

On the bainitic and martensitic phase transformation behavior and the mechanical properties of low alloy 51CrV4 steel

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Abstract

The role of the austenitization treatment (austenitizing temperature and austenitizing time) on the subsequent phase transformation behavior of low alloy 51CrV4 steel was investigated. The results indicate that the prior austenitization treatment strongly affects both the isothermal bainitic phase transformation kinetics and the onset of the martensitic transformation. In fact, higher martensite start temperatures (M_s) and longer phase transformation times were present when the austenite grain size increased concomitant with austenitization temperature. In addition, when the supercooled austenite was pre-strained and the subsequent transformation proceeded under a superimposed stress, transformation plasticity strains evolved, whereas higher end values at room temperature were obtained at a lower pre-deformation temperature (T^*), which is usually not incorporated in current models. Moreover, tensile tests performed at room temperature following pure bainitic, pure martensitic, or bainitic-martensitic reactions revealed that no linear mixture rule is present for the mixed microstructures. Overall, the current findings demonstrate the necessity of proper incorporation of T^* and a non-linear mixture rule for bainitic-martensitic microstructures in low alloy 51CrV4 steel while modeling complex manufacturing processes.

Keywords:

Phase transformation; martensite; bainite; supercooled austenite; pre-deformation temperature; mechanical properties.

Introduction

The reasons for the development of new production processes are manifold. Mostly the reduction of time, costs and energy is requested in order to be more competitive in today's markets. One promising way in this context is the development of production processes suitable to manufacturing work pieces with functionally graded properties. The functional gradation, i.e. the well-directed adjustment of local properties to local loads, can be achieved either by the combination of different materials [Praveen and Reddy 1998] or by using an optimized local temperature-time-deformation path during processing of a monomaterial (steel, aluminium or plastic) [Steinhoff et al. 2009]. Especially the monomaterial steel offers the possibility to adjust the mechanical properties in a large range, which is already used as an advantage in industrial press hardening of B-pillars [Steinhoff et al. 2009b]. By using optimized heating and/or cooling strategies prior to and during pressing, B-pillars with functionally graded properties are producible with local tensile strengths ranging from 600 to 1500 MPa [Paar et al. 2008]. However, the advantages of functional gradation are not restricted to sheet forming processes: a newly developed massive forging process showed potential for functional gradation [Steinhoff et al. 2005]. Certainly, the required three-dimensional distribution of desired properties complicates the production process immensely, but at the same time, it offers a great opportunity to reduce the process chain drastically, since no further heat treatment and less reconditioning by turning or milling would be required. However, especially for steel products, phase transformations have to be considered since they are accompanied by volumetric changes which might lead to distortion. Thus, for the optimization of the local temperature-time-deformation path a deep knowledge about how the different parameters, such as temperature and stress or strain, affect the phase transformation behavior, the mechanical properties and the dimensional stability, is indispensable. In particular, the austenitization treatment has a strong influence on the subsequent phase transformation kinetics since it dictates the degree of carbide dissolution and the austenite grain size. Especially when residual carbides are present the amount of carbon and chromium in the austenite is lower than in the bulk. Due to the lower carbon content, M_s increases since the amount of carbon dissolved in the austenitic matrix rather than the average carbon value in the work

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piece determines the transformation start temperature [Rose and Rademacher 1957]. In addition, temperature gradients during processing result in thermal stresses, and it is commonly known that stresses superimposed during the phase transformation accelerate the phase transformation kinetics, in the case of isothermal transformations [Ahrens et al. 2002, Shipway and Bhadeshia 1995, Bhadeshia et al. 1991] or lead to an increase in transformation start temperature when the transformation occurs continuously [Bhadeshia 1996, Denis et al. 1985, Patel and Cohen 1953]. Similar effects were observed following pre-straining of supercooled austenite prior to the isothermal bainitic transformation [Lambers et al. 2009, Ahrens et al. 2004]. However, when the pre-straining level exceeds a certain limit acceleration of the transformation kinetics is observed no longer, since the pre-straining stabilizes the austenitic matrix [Larn and Yang 2000, Shipway and Bhadeshia 1997].

Moreover, transformation plasticity strains, *i.e.* the anisotropic portion of the transformation strains, evolve when stresses even lower than the yield stress of the supercooled austenite are superimposed during the phase transformation [Su et al. 2003, Taleb et al. 2001, Ahrens et al. 2000]. Two mechanisms are widely accepted in order to explain the occurrence of transformation plasticity strains, the Magee mechanism [Magee 1966] and the Greenwood-Johnson mechanism [Greenwood and Johnson 1965]. According to the Greenwood-Johnson mechanism transformation plasticity strains stem from micro-deformations of the weaker phase. Since there is a volume difference between the parent and the product phase internal stresses evolve, which might be superimposed by additional externally applied stresses during the transformation. As a result, stress levels higher than the yield stress of the supercooled austenite are present, resulting in a plastic deformation of the supercooled austenite. Thus, the growth of the product phase at the expense of the weaker parent phase is accompanied by a plastic deformation along a certain direction, yielding transformation plasticity strains [Taleb and Petit 2006, Greenwood and Johnson 1965].

The Magee approach explains the evolving transformation plasticity strains with the alignment of martensitic or bainitic variants in the presence of superimposed stress fields. The additional mechanical driving force leads to variant selection, such that variants having a preferred orientation with respect to the local stress field directions grow at the expense of others. Finally, this brings about the anisotropic transformation plasticity strains [Kundu et al. 2007, Bhadeshia 1996, Magee 1965]. Both mechanisms can be simultaneously active; however, the predominating mechanism depends on the transformation temperature, the superimposed stress level and the type of transformation [Dalgic et al. 2008, Taleb and Sidoroff 2003, Marketz et al. 1996].

In recent years several models were proposed utilizing either only one or a combination of both mechanisms [Wolff et al. 2005, Taleb and Sidoroff 2003, Fischer et al. 2000, Marketz and Fischer 1994, Leblond et al. 1989], successfully simulating the occurring transformation plasticity strains for superimposed stress levels lower than half times the conventional 0.2 % offset- yield stress of the supercooled austenite [Taleb et al. 2001]. When the applied stress level exceeds this value the nonlinearity of the transformation plasticity strain vs. superimposed stress curve has to be considered.

However during forging, not only high stresses but also plastic strains are present further complicating the modelling of the occurring transformation plasticity strains. Moreover, in several previous studies a fully austenitic microstructure was adjusted prior to the subsequent phase transformation for monitoring the effect of stresses or strains on the transformation behavior [Frerichs et al. 2007, Wynne et al. 2007, Ahrens et al. 2004, Larn et al. 2000, Shipway and Bhadeshia 1995b]. However, during forging, a locally incomplete austenitization treatment is present, and thus, has to be considered for a realistic modeling effort.

Depending on the subsequent cooling conditions different microstructures can evolve, such as pure martensite, pure bainite, or a mixture of both. During continuous cooling experiments of a 16MND5 steel, it was shown that the mechanical properties of mixed bainitic-martensitic microstructures followed a linear mixture rule [Petit-Grostbussiat et al. 2004]. Certainly, during the subsequent cooling sequence of the forging process not only a continuous bainitic-martensitic transformation, but also a combination of an almost isothermal bainitic formation followed by a continuous martensitic reaction is possible. Thus, experimental data is required to decide whether a linear mixture rule following the aforementioned cooling scenario for the bainitic-martensitic microstructure of the 51CrV4 steel is also present.

In order to shed light on several of the ongoing phenomena during the considered forging process a thorough experimental investigation was undertaken, including the influence of the prior austenitization treatment on the bainitic and martensitic phase transformation kinetics, the effect of pre-deformation temperature in the case of combined pre-strain – stress experiments as well as the influence of the bainitic volume fraction in bainitic-martensitic microstructures on the stress-strain response. The results clearly show that the prior austenitization treatment has a dominant influence

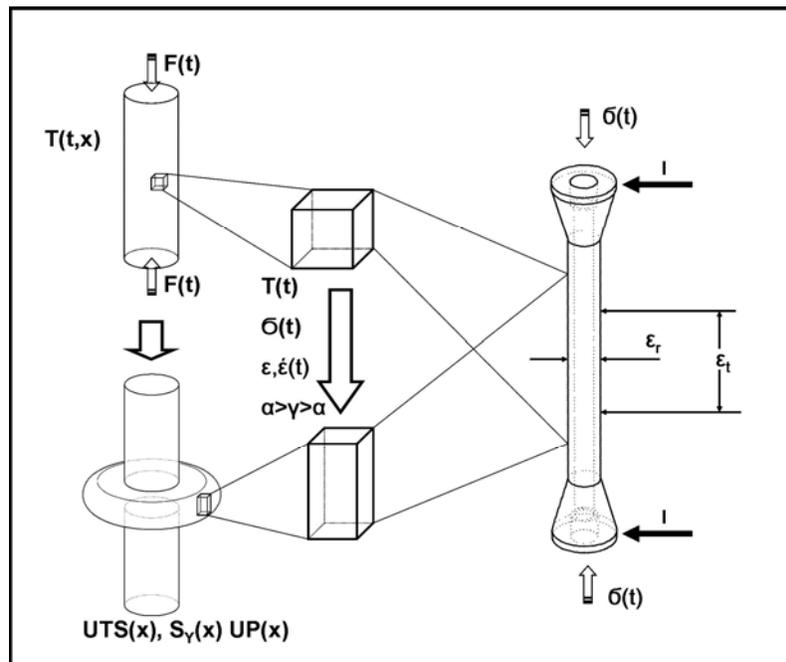


Figure 1: The thermo-mechanical coupled forging process investigated is influenced by several parameters such as strain (ϵ), strain rate ($\dot{\epsilon}$), temperature (T), and stress (σ), which can all vary with time (t) and location within the work piece. Thin-walled hollow specimens were used in order to simulate the local temperature-time-deformation path of selected volume elements in the work piece, schematic recompiled from [Maier et al. 2009].

On the subsequent bainitic and martensitic phase transformation kinetics, whereas the qualitative directions – higher M_s temperatures and longer incubation times for isothermal bainitic reactions with raising austenite grain size – is in accord with published data [Yang and Bhadeshia 2009, Garcia-Junceda et al. 2008, Yang and Bhadeshia 2007, Ahrens et al. 2002, Shipway and Bhadeshia 1995, Bhadeshia et al. 1991, Umemoto and Owen 1974]. However, the observed increase in transformation plasticity value while decreasing difference between T^* and M_s is not predictable with the existing models. Therefore, it is believed that the results presented herein will stimulate further modeling efforts in the field of phase transformation behavior, such that not only the effect of locally different austenitization treatments, but also the effect of simultaneously active strains and stresses during the phase transformation on the dimensional stability can be more precisely predicted.

Experimental details

The material used for the bainitic phase transformation experiments in this study was low alloy 51CrV4 steel from a single production batch. This guaranteed that sample-to-sample variations in the chemical composition for the bainitic phase transformation experiments were negligible (Table 1), which is crucial for the reliability of the experiments.

Table 1. Chemical composition of the low alloy 51CrV4 steel studied. The data shown represent minimum and maximum values obtained from various samples.

Element	C	Cr	Mn	S	Pb	Si	Cu
wt.-%	0.48	0.88	0.72	0.018	0.004	0.26	0.17
	0.48	0.94	0.77	0.021	0.010	0.30	0.19
Element	Al	Ni	Mo	Nb	Ti	P	Fe
wt.-%	0.007	0.09	0.02	0.001	0.011	0.017	Balance
	0.009	0.10	0.02	0.002	0.014	0.021	Balance

Precisely machined thin-walled hollow specimens having an outer diameter of 10 mm and a wall thickness of only 1 mm were used in this study in order to simulate the local temperature-time path of a specific volume element of the considered forging process (Figure 1). Moreover, the effect of stresses and strains on the phase transformation behavior was also detectable with the chosen specimen design. Due to the homogeneous wall thickness a uniform electrical resistance was obtained in the gage section allowing for heating up the specimens within seconds to a temperature as high as 1200 °C via direct current heating. For the phase transformation experiments a custom built test-rig utilizing a servo hydraulic test frame equipped with an axial and a diametral extensometer was used. Nozzles homogeneously distributed around the specimen allowed for a uniform airflow during cooling via gas quenching, and thus, only negligible temperature gradients were present. The temperature was controlled using a two-color pyrometer with a fast response time. In order to ensure steady state temperatures during austenitization and isothermal bainitic transformation an additional thermocouple was mounted inside the thin-walled hollow specimens. For experiments including an austenite-to-martensite reaction the thermocouple was micro-spot welded directly in the gage section of the specimen.

Since the temperature distribution is strongly anisotropic throughout the forging process of a real work piece, different austenitization treatments, as well as cooling scenarios, were considered in this study. For the bainitic and the martensitic phase transformation experiments, three different austenitization treatments were chosen. The first one is a conventional one typically used for the preparation of Time-Temperature-Transformation (TTT)-diagrams [Wever and Rose 1961] utilizing an austenitization treatment at 880 °C for 5 min (heated up within 2 min.). In order to simulate the temperature-time path of a specific volume element of the considered forging process during the austenitization sequence two other austenitization treatments utilizing a heating rate of 45 °C/s and austenitization temperatures of 1050 °C and 1200 °C, were chosen. For the bainitic transformation experiments the holding time was 10 s in both cases. However, for the martensitic phase transformation experiments, utilizing specimens from a new production batch with slightly higher carbon content than tabulated in Table 1 [Lambers et al. 2010], the isothermal holding time at 1050 °C was increased from 10 to 60 s, since an isothermal holding time of 10 s was no longer sufficient to dissolve all carbides during austenitization. As a result the austenite grain size increased moderately from 16 µm (10 s holding) to 22 µm (60 s holding) [Lambers et al. 2010], whereas austenite grain sizes of 16 and 44 µm were obtained following austenitization treatments at 880 °C for 5 min and 1200 °C for 10 s, respectively [Lambers et al. 2009, Lambers et al. 2010]. Since the local cooling conditions, which determine the resulting microstructures and the corresponding mechanical properties, differ depending on the position in the work piece, several cooling scenarios were considered in this study, such that the resulting microstructure was either pure martensitic, pure bainitic, or exhibited a mixture of both. In order to trigger a pure martensitic microstructure, the specimens were rapidly cooled down from the austenitization temperature to room temperature (RT), whereas the chosen cooling rate was fast enough to avoid any other transformation before reaching M_s . For adjusting a pure lower bainitic microstructure, an isothermal hold at a temperature of 340 °C for 30 minutes was chosen before cooling the specimens down to RT. Thus, the aforementioned experimental scheme allowed for determining the effect of the austenitization treatment on both, the isothermal bainitic phase transformation kinetics and on the martensite start temperature (M_s temperature).

In addition, during forging, stresses and strains are simultaneously active, affecting the phase transformation behavior even further. In many previous studies examining the effect of pre-deformation on the dimensional stability, the pre-deformation temperature was fixed [Grostabussiat et al. 2001, Larn and Yang 2000, Larn and Yang 1999, Shipway and Bhadeshia 1995b]. However, during forging the deformation sequence occurs in a broad temperature range, such that the local pre-deformation temperature (T^*) and its effect on the phase transformation kinetics and the dimensional stability has to be considered. Therefore two different pre-deformation temperatures were employed in this study following an austenitization treatment at 880 °C for 5 min. The M_s following the aforementioned treatment was determined to be 330 °C [Lambers et al. 2010], such that the T^*-M_s values for the considered pre-deformation temperatures of 400 and 500 °C in this study were 70 and 170 °C, respectively. Measuring both, axial (ε_a) and diametral (ε_d) strains throughout the experiments allowed for determining the transformation plasticity (ε_{TP}) strains with respect to the loading axis:

$$\varepsilon_{TP}(t) = \varepsilon_a(t) - \frac{\Delta V}{3V}(t) \approx \frac{2}{3}(\varepsilon_a(t) - \varepsilon_d(t)) \quad (1)$$

Figure 2 illustrates schematically the temperature-time-path for triggering pure bainitic or pure martensitic microstructures. For triggering a mixed microstructure consisting of x % bainite and (100-x) % martensite, with x = 30, 60 or 75, the isothermal holding time at 340 °C was adjusted accordingly, ranging from 60 to 90 s following an

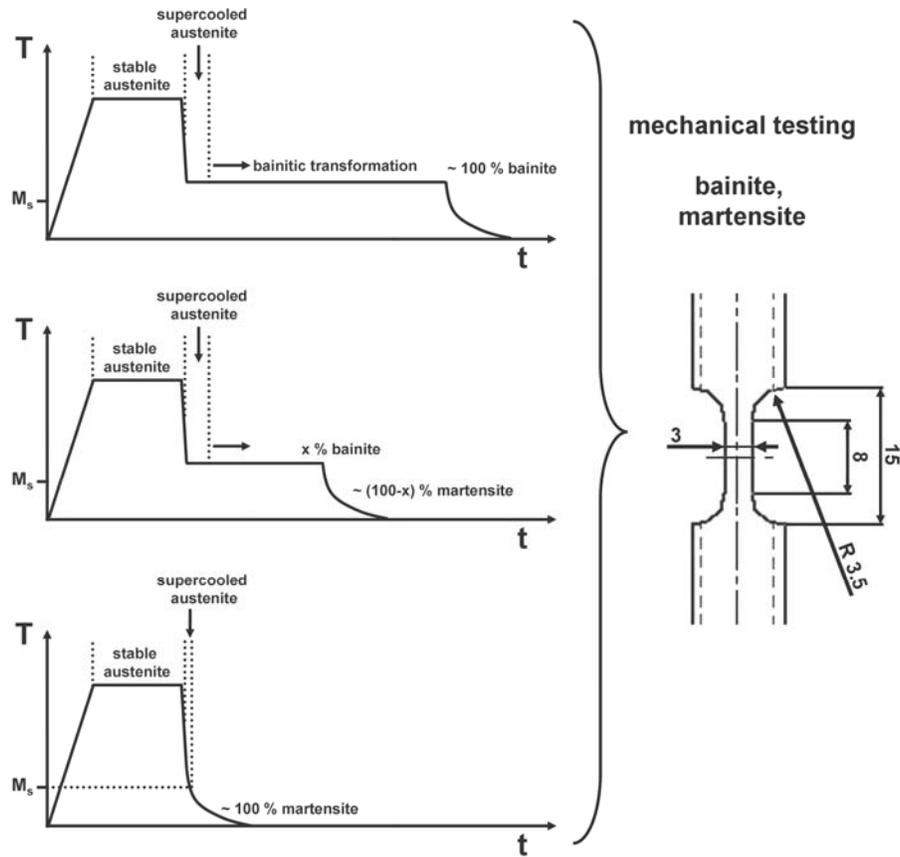


Figure 2: Schematic illustration of the cooling conditions considered in the present study, resulting in pure bainitic (a), pure martensitic (c) or a mixture of both microstructures (b) at RT. For the subsequent tensile test the cross-section in the gage section was reduced in order to guarantee for a failure within this regime.

austenitization treatment at 880 °C for 5 min. After the isothermal hold the specimens were rapidly cooled down to RT, such that the residual supercooled austenite completely transformed to martensite.

In order to determine the stress-strain behavior of the bainitic, martensitic and bainitic-martensitic microstructures, the cross-sections in the gage length were reduced after the phase transformation was completed as indicated in Figure 2. Without the reduction of the cross-section, a failure within the gage section could not be guaranteed since the completely transformed specimens had substantially higher strength in the gage length than in the remainder of the sample. The chosen nominal strain rate for all tensile tests conducted at RT was $4.3 \times 10^{-5} \text{ s}^{-1}$ which is in accord with the German Standard for tensile tests at RT [DIN EN 10002_1].

Since the main deformation sequence during forging occurs when the material is in its supercooled austenitic state, data on the stress-strain response of the supercooled austenite were also required in order to accurately simulate the forging process. However, there is only a small time gap, in which the stress-strain response of the supercooled austenite can be determined (Figure 2). Furthermore, when stresses or strains are prevalent the transformation settles in earlier, such that the time gap for tensile testing is even smaller than expected from the transformation behavior under zero-stress conditions. Thus, for determining the stress-strain response of the supercooled austenite at 340 °C following an austenitization treatment at 1200 °C for 10 s, a relatively fast strain rate of $1 \times 10^{-3} \text{ s}^{-1}$, faster than the recommended strain rate for high temperature tensile tests ($3.3 \times 10^{-4} \text{ s}^{-1}$ to $3.3 \times 10^{-3} \text{ s}^{-1}$), was chosen [DIN EN 10002_5]. Using the measured values of both, axial (ϵ_a) and diametral (ϵ_d) strains the relative volume change ($\Delta V/V$) can be determined:

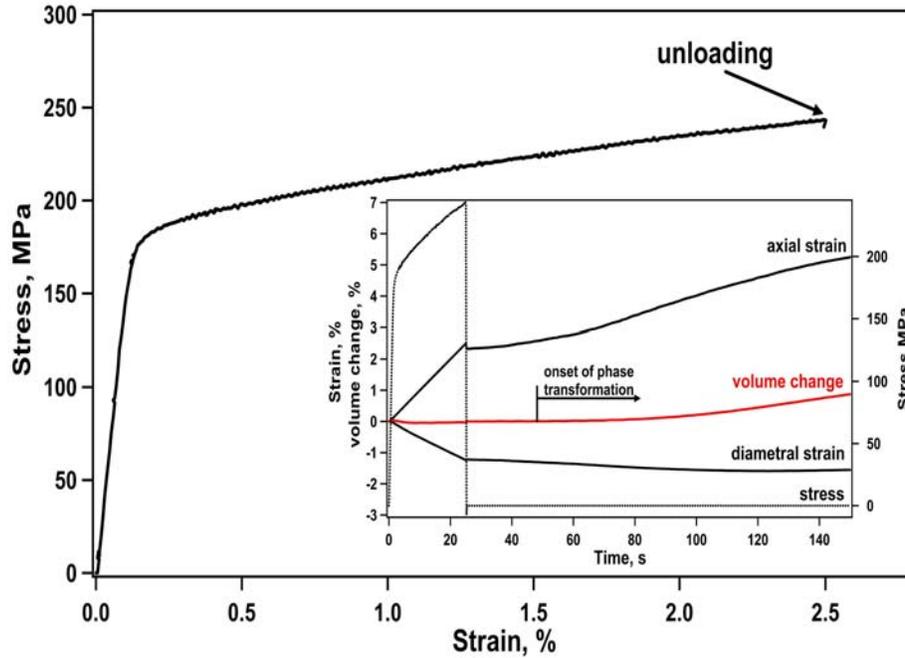


Figure 3: Stress-strain response of the supercooled austenite at 340 °C following an austenitization treatment at 1200 °C for 10 s. The chosen nominal strain rate was $1 \times 10^{-3} \text{ s}^{-1}$. In order to avoid the onset of transformation during tensile testing, the specimen was unloaded at 2.5 % total strain.

$$\frac{\Delta V}{V} = (1 + \varepsilon_a) \cdot (1 + \varepsilon_d)^2 - 1 \quad (2)$$

Thus, the onset of the transformation during tensile loading was detectable, since it is accompanied by a volumetric change.

Results and Discussion

In the current study a thorough experimental program was undertaken in order to allow for modeling and distinguishing between the effect of prior austenitization treatment, superimposed stress level or pre-strains on the bainitic and martensitic phase transformation behavior. In addition, the mechanical properties of several microstructures present at the end of the process were determined, including pure bainitic, pure martensitic as well as supercooled austenitic microstructures.

Especially the experimental determination of the flow behavior of supercooled austenite was of interest, since the estimation of mechanical properties of supercooled austenite using data obtained from stable austenitic steels is accompanied by large discrepancies, as evidenced in an earlier study [Ahrens 2003]. Therefore, the stress-strain response of the supercooled austenite at 340 °C following an austenitization treatment at 1200 °C for 10 s was determined (Figure 3). In order to avoid any transformation during the tensile test, the specimen was unloaded after reaching a total strain of 2.5%. Higher strains were not considered, since they would shift the onset of the transformation to earlier times, even earlier than indicated in the insert of Figure 3. The fact that no transformation occurred during the tensile testing can be easily seen from the insert in Figure 3 showing the evolution of axial, diameteral and volumetric strain as well as stress throughout the tensile test and the subsequent isothermal hold. Due to the applied strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ the axial strain increased linearly with the aforementioned rate, and the diameteral strain decreased. However, no change in volume throughout the tensile test was observed, clearly showing that no transformation occurred during tensile loading. The determined Young's Modulus of approximately 150 GPa and the 0.1 % offset yield stress of 170 MPa for the supercooled austenite at 340 °C are in good agreement with results of an

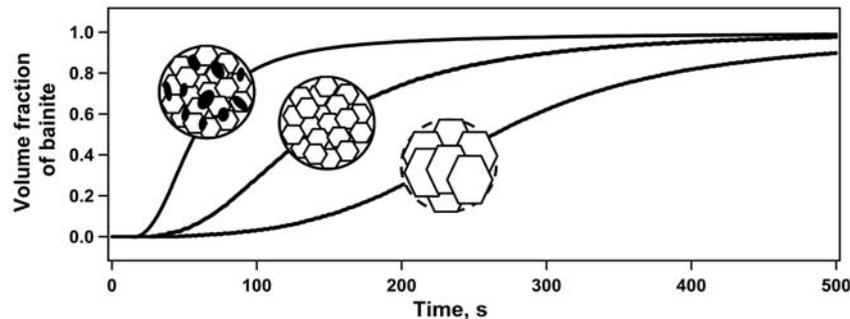


Figure 4: Influence of the prior austenitization treatment on the isothermal bainitic phase transformation kinetics at 340 °C. Austenitization treatments at 880 °C for 5 min, 1050 °C for 10 s and 1200 °C for 10 s resulted in small austenite grains with residual carbides (indicated as black ellipsoids), small austenite grains without carbides and large austenite grain sizes, respectively. The transformation curves following the austenitization treatments at 880 and 1050 °C were recomputed from [Lambers et al. 2009].

earlier study, where the stress-strain response of a low alloy steel with 0.4 wt-% carbon in its supercooled austenitic state at 350 °C was determined [Ahrens et al. 2004].

Figure 4 shows the evolution of the bainite volume fraction with time at 340 °C depending on the prior austenitization treatment. In fact, three different austenitization treatments, 880 °C for 5 min, 1050 °C for 10 s and 1200 °C for 10 s were considered, leading to small austenite grains with residual carbides, small austenite grains and large austenite grains, respectively. Obviously the onset of the bainitic phase transformation is shifted to later times with increasing grain size and increasing degree of carbide dissolution. When carbides are present due to insufficient austenitization treatment, the carbon content in the austenitic matrix is lower than expected from the average value in the specimen as examined via spectroscopy. The lower carbon content in the austenite matrix helped reducing the energy needed for bainitic transformations, accompanied by displacive as well as diffusive, transformation mechanisms, such that the start and end points of the transformation were shifted to earlier times [Ahrens et al. 2002, Shipway and Bhadeshia 1995, Bhadeshia et al. 1991]. Similarly, nitrogen hardening or decarburization during the austenitization treatment might explain the observed shift in transformation times. However, the aforementioned reasons could be excluded from the list of possible ones in an earlier study [Lambers et al. 2009]. Since the austenite grain size was the same following the austenitization treatments at 880 °C for 5 min and 1050 °C for 10 s, only the reduced carbon content in the austenite due to an insufficient austenitization treatment at 880 °C was responsible for the shift in transformation times to earlier times in the presence of residual carbides (Figure 4, small grains with residual carbides) [Lambers et al. 2009]. When the austenitization treatment is changed from 1050 °C for 10 s to 1200 °C for 10 s the isothermal bainitic phase transformation kinetics were slowed down even further (large grains, Figure 4).

Similar results were obtained during the formation of upper or lower bainite in continuous cooling experiments, where the overall transformation rate decelerated with increasing austenite grain size [Lee et al. 2008]. Since the austenite grains are potential nucleation sites for the bainitic reaction, a decrease in austenite grain size increases the austenite grain surface area, and thus, the number of potential nucleation sites [Jacques 2003], leading to an accelerated transformation. Indeed, this is only the case when a slow transformation rate determines the overall reaction time. When a fast growth rate from a limited number of nucleation sites dictates the transformation behavior, smaller austenite grains might reduce the volume transformed per nucleus, and thus, the reaction time would increase [Matsuzaki and Bhadeshia 1999]. Mostly, the bainitic reaction is accompanied by an autocatalytic process, *i.e.*, nucleation and growth at the tip of the bainite platelet. However, when the austenite grain size is reduced to the length of one bainite platelet the autocatalytic process is no longer possible, and thus, the slope of the bainite volume fraction vs. time curve is smaller for very small austenite grains [Jacques 2003]. In this study the total transformation time was increased with increasing austenite grain size indicating that the evolution of the bainite fraction with time was determined by a slow growth rate [Matsuzaki and Bhadeshia 1999]. Due to the decreased grain boundary area per volume when the austenite grain size was increased less possible nucleation sites were present resulting in a delayed start of the transformation. Moreover, also a smaller transformation rate was found for larger grains, and thus, the autocatalytic process was still active for the smaller austenite grain sizes.

Figure 5 shows the evolution of martensitic volume fraction with temperature for a stress-free transformation following different austenitization treatments. Three austenitization treatments were considered in this study and it was

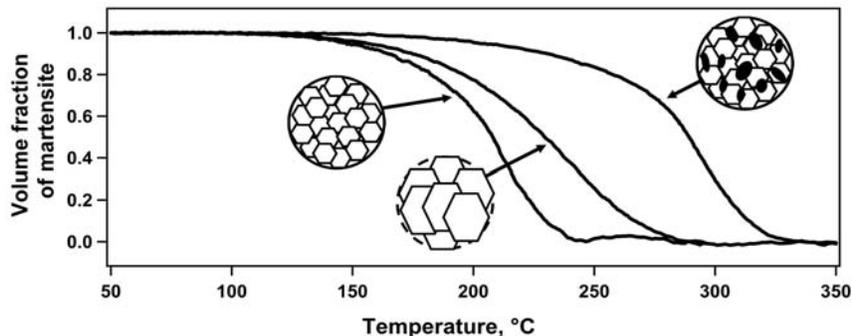


Figure 5: Influence of the prior austenitization treatment on the martensitic phase transformation behavior, data compiled from [Lambers et al. 2010]. Austenitization treatments at 880 °C for 5min, 1050 °C for 60 s and 1200 °C for 10 s resulted in small austenite grains with residual carbides (indicated as black ellipsoids), small austenite grains without carbides and large austenite grain sizes, respectively.

shown that the specimens having the smallest austenite grains in the presence of residual carbides (austenitized at 880 °C for 5 min) exhibited the highest M_s . When all residual carbides were dissolved during the austenitization treatment an increase in M_s with increasing austenite grain size was observed.

In literature several reasons are mentioned in order to explain the observed increase in M_s with increasing austenite grain size. In some studies it was argued that the increase in austenitization temperature increases the quenching temperature, and thus, increases the number of frozen-in vacancies. These structural defects might act as potential nucleation sites for martensitic transformations, such that the specimen austenitized at higher temperatures exhibited higher martensite start temperatures [Nishiyama 1978]. Therefore, the change in austenite grain size due to a change in austenitization treatment would only be a dependent variable, and the dominant factor would be the quenching temperature. By contrast, it was shown for the bursting-type Fe-Ni-C and Fe-Ni alloys that the quenching temperature itself does not affect the onset of martensitic transformation [Umemoto and Owen 1974]. In that study double step austenitization treatments were performed, including a high temperature austenitization to generate coarse austenite grains and a second austenitization step at a lower temperature, such that M_s following this treatment was the same as following a single step austenitization at the higher temperature [Umemoto and Owen 1974]. When the austenitization temperature is increased, coarser austenite grains are prevalent, such that the interference between austenite grain size and the autocatalytic behaviour of the burst type transformation is lower, as compared to the case of smaller grains. Thus, for a burst type transformation it was concluded that the austenite grain size itself is the decisive parameter and not the quenching temperature per se.

A lath type martensitic transformation proceeds more gradually. Since this kind of transformation is accompanied by nucleation at austenite grains, as well [Brofman and Ansell 1983], the aforementioned explanation is not valid. A decrease in austenite grain size would be accompanied by an increase in potential nucleation area per volume, and the martensite start temperature should therefore increase, but in [Brofman and Ansell 1983] the opposite case was observed. It was concluded that the only viable explanation for the observed behavior is that a decrease in austenite grain size increases the strength of the austenite via the well known Hall-Petch relationship. The strengthening of the austenitic microstructure, therefore, depresses the martensitic transformation [Brofman and Ansell 1983]. Several grain size strengthening theories were reviewed in [Hirth 1972] and some experimental evidence was found that an increase in dislocation density is accompanied by a decrease in austenite grain size. Thus, an increase in dislocation density strengthens the austenitic matrix by increasing the resistance against plastic deformation both locally and macroscopically. This strengthening is accompanied by an increase in non-chemical free energy which opposes the martensitic transformation, and thus, leads to a decrease in martensite start temperature [Garcia-Junceda et al. 2008].

Since the Hall-Petch relationship is associated with slip across grain boundaries, it is possible that a more localized resistance against dislocation motion has to be considered in terms of M_s [Yang and Bhadeshia 2009]. Since the determination of M_s by dilatometric tests is dependent on a volume change, the measured temperature does not represent the first occurrence of a martensite nucleus, but the growth of the nuclei to some extent. A geometrical partitioning model suggested that the volume fraction of martensite formed in the early stage of transition is proportional to the cube of the austenite grain size. Therefore, the amount of transformed martensite required to detect M_s is reached at a smaller degree of undercooling for the larger austenite grains, and thus, the M_s temperature is higher for larger grains [Fisher et al. 1948]. Since the measured M_s temperatures refer to the growth of the first nuclei, and not

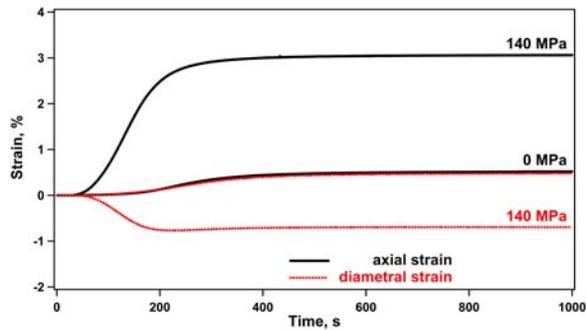


Figure 6 Evolution of axial and diametral strain during isothermal bainitic phase transformation at 340 °C following an austenitization at 1200 °C for 10 s with and without superimposed stress.

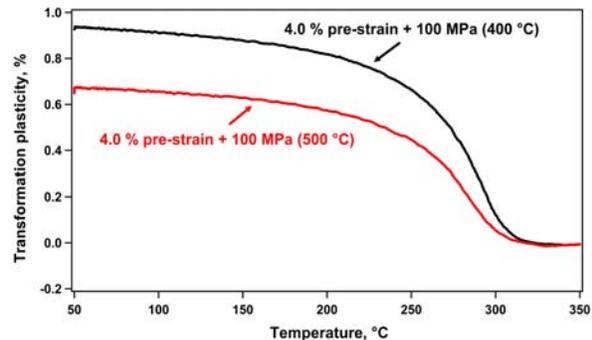


Figure 7: Evolution of transformation plasticity strain with temperature during martensitic transformation. Two pre-deformation temperatures were considered following an austenitization treatment at 1200 °C for 10 s. Data recomputed from [Lambers et al. 2010].

to the occurrence of the first nucleus, the results are strongly dependent on the sensitivity and resolution of the employed equipment [Yang and Bhadeshia 2007].

In this study M_s temperatures of 330, 280 and 250 °C were determined following the austenitization treatments at 880 °C – 5 min, 1200 °C – 10 s and 1050 °C – 60 s, respectively. Thus, the increase in M_s temperature with increasing austenite grain size when the austenitization treatment is changed from 1050 °C – 60 s (austenite grain size: 22 μm) to 1200 °C – 10 s (austenite grain size: 44 μm) is in accord with the aforementioned theories. When the austenitization took place at 880 °C for 5 min, even smaller grains sizes of 16 μm were present, and thus, a further decrease in M_s temperature should be expected. However, the treatment at 880 °C for 5 min was not sufficient to dissolve all residual carbides such that the carbon and chromium contents in the austenite matrix were lower than expected from the measured chemical composition of the bulk material [Lambers et al. 2009]. Thus, due to the decreased carbon and chromium contents in the austenite the energy needed for the martensitic reaction is reduced leading to higher M_s temperatures.

Moreover, during forging not only the temperature but also the stress and strain distributions throughout the process are inhomogeneous. Especially stresses superimposed during the phase transformation lead to transformation plasticity strains, which has to be considered in order to avoid dimensional instabilities. When the isothermal bainitic phase transformation at 340 °C following an austenitization at 1200 °C for 10 s proceeded without any superimposed stress almost no differences between axial and radial strains throughout the transformation were visible (Figure 6), showing that the occurring volume change was isotropic, and only negligible temperature gradients were present in the gage length. In contrast, when an elastic stress of 140 MPa was superimposed during the transformation the occurring volume change was anisotropic and brought about transformation plasticity strains. In the current case axial and diametral strains of 3 % and -0.7 % were observed at the end of the transformation, respectively, such that the total transformation plasticity strain was 2.5 %.

However, during forging not only stresses but both strains and stresses are simultaneously active further affecting the dimensional stability. Recent studies showed that the transformation plasticity strain at the end of the transformation is independent of the applied temperature as long as the superimposed stress level is lower than the yield stresses of the supercooled austenite. In fact, overall transformation plasticity strains of 0.6 % were observed when an elastic stress of 100 MPa was applied either at 500 or 400 °C [Lambers et al. 2010]. However, when the specimens were pre-strained prior to the phase transformation, the pre-deformation temperature strongly affected the amount of transformation plasticity at the end of the transformation. Significant transformation plasticity strains evolved during the isothermal bainitic transformation when pre-straining occurred at the transformation temperature of 340 °C whereas no transformation plasticity strains evolved when the specimens were pre-strained at the austenitization temperature of 880 °C [Lambers et al. 2008]. Similar results were found for the martensitic reaction following an austenitization treatment at 880 °C for 5 min. In that case an increase in pre-deformation temperature of 100 °C (from 400 to 500 °C) decreased the resulting transformation plasticity values at the end of the transformation from 0.15 % to 0.05 %. [Lambers et al. 2010].

By simply adding the transformation plasticity strains obtained from single experiments, *i.e.* only pre-straining (T^* dependent) or only the superimposition of 100 MPa during the transformation (T^* independent), overall transformation plasticity values of 0.65 % ($T^* = 500$ C) and 0.75 % ($T^* = 400$ °C) were expected for the experiment

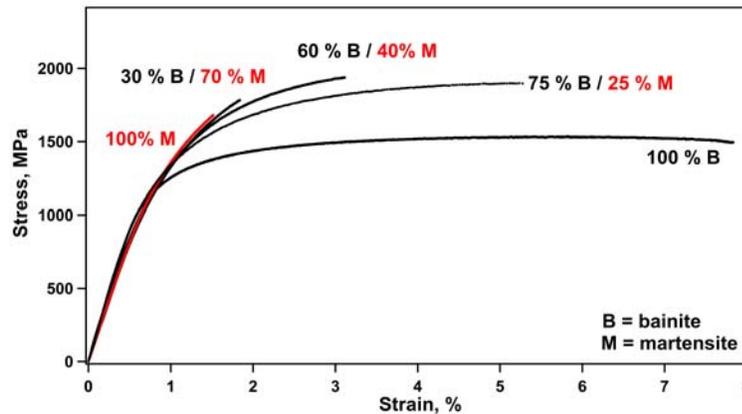


Figure 8: Stress-strain response of 51CrV4 steel in different microstructural states. The different volume fractions of bainite were formed isothermally at 340 °C following an austenitization treatment at 880 °C for 5 min.

combining both pre-straining and the superimposition of stress during transformation [Lambers et al. 2010]. However, the expected value was only obtained when the pre-straining occurred at 500 °C ($T^*-M_s = 170$ °C) as shown in Figure 7. When T^* decreases to 400 °C the simple addition of transformation plasticity values from single experiments underestimated the resulting transformation plasticity strains in the combined experiment. In fact, the observed transformation plasticity strain in the combined experiment was 20 % higher than the value obtained from simply adding the results of the single experiments. The reason for the observed behavior is attributed to the superimposition of internal stresses, generated during pre-straining, and externally superimposed stresses such that the overall stress state is higher. Due to the higher prevalent stresses more preferred variants evolve leading to higher transformation plasticity strains as expected from superposing single experiment results [Lambers et al. 2010].

Figure 8 shows the stress-strain response of pure bainitic, pure martensitic and mixed microstructures at RT, whereas three mixed microstructures containing 30, 60 or 75 % bainite were considered in this study. Obviously, the elongation to failure is strongly dependent on the present microstructure and increased with increasing bainitic volume fraction. However, no systematic increase in ultimate tensile strength was observed when the microstructure gradually changed from pure bainite to pure martensite. In fact, the tensile strength increased from 1530 MPa to 1940 MPa while changing the microstructure from pure bainitic to 60 % bainite and 40 % martensite and then decreased to 1680 MPa upon further increasing the martensite volume fraction to 100 %. Thus, no linear mixture rule was present for bainitic-martensitic microstructures of low alloy 51CrV4 steel as observed earlier for bainitic-martensitic microstructures of 16MND5 steel [Petit-Grostabussiat et al., 2004]. However, immense differences between both steels in carbon content in the austenite prior to the martensitic transformation were present, 0.16 wt-% in the case of 16MND5 and at least 0.41 wt-% in the case of 51CrV4 steel. The last value was calculated using the M_s temperature and the chemical composition as input parameters in order to determine the carbon content with the equation given in [Capdevilla et al. 2002]. Due to the higher carbon content in the 51CrV4 steel the amount of embedded carbon in the martensite is also increased, resulting in an even more distorted matrix. The distortion hinders the dislocation motion, and thus, an increase in strength is expected. However, also the notch-sensitivity increased and when a critical value of carbon dissolved in the martensite matrix was reached, the increase in notch-sensitivity dominated over the expected increase in strength. As a result, the martensitically transformed specimens failed almost within the elastic regime. Thus, due to the high notch sensitivity of 51CrV4 steel in its martensitic state no linear mixture rule was present for a mixed bainitic-martensitic microstructure.

The current work was undertaken in order to provide data suitable for modelling complex manufacturing processes including austenite-to-bainite and austenite-to-martensite phase transformations more accurately. It was found that the effect of the austenitization treatment accompanied by a change in austenite grain size and carbide dissolution on the subsequent bainitic phase transformation kinetics and on the onset of the martensitic transformation was qualitatively in accord with published data. However, no linear mixture rule for bainitic-martensitic microstructures was observed for 51CrV4 steel, which needs to be incorporated in current models. In addition, also T^* has to be considered when modeling complex production processes, where stresses and strains are simultaneously active, in order to avoid dimensional instabilities.

Conclusions

In this study the effect of prior austenitization treatment, pre-deformation temperature and stress-state on the subsequent phase transformation behavior during bainitic and martensitic transitions was investigated. Three different austenitization treatments were chosen in order to distinguish between the effect of austenite grain size and residual carbides on the isothermal bainitic phase transformation behavior and the onset of the martensitic transformation. In addition, the stress-strain responses of the supercooled austenite, as well as pure bainitic, pure martensitic, and bainitic-martensitic microstructures were determined. From the results presented herein the following conclusions can be drawn:

- The prior austenitization strongly affects the subsequent isothermal bainitic phase transformation kinetics, and it was shown that start and end of the bainitic transformation were shifted to earlier times when the austenite grain size was reduced from 44 μm (austenitized at 1200 $^{\circ}\text{C}$ for 10 s) to 16 μm (austenitized at 1050 $^{\circ}\text{C}$ for 10 s). Since the austenite grains can act as potential nucleation sites less sites are available when the austenite grain size is increased resulting in longer transformation times. Following an austenitization at 880 $^{\circ}\text{C}$ for 5 min, resulting also in an austenite grain size of 16 μm , even shorter transformation times due to the existence of residual carbides not dissolved during the austenitization treatment were detected.
- The onset of the martensitic transformation was also dependent on the prior austenitization, whereas the specimens having the smallest austenite grains in the presence of residual carbides (austenitized at 880 $^{\circ}\text{C}$ for 5 min) showed the highest M_s . Due to the residual carbides the amount of carbon and chromium in the austenitic matrix is reduced resulting in higher M_s temperatures. When all residual carbides were dissolved during the austenitization treatment an increase in M_s temperature with increasing austenite grain size was observed.
- Transformation plasticity strains evolved during both - isothermal bainitic and continuous martensitic phase transformations, when stresses even lower than the yield stress of the supercooled austenite were superimposed during the transformation. In addition, when the martensitic transformation proceeded following a pre-straining under stress the transformation plasticity value at RT strongly depended on the pre-deformation temperature T^* , and higher transformation plasticity strains at RT were observed for lower T^* temperatures. In this case internal stresses generated during pre-straining were still active at M_s , and thus, can interact with the additional superimposed elastic stresses. As a result, higher overall stress values were present, resulting in higher transformation plasticity strains at RT.
- The tensile stress-strain responses of pure bainitic, pure martensitic and bainitic-martensitic microstructures at RT were determined, demonstrating that no linear mixture rule for the bainitic-martensitic microstructures was present. Due to the high amount of carbon dissolved in the 51CrV4 steel the martensitically transformed specimens were highly notch-sensitive and failed almost in the elastic regime.

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