Influence of Cooling and Strain Rates on the Hot Ductility of High Manganese Steels Within the System Fe–Mn–Al–C

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The ductility curves of high manganese steel grades with a stepwise increasing [Al] content of 0, 1, 3, 5, and 8 wt% for two strain rates (0.01 and 0.001 s⁻¹) and for a varying cooling rate (−3 and −7 K s⁻¹) have been determined to investigate their influence on the hot ductility behavior. The tests have been conducted at the hot tensile testing unit of the Department of Ferrous Metallurgy of RWTH Aachen University. This equipment is able to perform testing of semisolid samples, e.g., during solidification. After the tensile testing, the specimens are investigated using the scanning electron microscope/energy-dispersive X-ray spectroscopy as well as by light optical microscopy to clarify the role of the cooling and strain rates on the hot ductility of high manganese steels and on the formations of precipitates. Furthermore, thermodynamic modeling using the commercial software ThermoCalc is performed. A shifting of the decay of the ductility maximum to lower temperatures down to ≈1273 K for both an increasing strain rate and an increasing cooling rate is determined and this effect is explained through their influence on the microstructure and fracture behavior. Micrograph analyses show that (MnS) precipitates form in contact with early (AlN) precipitates.

1. Introduction

One aspect of the collaborative research center SFB 761 “Steel—ab initio” deals with the investigation of the ductility behavior at high temperatures of high manganese steel grades with varying aluminum contents. The hot ductility behavior of such steel grades is a very important factor for the solidification via the continuous casting (CC) route. High manganese steels exhibit a remarkable combination of ductility (elongation ranging from 55% to 65%) and tensile strength (Rm = 650–1000 MPa) at room temperature.[1] These grades can be classified into two different steel types: the transformation-induced plasticity (TRIP) and twinning-induced plasticity (TWIP) steels. Those steels are designed, e.g., for the automotive industry because of their increased strength-to-weight ratio and their high ductility.[2] Those outstanding mechanical performances lead to the opportunity to implement new ways of designing structural components of a car to fulfill the high safety regulations and to decrease the car body weight.[3]

Within the temperature range between 973 and 1273 K, the hot ductility behavior for regular steel grades is well known. The straightening operation of continuously cast strands induces strains at the surface and below it, especially at oscillation marks and edges, which can result in cracks. Cracks appear primarily at the γ grain boundaries.[4] Using the remelting hot tensile test, the sensitivity to cracking can be investigated in a laboratory scale. The chemical composition, the test temperature, the cooling, and strain rates are main factors influencing and controlling the reduction of area (RA) as a measure of ductility. Hot plasticity behavior of high manganese steel grades distinguishes themselves in several aspects compared to regular steel grades. The ductility versus temperature curve for regular steel grades reveals two characteristic ductility minima.[5] Located directly below the solidus temperature, the first minimum is caused by microsegregation, which results in the formation of liquid films within the interdendritic zone. The formation of sulfides, nitrides, carbides, and thin proeutectoid ferritic films at the austenite boundaries leads to the formation of the second minimum (973–1273 K). Between the first and second minimum, the RA values can reach up to 100%. In contrast, for high manganese steels, only the first minimum is observed and the values of the RA barely exceed 50 pct.[6] To avoid cracking during the CC process, RA values higher than 40% should be reached.[7] However, as established in the literature, several tested high manganese steel grades do not exceed this critical RA value.[8–11]

The single-phase austenitic structure is one of the reasons for the low ductility values of the TWIP steels. The RA values of steels containing ferrite can reach RA values up to 80–100%, whereas the values in a fully austenitic structure are between 40% and 50%.[13] Caused by the absence of the thin ferritic film...
in single-phase steels, the ductility is mainly influenced by precipitates and second phase particles. (MnS) particles formed at high temperatures during cooling can act as nuclei and assist the formation of complex precipitates.\textsuperscript{[10,14]} Furthermore, due to the high [Al] contents, aluminum nitrides can occur at the grain boundaries. To minimize the formation of second phase particles and to improve ductility, low [S] and [N] levels are required.\textsuperscript{[8]} The transformation temperature is increased by high [Al] contents in conventional TRIP steels, and therefore, large amounts of austenite might form before the straightening in the CC process and lead to a ductility improvement. In case of 1 wt% [Al] conventional TRIP steels, the ductility at the critical temperature range can be very poor, ascribable to the presence of a thin ferritic film. Due to the coarse shape of the precipitates, a smaller influence of the formed (AlN) has been observed.\textsuperscript{[12]}

In addition to the chemical composition, the cooling and strain rates are also reported to influence the hot ductility behavior of steels. Precipitate sizes and volume fractions can be affected by both parameters.\textsuperscript{[15–17]} There is no general rule for the influence of the cooling rate on the hot ductility of steels, and its effect varies case-by-case. An increase in the cooling rate leads to a shift in the ductility curve towards lower temperatures and a widening of the low ductility zone. A decrease in the cooling rate is reported to improve ductility. One reason for this increase in ductility at lower cooling rates is the coarsening of precipitates, whereas a higher cooling rate results in finer grain size and precipitates formation. Moreover, the formation of microsegregation is reduced with a higher cooling rate.\textsuperscript{[8,11,13,17–19]}

In the presence of certain precipitate formers, the opposite effect can be observed as follows: with increasing cooling rate, there is no enough time for (MnS) to precipitate and form in combination with [Al] and [N] complex precipitates, which would result in a loss of ductility.\textsuperscript{[14,19]}

At the third ductility minimum (973–1173 K), a higher influence of the strain rate on the ductility is observed as opposed to temperatures above 1173 K (2nd ductility minimum). The lower temperatures combined with a decreasing strain rate widen the ductility trough for [C] steel (0.6C 0.27Si 1.6Mn). Corresponding to an increasing cooling rate, a declining strain rate is reported to result in a wider and deeper ductility trough.\textsuperscript{[13,20,21]} A lower strain rate results in a longer deformation time, and thus, precipitates can grow to larger nitrides and sulfides.\textsuperscript{[14,15]} In the presence of sulfide formers, by increasing the strain rate, a reduction of ductility may occur due to the formation of sulfides. An opposite effect is observed in the case of nitrides, where formation is inhibited by an increase in the strain rate.\textsuperscript{[22]}

However, the strain rate interval in which the embrittlement due to sulfides is observed is ~four times the size of the interval where embrittlement due to carbides can occur. The cause for those differences is reasoned in the different precipitation kinetics for (Fe,Mn)S, carbides, and nitrides.\textsuperscript{[22]}

Steenken et al. already conducted research on the influence of a varying aluminum content on the hot ductility of high manganese steels. The results are shown in Figure 1.\textsuperscript{[23]}

To gain a deeper insight into the previously less researched subject of the hot ductility behavior of high manganese and aluminum steels and their corresponding castability via the CC route, it is important to investigate the influence of [Al] for single- and multi-phase steels as well as the influence of different cooling rates and strain rates on the hot ductility. Therefore, five grades with varying [Al] contents (0, 1, 3, 5, and 8 wt% [Al]) were cast and analyzed focusing on the hot ductility behavior. To investigate the influence of the cooling and strain rates, those steel grades were cooled and strained to failure with varying parameters. The test temperatures were selected using characteristic temperatures of each hot ductility curve shown in Figure 1. To detect the responsible failure mechanisms, different analyses such as fractography, phase fraction, and microstructure analyses are conducted and compared to equilibrium state (ThermoCalc) calculations.

2. Experimental Section

2.1. Material Sampling and Steel Composition

At the IEHK Vacuum Steel Center of the Department of Ferrous Metallurgy RWTH Aachen University, five different steel grades were cast. Experiments were conducted using a two-chamber vacuum induction furnace with a heat weight of 100 kg to cast 140 mm square ingots. Specific tensile specimens with a diameter of 20 mm and length of 130 mm were produced from those ingots.

The composition of the investigated steel grades (in wt%) is shown in Table 1. All specimens had a manganese content of nominal 17 wt%, a carbon content of 0.3 wt%, and a varying aluminum content from 0 to 8 wt%.

![Figure 1. Hot ductility curves of the tested steel grades with a cooling rate of ~ 3 K s\(^{-1}\) and a strain rate of \(\dot{\varepsilon} = 0.001\) s\(^{-1}\), based on Steenken et al.]()
2.2. Hot Tensile Testing

The specimens were tested at the Department of Ferrous Metallurgy RWTH Aachen University using a CC simulator with a melting/solidification unit (in situ testing). Different testing conditions were established by varying the strain rate while keeping the cooling rate constant and vice versa to evaluate the hot ductilities. The specimen was heated up above its corresponding liquidus temperature, and after holding the temperature ($\approx T_{\text{liq}} + 20 \, \text{K}$) for about three minutes to homogenize the microstructure, the influence of the cooling and strain rates on the hot ductility was tested. A constant cooling rate ($\sim 7$ or $\sim 3 \, \text{K s}^{-1}$) was applied to the specimen and then strained to rupture with a preset strain rate (0.01 or 0.001 s$^{-1}$). The cooling rate was close to the cooling rate for conventional CC machines near the strand surface ($60–100 \, \text{K min}^{-1}$) and thin slab casting machines ($200–300 \, \text{K min}^{-1}$). To avoid oxidation, the experimental chamber was flooded with argon. The temperature course is schematically shown in Figure 2.

After separation, the area of fracture was measured and the RA was calculated by using Equation (1)

$$\text{RA} = \left( \frac{A_0 - A_f}{A_0} \right) \times 100\% \quad (1)$$

In Equation (1), $A_0$ is representing the cross section before testing and $A_f$ the cross section after testing. The different parameters for the experiments are summarized in Table 2.

Various investigation methods were used to analyze the type and cause of fracture, to detect precipitates, and to obtain determining factors influencing the hot ductility. The fracture surface itself was investigated using the scanning electron microscope (SEM), energy-dispersive X-ray spectroscopy (EDX), and wavelength-dispersive X-ray spectroscopy (WDX). To measure the grain size of the as-cast structure and the secondary arm spacing (SDAS), the area under the fracture surface was examined using SEM and light microscopy. The grain sizes were measured using the linear intercept method (DIN EN ISO 643). Furthermore, calculations and simulations under equilibrium solidification conditions were conducted by using the commercial software ThermoCalc with database TCFE7.

3. Results

3.1. Hot Tensile Testing

In Figure 3, the results of the hot ductility testing with varying cooling rates of the steel grades X30Mn17 to X30Mn17-Al8 are shown. For all investigated steel grades, the ductility increases with increasing temperature at a cooling rate of $\sim 3 \, \text{K s}^{-1}$. X30Mn17 and X30Mn17-Al8 reveal a different behavior in the results of hot ductility testing for different cooling rates at high temperatures. In case of X30Mn17, the difference between the two cooling rates at 1623 K is 51% and 41% for X30Mn17-Al8 steel at 1573 K. Almost no difference between the curves can be observed for steel grade X30Mn17-Al1 with varying cooling rates. At 1373 K, the steel grades X30Mn17-Al3 and X30Mn17-Al5 show a higher RA for the greater cooling rate of $\sim 7 \, \text{K s}^{-1}$. The contrary can be observed at higher temperatures (1573 K) with the lower cooling rate.

In Figure 4, the hot ductility results for varying strain rates of $\dot{\varepsilon} = 0.01 \, \text{s}^{-1}$ and $\dot{\varepsilon} = 0.001 \, \text{s}^{-1}$ are presented. At temperatures of about 1423 K, the ductility for both strain rates is approximately equal. With increasing temperature, the ductility curves start to separate. Higher temperatures (about 1573 K) combined with the smaller strain rate of $\dot{\varepsilon} = 0.001 \, \text{s}^{-1}$ result in at least 30% better hot ductility than at a strain rate of $\dot{\varepsilon} = 0.01 \, \text{s}^{-1}$. At high temperatures, the proximity to the solidus temperature is decisive for the influence of the strain rate on hot ductility. With a slower strain rate, more time is available for the solidification processes.

### Table 1. Chemical compositions of the laboratory ingots in weight percent.

| Grade     | [C] | [Mn] | [Al] | [Fe] | [Si] | [S]  | [P]  | [N]  | [Ti] | [V]  | [H]  | [O]  |
|-----------|-----|------|------|------|------|------|------|------|------|------|------|------|
| X30Mn17   | 0.328 | 17.3 | <0.01 | 82.372 | 0.08 | 0.012 | 0.003 | 0.0048 | <0.001 | <0.001 | 0.0001 | 0.0026 |
| X30Mn17-Al1 | 0.309 | 17.6 | 0.9  | 81.171 | 0.03 | 0.011 | 0.003 | 0.0075 | <0.001 | <0.001 | 0.00073 | 0.0003 |
| X30Mn17-Al3 | 0.329 | 16.9 | 3.2  | 79.571 | 0.02 | 0.01  | 0.001 | 0.0032 | <0.001 | <0.001 | 0.00072 | 0.0003 |
| X30Mn17-Al5 | 0.346 | 17.6 | 5.4  | 76.654 | 0.02 | 0.009 | 0.001 | 0.0032 | <0.001 | <0.001 | 0.00059 | 0.0003 |
| X30Mn17-Al8 | 0.37  | 17.5 | 8.5  | 73.630 | 0.02 | 0.012 | 0.001 | 0.0020 | <0.001 | <0.001 | 0.00055 | 0.0003 |

### Table 2. Summary of parameters which were varied within the hot ductility experiments.

| Test condition | Strain rate [$\text{s}^{-1}$] | Cooling rate [$\text{K s}^{-1}$] |
|---------------|-------------------------------|-------------------------------|
| 1             | 0.001                         | –3                            |
| 2             | 0.001                         | –7                            |
| 3             | 0.010                         | –3                            |
Furthermore, previous research showed higher strain rates to induce smaller precipitates at the grain sizes which can diminish hot ductility values.\textsuperscript{15,21,25} The hot ductility of the steel grade X30Mn17-Al1 at low temperature (1173 K) and high temperature (1373 K) is nearly equal for both strain rates. At 1273 K, the hot ductility at a strain rate of $\dot{\varepsilon} = 0.01 \text{s}^{-1}$ is higher, showing a RA value of 51%, which is 26.5% higher than for the lower strain rate.

3.2. Fractography

To detect the responsible mechanism of ductility losses, SEM and EDX analyses have been conducted.

\textbf{Figure 5a} shows the fracture surface of steel X30Mn17 tested at 1423 K. A fine (<1 $\mu$m) structure with similarity to precipitates is found. Those suspected precipitates are too small (<1 $\mu$m) to be analyzed by the used EDX. Figure 5b shows the fracture surface of X30Mn17-Al1. Two different surface areas can be observed. On the left side of the figure, some dimples (A) and a dendritic structure in the upper right corner (B) are pictured. The size of the dimples varies from small (about 10 $\mu$m) to big (about 180–200 $\mu$m). Inside some of the dimples, precipitates can be found. Presumably in most of them, precipitates should be expected, but due to specimen preparation, some precipitates have been removed. Figure 5c also shows the fracture surface of X30Mn17-Al1, illustrating the mentioned precipitates inside the dimples, however, at other testing conditions.

In \textbf{Figure 6a}, different fracture surfaces of the X30Mn17-Al3 steel can be seen. Near to the edge of the specimen a dendritic

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{fracture_surfaces.png}
\caption{Hot ductility course of X30Mn17 to X30Mn17-Al8 for different cooling rates ($dT/dt = -3 \text{ K s}^{-1}$ and $dT/dt = -7 \text{ K s}^{-1}$) and constant strain rate ($\dot{\varepsilon} = 0.001 \text{s}^{-1}$).}
\end{figure}
Figure 4. Hot ductility course of X30Mn17 to X30Mn17-Al8 for two different strain rates ($\dot{\varepsilon} = 0.001 \text{s}^{-1}$ and $\dot{\varepsilon} = 0.01 \text{s}^{-1}$) at constant cooling rate ($dT/dt = -3 \text{K s}^{-1}$).

Figure 5. Fractography investigations: a) X30Mn17, $V = 2000 \times$, $T_{\text{test}} = 1423 \text{K}$ cooling rate $T = -3 \text{K s}^{-1}$, and strain rate $\dot{\varepsilon} = 0.01 \text{s}^{-1}$; b) X30Mn17-Al1, $V = 50 \times$ at $T_{\text{test}} = 1373 \text{K}$, cooling rate $T = -3 \text{K s}^{-1}$, strain rate $\dot{\varepsilon} = 0.01 \text{s}^{-1}$, dimples (A), and dendritic structure (B); and c) X30Mn17-Al1, $V = 400 \times$ at $T_{\text{test}} = 1173 \text{K}$, cooling rate $T = -7 \text{K s}^{-1}$, and strain rate $\dot{\varepsilon} = 0.001 \text{s}^{-1}$.
structure (B) can be observed. Further into the middle of the sample, dimples are present (C). Figure 6b shows the fracture surface of the X30Mn17-Al5 steel. In Figure 6b, a structure consisting of dimples can also be observed. In all specimens, Mn(S,Se) and (AlN) precipitates can be observed. In Figure 6c, one of those precipitates is exemplary pictured. Precipitates are observed at grain boundaries as well as on the top of the dendritic structure. Most of the precipitates have a round or panel-like shape and a size between 2 and 10 μm. Equally in most of the investigated fracture surfaces, shrinkage holes can be seen, which might result out of shrinkage during solidification. Most of those shrinkage holes can be detected at the dendritic surfaces and close to surfaces containing dimples. In the vast majority of the investigated specimens, at least two or three different types of fracture surfaces are present. Mostly at the edge of the sample, a dendritic structure can be noted, further to the middle some interdendritic and intergranular fracture surfaces as well as dimples can be perceived. The size of the dimples varies from 20 to 200 μm, but their sizes are mostly below 100 μm. A relation between the size of the dimples and the ductility cannot be detected for these investigated steel grades. A sample with a value for RA of 71 pct. (0 wt% [Al], dT/dt = −3 K s⁻¹, 1623 K) does not show any dimple structure, whereas another sample for a different cooling rate and a lower RA value of 12 pct. (8 wt% [Al], dT/dt = −7 K s⁻¹, 1523 K) does not show any dimples.

### 3.3. Microstructure

To gain further information about the formation of precipitates, additional SEM analyses and ThermoCalc calculations were conducted. Figure 7 shows a micrograph of a X30Mn17-Al5 specimen and pictures the two matrix phases, ferrite and austenite, as well as the homogeneous distribution of the precipitates. EDX analyses show an enrichment of [Al] in the high-relief areas, which indicate a ferrite phase. In the background, an enrichment of [Mn] can be detected which is indicating the austenite phase. ThermoCalc calculations for all investigated steel grades were conducted and revealed for steel grades containing less than 3 wt% [Al], a fully austenitic solidification and for steel grades containing more than 3 wt%, a two-phase solidification.

In Figure 8, the formation of precipitates for the X30Mn17-Al8 steel is investigated. It shows a microscope image (magnification of 1000x) of a combined (AlN) and (MnS) precipitate (Figure 8a). It reveals that the (MnS) and the (AlN) precipitates are forming together, which is confirmed by WDX analysis (Figure 8b–e). The (MnS) is forming around the (AlN) precipitates, therefore (AlN) is reducing the necessary activation energy for the formation of a (MnS) precipitate and acts therefore as a substrate to form complex precipitates.

Considering the equilibrium state calculation of precipitates formation (Figure 9) using ThermoCalc for this specific alloy, it can be seen that (AlN) is forming first, at higher temperatures (1817 K), later the (MnS) (1616 K). Therefore, (AlN) can serve as a substrate to form complex precipitates. In addition to this, with increasing [Al] content, the aluminum activity is increasing and more (AlN) will be formed. For X30Mn17-Al1, the (AlN) and (MnS) precipitates might form at the same time (Figure 9b).

Using the light microscope, grain sizes and SDASs of the investigated steel grades at different testing conditions as well as the size and number of precipitates were measured. Table 3 shows the size and number of precipitates with varying aluminum content. The number of precipitates per examined area is decreasing with increasing aluminum content, yet the size of the precipitates is increasing, so the area fraction of the precipitates is almost constant. X30Mn17-Al1 reveals the highest
number of precipitates. This can be explained by the high amount of [N] (75 ppm at 1 wt% [Al] and 20–32 ppm for the other steel grades). The grain size was measured according to the linear intercept method (DIN EN ISE 643).

To reveal the grains of the as-cast structure, the samples were etched using a HNO$_3$-based reagent. It acted by both revealing the grain boundaries themselves and by giving slightly different colors to different crystal orientations, depending on the specific region and alloy. Grain boundaries were reinforced by using an image editing software, just before the measurement using the DIN standard. Secondary grains revealed by this methodology often coincided with the original dendritic networks but grain boundaries also shifted sometimes. Depending on the alloy, grain boundaries may have shifted due to local stresses, phase transformation processes, or grain boundary migration.

The grain size at high temperatures (1473–1573 K) is decreasing with an increasing aluminum content (Table 3). In addition to the aluminum content, the deformation temperature also seems to influence the grain size. Examined steel grades containing less than 5 wt% aluminum exhibit smaller grain sizes at lower testing temperatures (973 K) than those for higher temperatures. The grain growth is faster at higher temperatures, thus the longer the sample stays at high temperatures (1473–1573 K), the more time for grain growth is available. For examined steel grades containing 5 and 8 wt% aluminum, no differences in grain size between high and low temperatures have been

![Figure 8](https://www.advancedsciencenews.com/...)

**Figure 8.** X30Mn17-Al8, $V = 1000$ at $T_{\text{test}} = 1623$ K cooling rate $\dot{t} = -3$ K s$^{-1}$ and strain rate $\dot{\varepsilon} = 0.001$ s$^{-1}$. a) Micrograph: (AlN) and (MnS) precipitates and (MnS) precipitates around (AlN) precipitates. (b–e) Element mapping elements images of [Al], [N], [Mn], and [S].

![Figure 9](https://www.advancedsciencenews.com/...)

**Figure 9.** Precipitation simulation using ThermoCalc calculation with database TCFE7 for a) X30Mn17-Al8 and b) X30Mn17-Al1.

| Table 3. Grain size, number of precipitates, and the size of precipitates for the tested steel grades with varying aluminum content. |
|---------------------------------------------------------------|
|                     Average grain          | Average grain          | Number of precipitates | Average of precipitates, area at 1073 K |
| steel size at 1473–1573 K | steel size at 973 K | at 1073 K              |                                      |
|--------------------------|--------------------|------------------------|--------------------------|
| X30Mn17                  | 865 μm             | 642 μm                 | 180                      | 18.33 μm$^2$          |
| X30Mn17-Al1              | 755 μm             | 490 μm                 | 242                      | 19.37 μm$^2$          |
| X30Mn17-Al3              | 556 μm             | 244 μm                 | 120                      | 16 μm$^2$             |
| X30Mn17-Al5              | 117 μm             | 152 μm                 | 90                       | 17.21 μm$^2$          |
| X30Mn17-Al8              | 38 μm              | 41 μm                  | 69                       | 21.04 μm$^2$          |

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detected. Their interfacial energy seems to be the main influence on the grain size, see the Discussion section.

SDAS average varies from 37 to 73 μm. The different test temperatures, strain, or cooling rates do not influence the SDAS in any way for the investigated steel grades.

4. Discussion

Regarding to the results of hot ductility testing, a shifting of the ductility trough can be observed. An increase in the strain rate results in a shift of the hot ductility maximum to lower temperatures, and the overall ductility decreases (Figure 10a,b).

At a strain rate of 0.001 s⁻¹, ductility values go up to 84% for the X30Mn17-Al5 steel, whereas at a strain rate of 0.01 s⁻¹, the hot ductility only increases to 56%. Previous research explained those effects by a finer precipitates formation at the grain boundaries. For the investigated steel grades, the size or composition of the precipitate is not influenced by the different strain rates. At both strain rates, (MnS) as well as (AlN) precipitates are found. The range of the tested strain rates is probably too small to find a significant influence of the strain rate on the precipitates composition and formation. The investigated solubility product [Al]x[N] > 2 x 10⁻⁴ issued by Mintz et al. for the formation of (AlN) precipitates in high manganese steels (>1.4% Mn) is reached for every investigated steel grade. With increasing [Al] content, the ferrite phase fraction is increasing, beginning at an [Al] content of 3 wt%. Therefore, the ductility of ferrite is lower than in austenite, and this is one additional factor decreasing the ductility in addition to the increasing strain rate and explains the low RA values of 15.8% of X30Mn17-Al8 at a strain rate of 0.01 s⁻¹. These low ductility values for the two-phase alloy may also be caused by the different specific volumes of the δ/γ phases. Examined high manganese alloys containing more than 3 wt% [Al] show a two-phase solidification. During the δ to γ transformation occurring during solidification, stress can be induced.

In addition to the strain rate, the cooling rate has an influence on the hot ductility behavior of the tested steel grades. The influence of the cooling rate on hot ductility is similar to the influence of the strain rate. As well as with increasing the strain rate, with increasing cooling rate, a shift of the ductility maximum to lower temperatures can be observed (Figure 10c,d). At a cooling rate of -3 K s⁻¹, a maximum RA value of 84% is reached.

Figure 10. Hot ductility course for all steel grades, a) strain rate 0.001 s⁻¹ and © 3 K s⁻¹, b) strain rate 0.01 s⁻¹ and © 3 K s⁻¹, c) cooling rate © 3 K s⁻¹ and strain rate 0.001 s⁻¹, and d) cooling rate © 7 K s⁻¹ and strain rate © 0.001 s⁻¹.
(X30Mn17-Al5), whereas \(-7 \text{ K s}^{-1}\) results in a maximum value of 60%. According to Schwerdtfeger, Mintz, and Kang, a better ductility due to slower cooling rate can be explained with a coarsening of the precipitation. The conducted SEM analyses show no influence of an increasing cooling rate on the precipitates formation. At high temperatures, the ductility at \(-7 \text{ K s}^{-1}\) decreases significantly for a steel containing 0 and 8 wt% aluminum relatively to their respective values for \(dT/dt = -3 \text{ K s}^{-1}\), in comparison with the other investigated steel compositions. This can be reasoned by a higher cooling rate raising the tendency for segregation for X30Mn17. For X30Mn17-Al8, the previously explained \(\delta/\gamma\) transformation might even have a greater impact, due to a faster solidification, more metastable \(\delta\)-ferrite will be retained in the system. This phase is unstable, and when it transforms to \(\gamma\) in the absence of a liquid phase, larger regions where plastic deformation locally takes place will appear. A higher amount of \(\delta\) will also lead to a higher amount of microsegregation causing less N und S to solute. Therefore, it is necessary to distinguish between two different mechanisms of precipitation formation. One is due to microsegregation (coarser) and the other due to phase transformation (finer).

Following results can be obtained by investigating the fracture surface using the SEM: fracture surfaces with low ductility show intradendritic and transdendritic fractures. The fracture surface of a ductile specimen is showing deeper dimples and stronger deformation, than a less ductile specimen. At brittle fracture surfaces, intercrystalline fractures can be observed, because of precipitates formation and low temperatures \((<1273 \text{ K})\). In addition to this, at most of the surfaces, different types of fracture can be detected, as well as different kinds of precipitates have been perceived. The differences in grain size might be explained by the formation of ferrite. Jacobi et al. found that the interfacial energy of ferrite is smaller than of austenite: \(\sigma(L)\delta < \sigma(L)\gamma\). This results in a greater formation of ferrite grains that act as substrates for austenite nucleation, and therefore, in smaller primary and secondary grain sizes for \([Al] > 3 \text{ wt\%}\). These differences in grain size can be related to the formation of a single phase for less than \(3 \text{ wt\%} [Al]\) and two phases for more than \(3 \text{ wt\%} [Al]\), as previously discussed. With increasing aluminum content, the equilibrium phase fraction of ferrite is increasing. Even if there is no ferrite present at room temperature according to the ThermoCalc calculations, there might be some metastable \(\delta\)-ferrite left from solidification, in the case of a peritectic solidification.

Through the micrographs, it is shown that aluminum nitrides as well as manganese sulfides are the precipitates that have been detected. ThermoCalc calculations confirm that those precipitates are expected. Some aluminum oxides have been observed too, but are not considered furthermore, because their occurrence is due to inclusions coming from contact with the \((\text{Al}_2\text{O}_3)\) crucible of the hot tensile testing machine. At some locations, it can be seen that the \((\text{MnS})\) is forming around the \((\text{AlN})\) (Figure 8). \((\text{AlN})\) is reducing as a substrate the necessary activation energy so that more complex precipitates consisting of \((\text{AlN})\) and \((\text{MnS})\) can form. A suppression of precipitates formation due to an increasing cooling rate cannot be observed for the investigated steel grades and for the investigated cooling rates.

5. Conclusion

Within this study, different experimental and theoretical investigations about the hot ductility behavior of high manganese steels with a varying aluminum content \((0, 1, 3, 5, \text{ and } 8 \text{ wt\% [Al]})\) have been conducted. The main results can be summarized as follows: 1) Increasing the cooling and strain rates results in a shifting of the hot ductility curves to lower temperatures. Furthermore, the overall ductility is decreasing with an increasing strain rate. 2) An increase in the cooling rate for X30Mn17 and X30Mn17-Al8 leads to a decrease in ductility which can be explained by the segregation for the 0 wt% [Al] alloy and the \(\delta/\gamma\) transformation-induced stress for the 8 wt% [Al] alloy. Lower temperatures combined with higher strain rates of 0.01 s\(^{-1}\) lead to better ductility, whereas at higher temperatures, ductility is better at the lower strain rate of 0.001 s\(^{-1}\). At lower temperatures, the influence of precipitates is dominating, whereas the matrix influence dominates at higher temperatures. 3) SEM analysis states that fracture surfaces with low ductility show intradendritic and transdendritic fractures. Fracture surfaces with high ductility show deep dimples and high deformation. At brittle fracture surfaces, intercrystalline fractures can be found, because of precipitates formation and a consequence of low temperatures \((<1273 \text{ K})\). 4) The grain size is decreasing with increasing

![Figure 11. ThermoCalc calculations for precipitates of a) X30Mn17, b) X30Mn17-Al3, and c) X30Mn17-Al8.](image-url)
aluminum content, as well as the number of precipitates. However, the size of the precipitates is increasing with increasing aluminum content, so that the fraction of the surface occupied by the precipitates is staying approximately constant. 5) Precipitates are found either as pure (AlN) or (MnS) or as complex combined precipitates. Aluminum nitrides serve as a substrate for (MnS) precipitation at relatively high [Al] concentration. (MnS) is forming around the (AlN), because (AlN) is reducing the necessary activation energy for (MnS) nucleation.

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Conflict of Interest
The authors declare no conflict of interest.

Keywords
cooling rate, high manganese steel, high temperature ductility, precipitates, strain rate

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