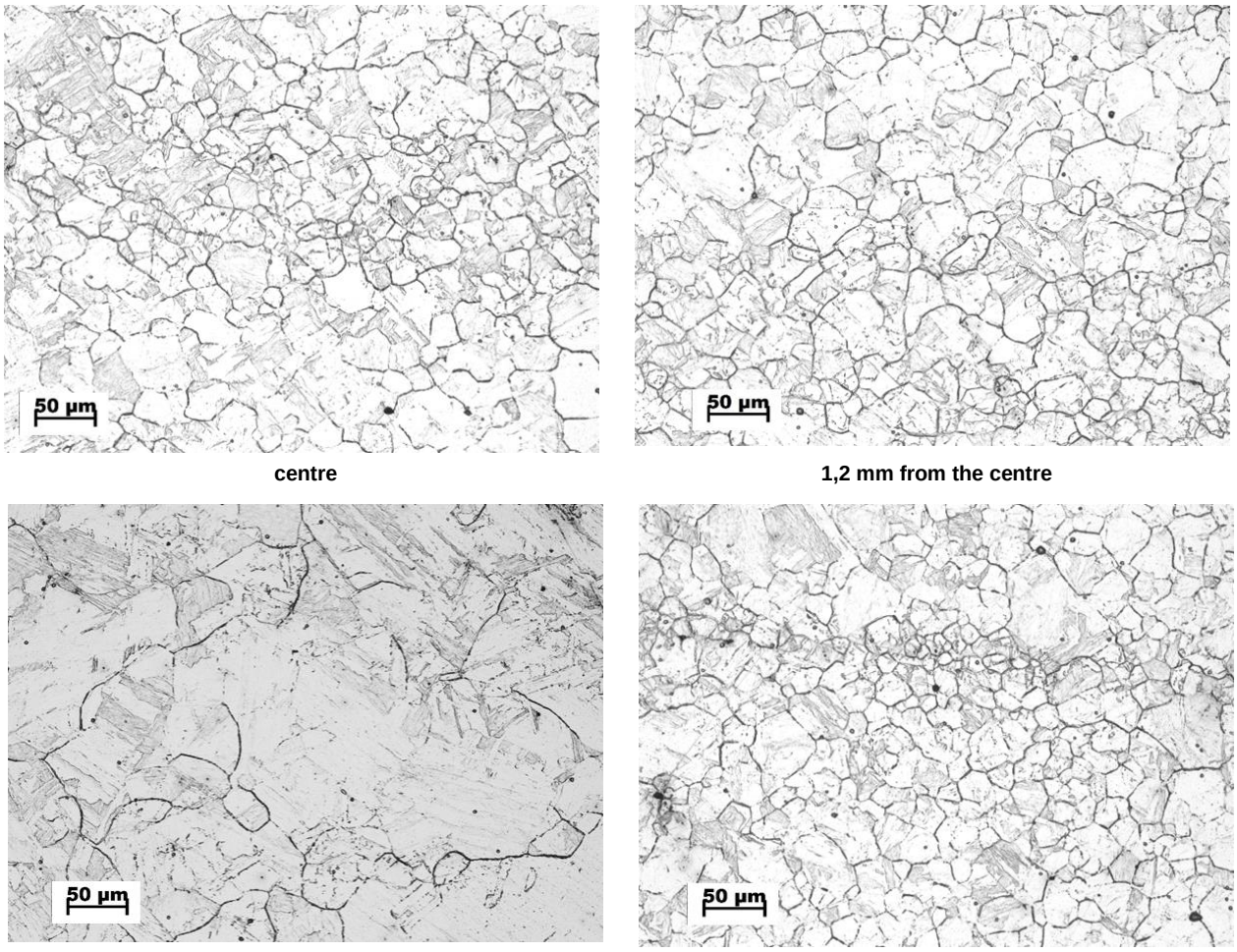

Figure 4 Size of the original austenitic grain at different distances from the centre of the L18 sample

The graphs in **Figure 5** show the evolution of the austenite grain size dependence on equivalent strain over time for all three temperatures tested. At 1 000 °C the average recrystallized grain size is around 15 µm. At a strain of 0.1 to 0.2, there is a progressive and relatively rapid grain coarsening. The maximum after 200 s is reached at a strain of 0.1 when a value of 37.5 µm was measured. At higher strains, no grain growth above 20 µm was observed (with the exception being the values for 4 s, which are out of trend). At 1 050 °C, the average recrystallised grain size is around 25.7 µm. A region of abnormal grain coarsening is seen here at strains between 0.05 and 0.2, where grain sizes above 60 µm were found, which is twice as much as in the case of the highest strains. At a temperature of 1 100 °C the average recrystallized grain size is around 27 µm. Significant coarsening occurs in the strain range from 0.02 to 0.15, up to a value above 160 µm. This is up to 5 times higher than the highest strains and the grains size is significantly inhomogeneous.

4. COMPARISON WITH THE MATHEMATICAL MODEL

A complex model can be used to describe the microstructure evolution during forming, as presented in the work of J.J. Jonas et al [4-10]. The model consists of several modules: 1. critical strain for initiation of dynamic recrystallisation (DRX), 2. grain size evolution (softening between passes through static (SRX) to metadynamic (MDRX) recrystallisation, strain accumulation between passes, recrystallised grain size, mean grain size for incomplete recrystallisation, grain growth (GG) after recrystallisation) and 3. precipitation model

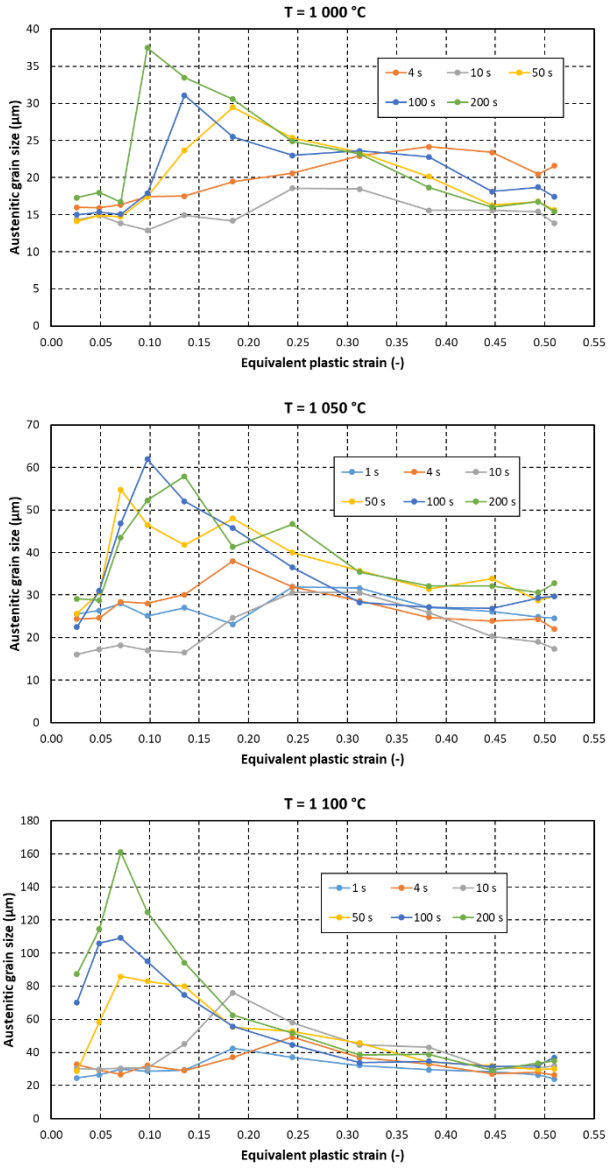

Figure 5 Time dependence of austenite grain size on equivalent strain

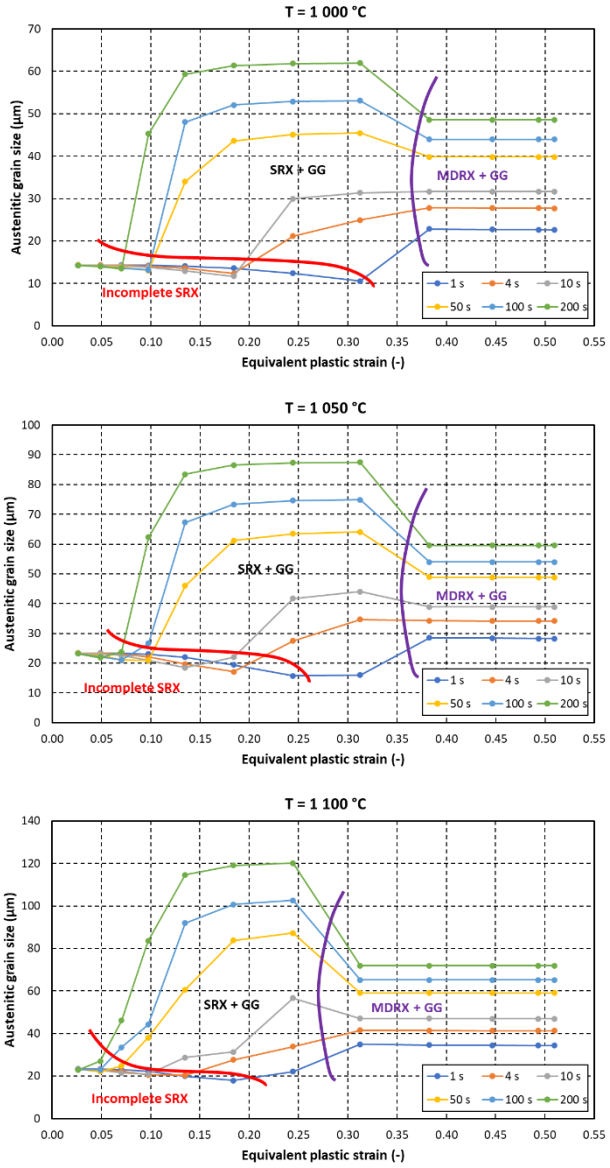

Figure 6 Evolution of austenitic grain size dependence on equivalent strain with time calculated by the model

(solubility of Nb carbonitride, precipitation start time). As our experiment takes place at temperatures where we do not expect precipitation to start, we used the first two models for comparison. For the actual model we used the equations for Nb steel presented by Siciliano e.g., here [8]. Since the equation for grain size determination according to SRX from [4] is not suitable for Nb contents above 0.032 wt. %, we used the following equation [11]:

$$D_{SRX}^{\gamma} = -1.25 + 24.4 \cdot (V + Nb)^{-0.2} \cdot N^{-0.04} \cdot D_0^{0.25} \cdot \varepsilon^{-0.55} \cdot \left[\exp\left(\frac{350\,000}{R \cdot T}\right)\right]^{-0.07} \quad (1)$$

Where:

V, *Nb* and *N* - elemental contents in wt. %
D_0 - initial grain size (µm)
ε - strain (-)
T - temperature (K).

The graphs in **Figure 6** show the evolution of the austenite grain size dependence on equivalent strain over time for all three tested temperatures calculated by the model. By using the model, we can easily identify significant regions in the graph. We can see that for small equivalent strain and short interpass time SRX is incomplete, due to this the grain size remains practically the same as the original one. For large equivalent strain above 0.35 (0.3 for T_d = 1 100 °C), MDRX followed by GG took place in the structure, since after MDRX the grain is finer than after SRX and coarsens more slowly we see a significantly smaller grain here than in the remaining region where SRX and GG occurred. At 1 000 and 1 050 °C the maximum austenitic grain size predicted by the model is significantly larger than our measurements, and conversely at 1 100 °C the model predicts a grain size of only about 120 μm, but we measured up to 160 μm (compare the plots in **Figure 5**). The region of maximum grain size predicted by the model, at all temperatures, is shifted more to the right towards a larger equivalent strain than in the actual samples and is much wider. The model also predicts significant grain coarsening after MDRX, which also does not match the measured results.

5. CONCLUSION

Experiments have shown that when a small equivalent strain is used, especially at higher temperatures and longer interpass times, an enormous grain coarsening can occur. This also affects the microstructure development in subsequent passes, which can result in a coarser final microstructure with an inhomogeneous grain size distribution. This may ultimately lead to a degradation of the ductile-brittle properties. From this perspective, it will be necessary to adjust the roll pass schedule for rolling strips of microalloyed steels at Liberty Ostrava a.s. A comparison of our results with the predictions of a common microstructure evolution model showed that the models originally developed for rolling strips on continuous lines (very short interpass time) are not very applicable to our conditions. Therefore, in the future we will have to develop our models using similar experiments.

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