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A Comparative Analysis of Austenite-to-Martensite and Austenite-to-Bainite Phase Transformation Kinetics in Steels

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This paper presents a comparison of the macroscopic transformation strain evolution as a function of the bainite and martensite phase fractions in steels. Specifically, the evolution of anisotropic strain with phase fraction follows a linear trend for the martensitic transformation due to continuous stress-induced variant selection. In the case of bainitic transformation, the anisotropic strain evolves non-linearly owing to diffusion, minimizing the distortion around the bainitic sheaves and further promoting stress-induced variant selection at the early stages of the bainitic phase transformation. However, this effectiveness is reduced when the bainitic sheaves start constricting the growth of each other.

Keywords: Austenite; Bainite; Martensite; Phase transformation; Anisotropic strain

1. Introduction Considering the current industrial trend for improving energy efficient mechanical systems, weight reduction becomes one of the most important targets for various applications, especially in aerospace, defence and automotive industries [1]. The major action taken to achieve weight reduction is sheet metal thickness reduction of structural subsystems, which requires metals with high strength and good formability [2]. This makes bainitic steel a good candidate material for such applications as it fulfils both strength and formability requirements. More specifically, even though formability and strength are trade off material properties, microstructure refinement that takes place during the bainitic phase transformation leads to the improvement of both properties. This, in turn, further emphasizes the importance of fully understanding the bainitic phase transformation and the corresponding microstructure evolution in order to accurately predict the structural performance of these steels and the associated final product quality [3]. Furthermore, production processes involving bainitic phase transformations are often accompanied by stress and temperature gradients yielding undesired shape changes in the final work pieces, i.e. non-isotropic strains evolving when stresses are superimposed during the phase transformation [4–6]. Thus, the work pieces often need post-processing by milling or turning, resulting in additional production costs. Consequently, modelling of anisotropic strain evolution gains significant importance in reducing the phase transformation-induced distortion, and thus, warranting a solid knowledge of the processes active at the microscopic level.

Studies carried out to understand the underlying mechanisms of bainitic phase transformation and the resulting microstructure – texture – mechanical property relationship have focused either on transformation kinetics theory [1,7–13], or the evolution of final transformation strains at the macroscopic scale and the corresponding stress-induced variant selection [14–17].

These works established the kinetics theory of the bainitic phase transformation, such that the transformation kinetics following different thermal treatments and the resulting microstructures can be predicted. However, when it comes to predicting the macroscopic transformation strains upon austenite-to-bainite phase

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transformation, the evolution of anisotropic strain throughout the transformation shall also be emphasized for a full understanding of the phase transformation processes. However, when it comes to predicting the macroscopic transformation strains upon austenite-to-bainite phase transformations, the martensite transformation theory is often directly adopted and used for calculating the anisotropic strains [14–17]. Even though the corresponding results are reasonable in most cases, it is important to develop a theory consistent with the already established kinetics of the bainitic transformation, including the lattice deformation prevalent during the bainitic phase transformation [17,18]. It is important to note that a key parameter, especially when variant selection is active during the phase transformation, is the microstructural evolution, dictating the transformation strains at the macroscopic scale. Thus, the evolution of transformation strains is unique for the austenite-to-bainite phase transformation, and hence, must be considered within the models.

The current work was undertaken with the purpose of showing the necessity of a theory specific to the austenite-to-bainite transformation, which is different than the austenite-to-martensite transformation theory commonly utilized in predicting macroscopic transformation strains for both the bainitic and martensitic phase transformations. For this purpose, the role of the microstructural processes active during the austenite-to-bainite phase transformation on the resulting transformation strains was examined as compared with those active in the austenite-to-martensite phase transformation in the presence of variant selection. Following the phase transformations, the resulting microstructures where investigated by transmission electron microscopy (TEM) to (i) identify the phases evolving in the phase transformations and (ii) determine whether diffusion was active during the phase transformation or not. The corresponding results evidence a non-linear relationship between the evolution of anisotropic strain and the volume fraction of bainite. In contrast to the bainitic phase transformation, the evolution of anisotropic strain and the volume fraction of martensite follows a linear relationship. Consequently, the current study constitutes a step towards a more accurate and complete modelling of the anisotropic strains upon isothermal austenite-to-bainite phase transformations.

2. Experimental details In order to expose the differences in the microstructural evolution of the austenite-to-bainite and austenite-to-martensite phase transformations, two completely transforming steels, namely a low-alloy 40 CrMnMoS 8-6 tool steel and a low-alloy 51 CrV 4 steel, were studied. The investigated steels exhibited only minor sample-to-sample variations in the chemical composition, which is important as the phase transformation kinetics strongly depend on the chemical composition as illustrated in Table 1 for the tool steel and in Table 2 for the low-alloy steel [13,19,20].

Table 1. Chemical composition of the 40 CrMnMoS 8-6 steel studied.

Element	C	Cr	Mn	S	Pb
Mass content in %	0.33 0.34	1.71 1.75	1.42 1.45	0.055 0.062	0.003 0.004
Element	Si	Ni	Mo	Nb	Fe
Mass content in %	0.27 0.28	0.10 0.11	0.13 0.14	0.002 0.004	Balance

Note: The data shown represent the minimum and maximum values obtained from various samples, recompiled from [20].

Table 2. Chemical composition of the 51 CrV 4 steel studied.

Element	C	Cr	Mn	S	Pb
Mass content in %	0.48 0.48	0.88 0.94	0.72 0.77	0.018 0.021	0.004 0.010
Element	Si	Ni	Mo	Nb	Fe
Mass content in %	0.26 0.30	0.09 0.10	0.02 0.02	0.001 0.002	Balance

Note: The data shown represent the minimum and maximum values obtained from various samples, recompiled from [13].

The experiments were carried out on a custom-built test rig featuring a servohydraulic test frame, extensometers measuring the axial and diametral strain simultaneously, a two-colour pyrometer for controlling the temperature-time paths and nozzles used for gas quenching, as well as providing a nitrogen atmosphere to avoid oxidation of the specimens at the austenitization temperature. All specimens were heated up via direct current to the austenitization temperature of 1200°C within 15 s and held at this temperature for 10 s. This austenitization treatment guaranteed that all existing carbides dissolved as already shown in a previous work on the low-alloy steel [13,20]. Afterwards the specimens were quenched by simultaneous reduction of the direct current and application of pressurized air at a nominal rate of 70 K/s, either down to the isothermal phase transformation temperature of 340°C to obtain lower bainite, or down to room temperature to promote martensite. Once the martensite start temperature of the steels was attained, the further nominal cooling rate was 15 K/s. For both steels, the chosen quenching rate guaranteed that no unintended phase transformation took place before reaching the isothermal phase transformation temperature or the martensite start temperature. A tensile stress was be superimposed during both the austenite-to-bainite and austenite-to-martensite phase transformations, such that

anisotropic strains evolve upon the displacive transformations. This stress level was intended to be high enough to observe a noticeable variant selection and low enough to prevent undesired distortions, i.e. below the yield strength of the supercooled austenite. Therefore, tensile experiments of the supercooled austenite were carried out at 400°C which was above the temperature, at which the loads should have been applied (380°C) to the samples of both steels.

All specimens had a constant wall thickness of 1 mm and were precisely machined in order to ensure a homogeneous temperature distribution in the gage length. However, the specimen geometries were different for both materials. In the case of the low-alloy 51 CrV 4 steel hollow specimens with a diameter of 10 mm were used, whereas the samples had a 10 mm wide flat gage section in the case of the 40 CrMnMoS 8-6 steel. The tool steel samples were machined with a different geometry to monitor the local strain evolution by digital image correlation to compliment the macroscopic strains recorded by the extensometers [21]. Therefore, a flat surface in the gage length was required to allow reliable calculation of the local strain fields. It should be noted that similar results should be obtained for both specimen geometries for the same material, as long as the temperature gradient within the specimen is small [22]. However, a comparison between the anisotropic strain evolutions of both specimen geometries should be considered critical as both the specimen geometry and the chemical composition varied between the experiments.

TEM in combination with energy dispersive X-ray spectroscopy was employed to determine whether carbides precipitated during the bainitic and martensitic phase transformations. It should be noted that TEM cannot directly show the diffusion in an *in-situ* manner, yet demonstrate its presence retrospectively. Specifically, martensitic transformation does not facilitate any long range diffusion, whereas small cementite carbides within the bainitic laths emerge, resulting from carbon diffusion during the tempering of the bainite following the formation, which can be observed in post-experimental TEM studies [23,24]. For the TEM analysis, thin slices were extracted from the transformed specimens, which were then ground down to a thickness of 150 μm. Afterwards twin-jet electropolishing utilizing a 5 pct perchloric acid solution under an applied potential of 25 V at −40°C was used to achieve electron transparency. The TEM utilized in this study was operated at a nominal acceleration voltage of 200 kV.

The evolution of the anisotropic strains (ε_A) during the phase transformations can be described, such that the anisotropic strains equal the anisotropic strain component less an assumed isotropic volume change

$$\varepsilon_A = \varepsilon_a - \frac{\Delta V}{3V} \approx \frac{2}{3}(\varepsilon_a - \varepsilon_d), \tag{1}$$

where $\Delta V/3V$ represents the volume change in one direction, and ε_a and ε_d stand for the axial and diametral strains as a function of time or temperature, respectively. The volume fractions of the evolving product bainite phase ($w_b(t)$) as a function of time can be determined on the basis of the momentary volume change $\Delta V/V$ normalized by the volume change at the end of the phase transformation, i.e. when no more phase transformation-induced volume change takes place:

$$w_b(t) = \frac{(\Delta V/V)(t)}{(\Delta V/V)(t \to \infty)}. \tag{2}$$

In addition, the volume fraction of the martensite phase ($w_m(T)$) as a function of temperature can be determined by means of the momentary volume change divided by the volume change of the completed martensitic phase transformation, i.e. when no more phase transformation-induced volume change takes place at low temperatures

$$w_m(T) = \frac{(\Delta V/V)(T)}{(\Delta V/V)(T \to M_f)}, \tag{3}$$

where M_f stands for the martensite-finish-temperature. The volume change $\Delta V/V$ can be obtained from the measured axial and diametral strains ε_a and ε_d:

$$\frac{\Delta V}{V} = (1 + \varepsilon_a)(1 + \varepsilon_d)^2 - 1. \tag{4}$$

The major advantage of the relationships used to compare the transformation kinetics of the martensitic and bainitic phase transformations is that the kinetics are directly ascertainable from ε_a and ε_d measured during the phase transformation experiments without making any assumptions. Many other relationships used to analyze the transformation kinetics utilize coefficients which need to be estimated or determined through an extensive series of experimentation and curve fitting.

3. Results and discussion In order to estimate the stress which should be superimposed during the phase transformations, tensile tests of the supercooled austenite were performed at 400°C for both steels prior to the phase transformation experiments. The important observation made during these tensile tests is that the yield strength of the supercooled austenite is higher than 150 MPa at 400°C for both steels investigated in this work [6]. Thus, the superimposed load during phase transformation was chosen as 100 MPa in order to facilitate noticeable variant selection and to avoid any undesired distortion.

A comparison of the experimentally obtained anisotropic strain evolutions with bainite and martensite volume fraction evolutions is presented for isothermal bainitic and martensitic phase transformations, respectively, in Figure 1. Specifically, the evolutions of both normalized anisotropic strain (Figure 1(a)) and absolute anisotropic strain (Figure 1(b)) are presented as a function

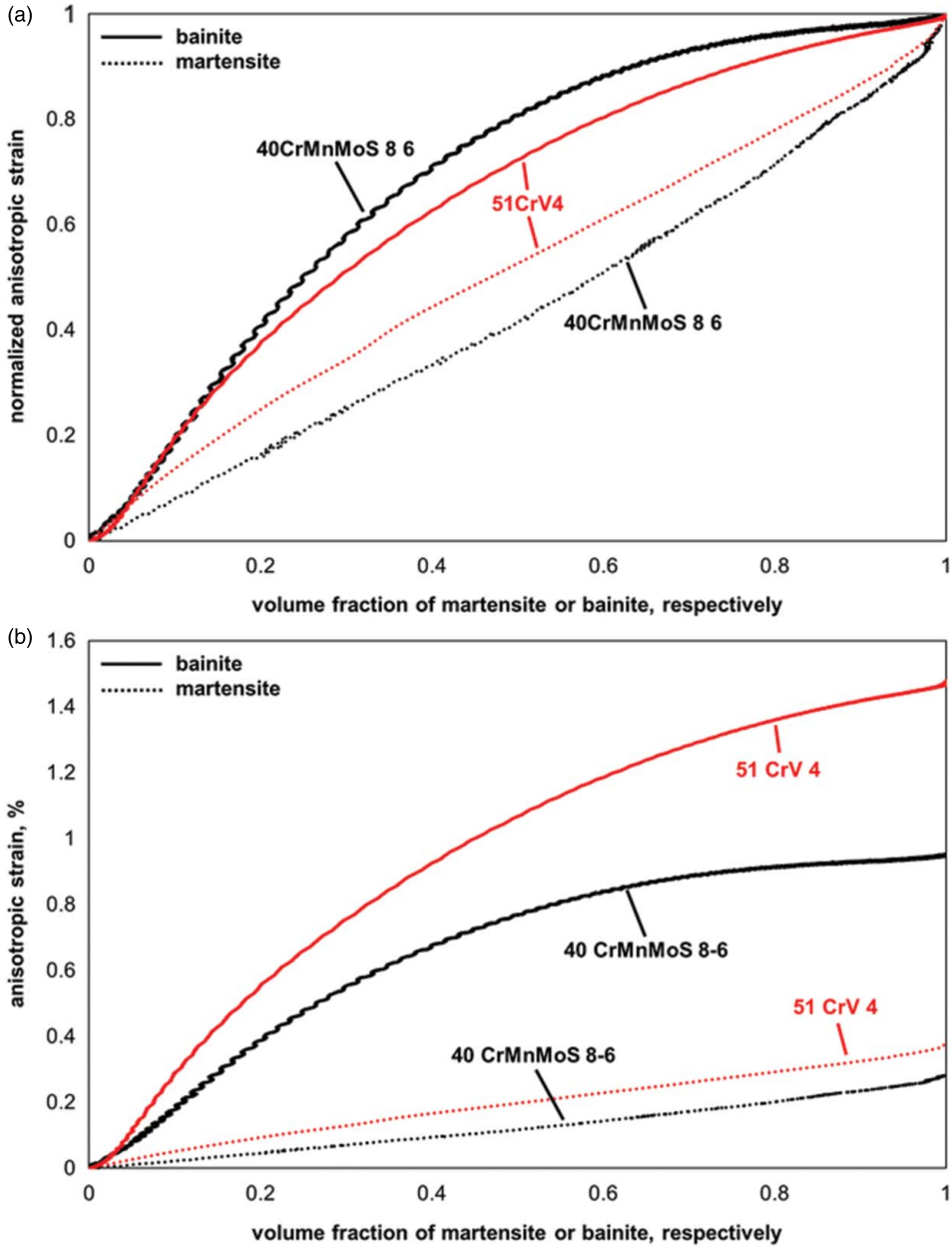


Figure 1. Evolution of anisotropic strain with volume fraction of martensite and bainite presented in (a) normalized and (b) absolute fashion for a martensitic and an isothermal bainitic phase transformation at 340°C under 100 MPa superimposed stress. The phase transformations followed an austenitization treatment at 1200°C for 10 s for both materials.

of the volume fraction. The evolution of the normalized anisotropic strain emphasizes the diverse transformation kinetics of the bainitic and the martensitic phase transformations, and the absolute anisotropic strain illustrates the absolute values of anisotropic strain evolving in the phase transformations, which is not feasible in the normalized way of representation. Note that both the austenite-to-bainite and the austenite-to-martensite phase transformations took place under 100 MPa superimposed stress. Accordingly, for both materials, the anisotropic strain evolution during martensitic transformation follows a nearly linear trend as expected, whereas its dependence on the bainitic volume fraction evolution is significantly more non-linear (Figure 1(a)). It should be noted that it is essential whether a non-linear or linear dependence of the microstructure – texture – mechanical property relationship is implemented in models as the simulated properties strongly depend on these relationships. The following absolute anisotropic strains were recorded in the phase transformation experiments: 0.3 for the martensitic and 0.9 for the bainitic transformations in the tool steel, and 0.4 for the martensitic and 1.4 for the bainitic transformations in the low-alloy steel, cf. Figure 1(b).

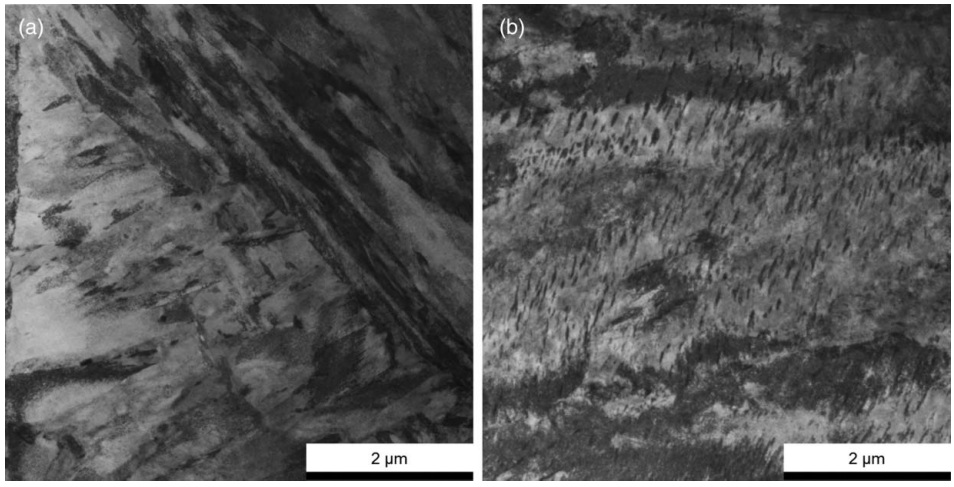


Figure 2. TEM images illustrating the microstructure upon (a) austenite-to-martensite and (b) austenite-to-bainite phase transformations under 100 MPa superimposed stress in a low-alloy 40 CrMnMoS 8-6 tool steel.

The observed linear trend in the martensitic transformation is associated with the continuous effectiveness of the stress-induced variant selection process throughout the transformation [4,25]. During martensitic phase transformation a lath grows rapidly through the austenite grain. The following lath, however; tends to grow perpendicular to the formerly grown one (Figure 2(a)). This happens despite the superimposed stress due to the large amount of distortion within the microstructure associated with the formation of the first martensitic lath, such that a volume difference between the parent and the product phases prevails and the variant selection is affected to a lesser extent by the superimposed stress leading to lower absolute anisotropic strains as compared with the bainitic phase transformation [5,24]. Furthermore, a careful inspection of the microstructure (Figure 2(a)) reveals that no cementite or other carbides precipitated during or after the phase transformation as only one phase (ferritic distorted) can be identified. Consequently, diffusion leading to the evolution of carbides was not active during the martensitic phase transformation and thus, transformation-induced dislocations by the variants distorting the microstructure cannot be depleted during the transformation due to diffusion.

However, in the case of the bainitic transformation, cementite carbides are identified in the microstructure by TEM (Figure 2(b)) in addition to the ferritic phase and, thus, one can deduce that at least carbon diffusion took place during or following the bainitic phase transformation. Hence, when diffusion is active the distortion imposed upon the dislocations at the tips of bainitic subsections forming sheaves of variants can be minimized, such that the growth of the further sheaves is less affected by the surrounding distortion. By means of diffusion, the dislocations straining the microstructure around the subsections of the sheaves can also annihilate each other, and hence, a less distorted microstructure results in the case of the bainitic microstructure as compared with the martensitic case. This, in turn, increases the efficiency of variant selection, such that the driving force for the growth of sheaves oriented with respect to the loading direction increases at the early stages of the phase transformation [23,26,27]. However, this variant selection procedure becomes less prominent as the bainitic transformation proceeds, resulting from a constriction of the growth of sheaves by each other, and thus, the anisotropic strains exhibit a non-linear dependence on the bainitic volume fraction (Figure 1).

It is important to note that both steels exhibit the same trend in terms of the normalized anisotropic strain evolution: a non-linear trend for the austenite-to-bainite and a linear trend for the austenite-to-martensite phase transformation. In this context, the observed relatively more significant difference in the normalized anisotropic strain evolution (Figure 1(a)) between bainite and martensite in the tool steel as compared with the low-alloy steel may indicate that the process of variant selection is more effective in the tool steel, which could stem from a higher content of alloying elements in the microstructure. Specifically, the alloying elements decelerate the phase transformation and, consequently, diffusion would have more time to deplete the deformation at the tips of the laths forming sheaves. Unfortunately, it is not totally clear if the observed effects stem from the higher amount of alloying elements in the tool steel, as both the specimen geometry and the chemical composition are different. Furthermore, not only the substitutional alloy elements but also the interstitial carbon is different in both steels constituting another deviant boundary condition. However, further experiments in the future, where the same specimen geometry for steels with only different substitutional alloying element contents will reveal the validity of this assumption. The larger magnitude of anisotropic strain attained during the bainitic phase

transformation compared with that of the martensitic transformation in both steels derive from a higher ratio of yield strength of the supercooled austenite to applied stress for the martensitic transformation (the ratio is initially 1.8 for the low-alloy steel and 1.4 for the tool steel, and increases with further supercooling) compared with the bainitic transformation (the ratio is constantly 1.5 for the low-alloy steel and 1.1 for the tool steel). This, in turn, means that the ratio of the mechanical to the chemical driving force for the variant selection is larger in the bainitic phase transformation, yielding absolutely larger anisotropic strains upon transformation of austenite to bainite.

4. Conclusions The current study investigated the evolution of macroscopic transformation strains as a function of the bainitic and martensitic phase fractions in low-alloy and tool steels, in order to emphasize the determinant role of the diverse microstructural evolution processes. The findings of this study can be summarized as follows:

- When superimposing a stress of 100 MPa to the austenite-to-martensite and austenite-to-bainite phase transformation, anisotropic strains evolved, whereas the evolution followed a linear trend in the martensite case and a significantly non-linear trend in the bainite case.
- The observed linear trend for the evolution of anisotropic strains with martensitic phase fraction can be attributed to the continuous effectiveness of the stress-induced variant selection.
- The observed non-linear trend for the evolution of anisotropic strains with bainite phase fraction can be attributed to an initially increasing effectiveness of variant selection in succession of diffusion minimizing the distortion around the bainitic sheaves, and a subsequent decreasing effectiveness by a constriction of the sheave growth by each other.
- The larger magnitudes of anisotropic strains in the bainitic phase transformation compared with the martensitic phase transformation can be attributed to a more effective variant selection active during the bainitic transformation in consequence of a larger driving force resulting from a lower ratio of yield strength of the supercooled austenite to applied stress as compared with the martensitic transformation.

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