Because of an excellent combination of strength and ductility, mono-phase low-alloyed steels with a bimodal grain size are an appropriate alternative to conventionally cold-rolled and annealed steels as well as to steels with a dual-phase microstructure. This study investigates how the microstructure of a low-alloyed air-hardening steel either with a homogeneous, a dual-phase, or a bimodal grain structure influences its mechanical and fatigue performances. The homogeneous ferritic grain microstructure of the steel sheets is adjusted by an intercritical annealing at 790 °C, along with subsequent air hardening to obtain a dual-phase state. Then, the ferritic-martensitic material is cold-rolled and annealed at 550–700 °C to produce different bimodal grain microstructures. The evolution of microstructure and mechanical properties are characterized. An annealing temperature of 600 °C is considered to be the optimal temperature resulting in pronounced bimodal grain size distribution. The sheet with a bimodal microstructure exhibits a higher strength and equal ductility compared to one with a homogeneous ferritic microstructure. Additionally, high-cycle fatigue tests of the material with a bimodal microstructure shows its superior fatigue behavior at a loading of above 800 000 cycles compared with both the homogeneous ferritic microstructure as well as the dual-phase microstructure.

1. Introduction

The automotive industry is forced to reduce the weight of its products in order to minimize its footprints in social challenges such as global warming or the finiteness of resources. One way to reach this goal is to use lighter materials in body-in-white products. Additionally, lighter structural parts can be produced by adapting their geometry to specific loading conditions. Therefore, besides using high strength materials that allow for a thickness reduction in structural parts, the sheet material should have a moderate to excellent formability. However, the use of different alloying concepts to meet the mentioned criteria is restricted by the material’s weldability and the resulting costs, which are of high importance to the automotive industry.[1]

New thermo-mechanical processing routes have enabled to produce steels with two or more phase components in their microstructure, i.e., dual-phase (DP) or multiphase steels.[2–4] In DP steels, the presence of both a soft ferrite phase and a hard martensite phase results in a high strength and a moderate formability. Here, the combinations of strength and ductility properties can vary immensely and are easily adjustable through the appropriate parameters of intercritical annealing.[2,3] However, due to localized straining at cold forming, which primarily occurs in the softer ferrite phase, these steels have a comparably low local formability.[5] In contrast to DP steels, complex phase steels consist mainly of bainite or tempered martensite. These phases are harder than soft ferrite but softer than untempered martensite in DP steels, resulting in a high local formability. Yet, due to the reduced ductility of the phases, these steels possess a relatively poor global formability.[5] Furthermore, their high yield strength (YS) results in a significant springback during cold forming and complicates the design process of the cold forming tool.[6]

Because of their microstructure, which largely consists of a ferrite matrix and retained meta-stable austenite, transformation-induced plasticity steels (TRIP steels) have high strength properties similar to those of DP steels, but with a better global formability due to the transformation of metastable retained austenite into martensite during cold forming.[5] In contrast to TRIP steels, twinning-induced plasticity steels (TWIP steels) possess a mainly stable austenitic microstructure at room temperature. Relatively low values of stacking fault energy of austenite in these steels enable cold deformation via the twinning mechanism, which results in a very high strain hardening and thus ductility.[7] However, to retain austenite at ambient temperatures, alloying

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FULL PAPER

Relationships between Microstructural and Mechanical Performance on Example of an Air-Hardening Steel

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elements need to be added, which increases the material cost and reduces the weldability.\textsuperscript{[5]}

Structural parts with both high formability and strength can be produced by a hybrid thermo-mechanical processing.\textsuperscript{[8]} These functionally graded parts have tailored properties, i.e., high ductility and formability in areas that are either to be cold formed or to absorb energy in the case of a crash, as well as a high strength in areas that should prevent intrusions in the case of a crash. This kind of processing can be complex in its practical implementation and needs to be adjusted on a case-by-case basis.

One way to improve a material’s strength independent of its phase content is to refine the microstructure. Grain boundaries, for instance, act as obstacles against crack propagation from one grain to another. Hence, a high amount of grain boundaries results in an improved high cycle fatigue life of structural parts.\textsuperscript{[9,10]} This can be achieved by various processing routes of severe plastic deformation.\textsuperscript{[9]} However, this leads to a reduction in ductility and formability, making fine-grained materials less suitable for automotive applications.\textsuperscript{[9,10]}

Designing a bimodal grain microstructure with heterogeneous grain size improves the ductility of fine-grained materials while maintaining their high strength and resistance to high cycle loads.\textsuperscript{[11–15]} These materials commonly show phase homogeneity, i.e., consisting of one phase, and simultaneously have a heterogeneous grain-size distribution, i.e., comprising coarse grains surrounded by fine grains.\textsuperscript{[13]} In addition to improving a fine-grained steel’s ductility, a bimodal grain microstructure could potentially enhance its corrosion resistance as well.\textsuperscript{[14]}

In the last decade, the topic of bimodal ferritic microstructures in bulk low-carbon steels has been researched extensively. The following processing techniques lead to a bimodal grain size in ferrite at room temperature: 1) Hot deformation at a temperature of 1000 °C, resulting in a partial recrystallization of austenite grains, and thus, in a heterogeneous grain growth of the ferrite during the subsequent controlled cooling.\textsuperscript{[16,17]} 2) Hot deformation slightly above the A\textsubscript{C1}-temperature, allowing for a partial strain-induced \(\gamma\rightarrow\alpha\) transformation, and thus, the formation of a heterogeneous distribution of ferrite grains sizes after a subsequent controlled cooling.\textsuperscript{[16,17]} 3) Hot deformation between the A\textsubscript{C1}- and A\textsubscript{C3}-temperatures, recrystallizing fine ferrite grains due to strain accumulation, and transforming austenite into coarse ferrite grains during a subsequent controlled cooling.\textsuperscript{[17,18]} 4) Hot deformation at 1050 °C, followed by a warm deformation at 550 °C or by a cold deformation with a subsequent rapid annealing between the A\textsubscript{C1}- and A\textsubscript{C3}-temperatures. The rapid annealing of the deformed ferritic microstructure with a heterogeneous distribution of cementite results in the formation of very fine austenite grains in the vicinity of cementite particles as well as in the coarsening of the existing deformed ferrite grains due to recrystallization. A final controlled cooling leads to the transformation of austenite into fine ferrite grains and coarse recrystallized ferrite grains.\textsuperscript{[19]} 5) Annealing between the A\textsubscript{C1}- and A\textsubscript{C3}-temperatures with\textsuperscript{[12,13,14]} or without a deformation,\textsuperscript{[12,13,14]} followed by quenching to obtain a DP microstructure consisting of martensite and ferrite. Cold rolling and annealing below the A\textsubscript{C1}-temperature with a subsequent controlled cooling results in an abnormal coarsening of severely deformed ferrite grains, which are surrounded by fine ferrite grains transformed from martensite.\textsuperscript{[12,13,15]}

The last processing technique, i.e., “annealing between A\textsubscript{C1}- and A\textsubscript{C3}-temperatures → cold rolling → annealing below A\textsubscript{C1}-temperature”, enables a more simple temperature control during the annealing and the deformation steps compared to techniques comprising warm or hot rolling. Due to a high dimensional control, cold rolling is an elementary deformation step to produce thin steel sheets, which are commonly used for structural parts of body-in-white. Therefore, the mentioned processing technique seems to be the most appropriate for the specific goals of the automotive industry.

The temperature during the last annealing step directly influences the extent of the ferrite recrystallization as well as the transformation behavior of martensite into ferrite. Hence, it can be adjusted to reach the desired combination of strength and ductility in steel sheets with a bimodal microstructure. However, this has only been researched superficially to date.\textsuperscript{[12,15]} Furthermore, there is no information about the fatigue behavior of low-carbon steels with a bimodal ferritic microstructure, and therefore, it has not been compared to the fatigue performance of the same steel in other microstructural states. Thus, the aim of this work was to establish relationships between the microstructure and the mechanical properties, i.e., depending on the temperature during the final annealing step, as well as to compare the fatigue behaviors of low-carbon steels with homogeneous ferritic, bimodal ferritic, and ferritic-martensitic DP microstructures.

2. Experimental Section

2.1. Processing of the Investigated Material

The air-hardening steel RobuSal 800 (in the past—LH800) with a thickness of 1.65 mm was examined both in a CR and an annealed state. Due to a well-designed alloying concept (see Table 1), this steel demonstrated a high formability in its annealed state as well as a high strength after austenitization and quenching. Simultaneously, its critical cooling rate amounts to 10 Ks\textsuperscript{-1}, i.e. the steel can be easily hardened by air cooling. More detailed information about this steel’s cold formability in an annealed state as well as its heat treatability can be found elsewhere.\textsuperscript{[2,20,21]}

As the first processing step, the strips cutoff from the sheet that initially had a homogeneous ferritic microstructure (homogeneous microstructure or HM state) were intercritically annealed to obtain a ferritic-martensitic microstructure, with each phase approximately making up 50% (DP microstructure or DP state). Both the temperature and the annealing time necessary to obtain this phase composition were chosen based on a previous study,\textsuperscript{[2]} amounting to 790 °C and 15 min, respectively. Subsequently, while some strips were air-cooled, others were cold-rolled with a reduction of 70% to produce a severely deformed DP microstructure (cold-rolled or CR state). Here,

| Table 1. Chemical composition (in weight percent) of the steel RobuSal800 | C | Mn | Cr | Si | Mo | Al | B | Ti + V + Nb | P | S |
|---------------------------|---|----|----|----|----|----|---|--------------|---|---|
| Min                       | 0.07 | 1.60 | 0.50 | 0.15 | 0.1 | 0.02 | 0.0015 | 0.05 | – | – |
| Max                       | 0.15 | 2.1 | 1.0 | 0.3 | 0.4 | 0.06 | 0.006 | 0.012 | 0.02 | 0.01 |

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the cold rolling direction corresponded to the rolling direction of the initial material. Lastly, the CR strips were annealed at temperatures between 550 and 700 °C, with an interval of 50 °C lasting for 45 min, and air-cooled to room temperature to form a ferritic bimodal grain microstructure (bimodal microstructure or BM state).

### 2.2. Characterization of the Mechanical Properties

The tensile tests were carried out using an MTS 858 Table Top System to measure the material’s 0.2% YS, its ultimate tensile strength (UTS), its uniform elongation ($E_u$), and its elongation at fracture ($E_f$).

At least three dog bone specimens (see Figure 1a) with a gage length of 8 mm and a width of 3 mm were sampled from strips of each investigated state by wire erosion—the samples’ longer side corresponding to the previous rolling direction. The specimens were tested with a cross-head speed of 0.01 mm s$^{-1}$.

For the fatigue tensile tests, samples with a geometry similar to ASTM E466-07 were used (see Figure 1b). After the wire erosion, all samples’ surfaces were ground gradually with SiC papers with a grit range of P500, P1200, P2500, and P4000. The fatigue tests were carried out using the MTS 858 Table Top System at a frequency of 10 Hz and a load ratio of $R = 0.1$. For every S–N curve, i.e., the stress amplitude versus the corresponding number of cycles to failure, at least ten specimens were tested for up to $10^6$ cycles. Subsequently, they were defined as run outs.

### 2.3. Characterization of the Microstructure

To characterize the grain structure, the samples were cold mounted, mechanically ground, polished, etched in 3% nitric acid dissolved in ethanol (Nital etching), and then examined with the scanning electron microscope (SEM) Zeiss Ultra Plus. The SEM was operated at an acceleration voltage of 20 kV by using a secondary electron detector (SE).

Additionally, some specimens were exposed to LePera etchant to reveal the microstructures’ ratios of martensite and ferrite.\textsuperscript{[23]} The etched microstructure was examined under a Keyence VHX5000 digital optical microscope. Four lines were drawn in each analyzed image: two diagonal lines with a horizontal and a vertical line at their intersection. These lines were used to measure the total length of the grains of each phase, and thus, to determine the percentage portions of each phase. To do so, a minimum of 200 grains of each phase were considered.

Additionally, the microhardness was measured with a testing force of 0.098 N (HV0.01) in order to reveal differences in strength properties between phases as well as between fine and coarse grains.

A detailed characterization of the microstructure was performed with the transmission electron microscope (TEM) CM 200 from Philips. The TEM sample was prepared by grinding the material to a thickness of 150 μm and by a subsequent twin-jet electropolishing, utilizing a 5% perchloric acid in ethanol solution under an applied potential of 25 V at $-40$ °C until electron transparency was achieved. The analysis was carried out at a nominal acceleration voltage of 120 kV.

### 3. Results

The material’s mechanical properties in different microstructural states are summarized in Table 2. Typical stress-strain curves are presented in Figure 2.

The material with a homogeneous ferritic microstructure forms the basis for the evaluation of both the properties and microstructure of the material with a DP or bimodal microstructure. A relatively low difference between the UTS and YS values, i.e., 27%, indicates a low strain hardening capacity of the material with a homogeneous ferritic microstructure. Simultaneously, its high ductility, i.e., $E_u$ and $E_f$ values, at low strength values indicates good formability, which is necessary for the cold forming operation (HM in Table 2). Lastly, a slightly pronounced stress plateau indicates that the plastic deformation in this area of the stress–strain curve proceeds by the growth of Lüders bands (HM in Figure 2).\textsuperscript{[24]}

In contrast to the HM material, the material with a DP microstructure has a high UTS value and lower $E_u$ and $E_f$ values, whereas its YS value is almost equal to that of the HM material (DP in Table 2). The difference between the UTS and YS values of the DP material amounts to 49%, indicating a pronounced strain hardening during the plastic deformation.

The YS and UTS values of the material with a severely cold deformed DP microstructure are very similar, i.e., strain hardening is almost absent, and thus, the ductility is very low (CR in Table 2).

The mechanical properties of the bimodal microstructure, which results from annealing the material in a CR state, strongly depend on the annealing temperature: 1) Annealing at 550 °C reduces the strength properties by 20% (UTS), with a slight increase in $E_u$ and $E_f$ values in comparison with the CR state.

| Table 2. Mechanical properties of the material in different states. |
|----------------------|------------------|-----------------|-----------------|-----------------|-----------------|-----------------|
|                     | HM [MPa]  | DP [MPa]  | CR [MPa]  | BM [MPa]  | BM [MPa]  | BM [MPa]  |
|----------------------|------------------|------------------|------------------|------------------|------------------|------------------|
| YS [MPa]             | 352 ± 3          | 363 ± 1          | 1068 ± 3         | 912 ± 5          | 474 ± 12         | 441 ± 7          | 255 ± 2          |
| UTS [MPa]            | 485 ± 1          | 711 ± 2          | 1152 ± 2         | 917 ± 7          | 538 ± 13         | 503 ± 6          | 587 ± 9          |
| $E_u$ [%]            | 22 ± 1           | 15 ± 0           | 3 ± 0            | 4 ± 0            | 24 ± 1           | 25 ± 1           | 24 ± 1           |
| $E_f$ [%]            | 47 ± 1           | 30 ± 1           | 5 ± 0            | 7 ± 0            | 39 ± 1           | 38 ± 1           | 31 ± 3           |

\[550 \text{°C}\]

\[600 \text{°C}\]

\[650 \text{°C}\]

\[700 \text{°C}\]
2) Annealing at 600 °C dramatically changes the ratio between strength and ductility compared to the CR state. Although the YS and UTS values are bisected, they are still higher than the corresponding values of the material in the HM state. Simultaneously, the $E_u$ values are lower and the $E_t$ value is higher than those in the HM state. 3) Annealing at 650 °C decreases the strength to the material’s initial level. The ductility, i.e., the $E_u$ and $E_t$ values, increase only slightly in comparison with the corresponding values of the material annealed at 600 °C. 4) Annealing at 700 °C results in an unexpected increase in the UTS value and a decrease in YS and ductility values. Interestingly, these samples show no Lüders strain compared to other samples that were annealed after cold rolling (Figure 2).

The representative micrographs of the materials with a homogeneous ferritic microstructure (HM), a DP microstructure, and a severely cold deformed microstructure (CR) are presented in Figure 3.

In the HM state, the material consists of equiaxed ferrite grains with carbides distributed both within them as well as on the grain boundaries (Figure 3a). In the DP material, separate martensite islands can be observed within the ferrite. Some ferrite grains still contain separate carbides, which can rarely be observed on the grain boundaries (Figure 3b). Generally, the microstructure of the DP state comprises approximately 44% martensite and 56% ferrite, which differs only slightly from the intended ratio of 50% between the two phases (Figure 3d). The CR microstructure consists of wavy elongated ferrite grains that bend around slightly deformed martensite islands (Figure 3c).

The microstructures of the material in the BM state, depending on the annealing temperature, are presented in Figure 4.

Annealing at the lowest temperature investigated (550 °C) does not significantly influence the morphology of ferrite grains that were elongated during the cold rolling process: the banded microstructure is very pronounced. Simultaneously, martensite islands are transformed into fine ferrite grains (black rectangles),
which are still acicular and contain numerous fine carbides (Figure 4a). Increasing the annealing temperature results in the formation of a recrystallized bimodal microstructure with fine equiaxed grains (white circles) distributed between coarse grains (black circles) and partially oriented along the rolling direction. While the martensite transformed into fine grains, the ferrite grains were coarsened during the annealing process. A banded microstructure can still be observed here (Figure 4b). After annealing at a temperature of 650 °C, the bimodal microstructure is no longer banded and consists entirely of equiaxed nonoriented ferrite grains with a high amount of carbides, both in ferrite grains as well as on the grain boundaries (Figure 4c). The size difference between the individual ferrite grains is less pronounced than that in the bimodal microstructure formed at 600 °C. Instead, the microstructure resembles that of the material in the homogeneous state. Annealing at the highest temperature (700 °C) with subsequent air cooling generally leads to a further coarsening of the ferrite grains, while islands of a different phase can be observed in some triple junctions of the grain boundaries (white triangles in Figure 4d). Additional LePera etching of this sample revealed the presence of martensite at 9% (Figure 4e).

Thus, the bimodal microstructure formed during the annealing of the CR material at 600 °C demonstrates both the most pronounced inhomogeneity of grain-size distribution (Figure 4e) and a good combination of mechanical properties, i.e., simultaneously a high-level of strength and ductility (Table 2). Therefore, this state was chosen to characterize the fatigue property.

The fatigue properties of the materials with a homogenous ferritic grain microstructure, in a DP state and in the bimodal state after annealing at 600 °C, were characterized in the high cycle fatigue region. The corresponding S–N curves and the SEM pictures of the samples' fracture surfaces in the different states are presented in Figure 5.

The material exhibits the worst fatigue properties in the homogenous state (solid line). To withstand $10^6$ loading cycles, the maximum stress amplitude must not exceed 197 MPa. In the DP state (dashed line), however, the material has a significantly better fatigue strength than in its initial state: the maximum stress amplitude allowing to withstand $10^6$ loading cycles amounts to 239 MPa. It is noteworthy that the fatigue-life curve of the material in the DP state has a higher slope than that of the material with a ferritic microstructure. At a low cycle number, i.e. up to approximately $8 \times 10^7$ cycles, the samples with a
bimodal grain microstructure (dash-dot line) show an intermediate fatigue behavior in comparison with other material states, i.e., at a given cycle number, the corresponding stress is higher than that in the initial state and lower than that in the DP state. However, the maximum cycle number of $10^6$ can be maintained at 245 MPa, which is higher than that in both the other states, i.e., for the ferritic state with a homogenous microstructure and the DP state. Furthermore, the slope of the fatigue life curve equals that of the material in the homogenous state.

Independent of the material state, a crack developed at the samples’ edges and then propagated inside (indicated by dashed lines in Figure 5b–d) until a forced fracture, which manifests itself in the characteristic surface of a ductile fracture with dimples (not shown), occurs. The fracture surfaces of the samples in the homogenous state (Figure 5e) and the bimodal state (Figure 5g) are similar, although the latter one mainly consists of smaller dimples. The surface of the material in the DP state also exhibits cracks and cleavages (white circles in Figure 5f), which are not present in other states.

4. Discussion

During the intercritical annealing at 790 °C, a partial $\alpha \rightarrow \gamma$ transformation occurs. Both the ferrite grain boundaries and the interfaces between carbides and multipoint junctions exhibit the lowest energy for the nucleation of austenite grains. Simultaneously, carbides on the grain boundaries dissolve, allowing for carbon diffusion in the nucleated austenite grains and their chemical stabilization.[25] Therefore, only martensite islands that transformed from austenite during the air-hardening process are present on the ferrite grain boundaries, whereas carbides are rarely observed (Figure 3b). The transformed martensite serves as a strengthening hard phase, which leads to a significant improvement in the material’s strength and a moderate reduction in its ductility (Figure 2 and Table 2).

In accordance with previous studies of Azizi-Alizamini et al.[15] and Okitsu et al.,[26] it was found that the deformation during the cold rolling of DP steels mainly occurs in ferrite, resulting in fine lamellar grains elongated in the rolling direction and bent around hard and significantly less deformed martensite islands. Such a severely deformed microstructure exhibits excellent strength properties, as its UTS value is 21% higher than that of materials in an as-quenched state.[26] However, the extremely low $E_u$ and $E_t$ values do not allow for a practical application of those materials. After the second annealing step, the material’s bimodal microstructure, and therefore its mechanical properties, are both strongly influenced by the annealing temperature. The lowest investigated annealing temperature of 550 °C generally only results in the precipitation of carbides and tempering of

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**Figure 5.** S–N curves of the material with the homogeneous ferritic microstructure (HM), the DP microstructure, and the bimodal microstructure (BM) (a). Fracture surface of broken samples: b) and e) overall and magnified view for the HM state; c) and f) overall and magnified view for the DP state; d) and g) overall and magnified view for the BM state. Black rectangles—areas of magnified view; white circle—areas with observed cracks. ND—normal direction; TD—transversal direction. Arrows in (a) indicate the corresponding run outs.
the martensite and its subsequent transformation into less distorted ferrite, which retains the acicular form of the parent martensite. As Speich et al.\cite{12} and Tokizane et al.\cite{13} have shown, the recrystallization of as-quenched martensite in low-carbon steels occurs at temperatures near the corresponding \(A_{\text{Cl}}\) temperature, i.e., the initial temperature of the \(\alpha\rightarrow\gamma\) transformation. Simultaneously, a cold plastic deformation of the martensite facilitates the recrystallization process and shifts it toward lower temperatures and duration times.\cite{14} Martensite grains are not recrystallized at 550 °C, i.e., no formation of new equiaxed ferrite grains can be observed, which confirms a minor plastic deformation of martensite occurring during the cold rolling process. The wavy form of the severely deformed ferrite grains remains almost equal to that of the material in a CR state. These observations indicate that no recrystallization process occurs at this annealing temperature. For this reason, the mechanical properties change only slightly from the CR state (Figure 2 and Table 2). 

Increasing the annealing temperature to 600 °C results in the recrystallization of the martensite islands and formation of very fine ferrite grains. This fine grain size, i.e., the absence of any coarsening due to recrystallization, can be attributed to a pinning effect of the carbides that precipitated during the tempering of the martensite.\cite{15} Additionally, coarse ferrite grains can be observed, some of which are still elongated in the rolling direction. This was only true in areas that previously consisted of severely deformed ferrite grains, indicating a faster recrystallization. The difference to fine grains can also be explained by the absence of a pinning effect in the former martensite areas.\cite{16} Furthermore, although both cold deformed ferrite and martensite have nearly the same dislocation density, the dislocation distribution in martensite is homogeneous, whereas a cellular dislocation substructure is formed in cold deformed ferrite. Because these dislocation boundaries between the separate cells act as starting points for recrystallization, the recrystallization process starts earlier than that in martensite.\cite{17}

A detailed observation of the bimodal microstructure in the TEM is presented in Figure 6.

Both the fine (Figure 6a) and coarse grains (Figure 6b) of the bimodal microstructure are free from any dislocations, thus indicating a complete recovery. In contrast to the coarse grains, the fine grains contain numerous equiaxed carbides. The migration of their grain boundaries seems to be restricted in comparison with the free carbide grains, which is a result of the abovementioned pinning effect. Furthermore, due to the more heterogeneous distribution of the alloying elements, i.e., a higher percentage of fine carbides in fine grains, the individual fine grains exhibit a higher microhardness (200 ± 12 HV0.01) than the coarse grains (161 ± 1 HV0.01). According to Azizializamini et al.\cite{18}, the fine grains seem to act similarly to martensite in the DP microstructure. During the tensile test, however, the behavior of the material with the bimodal microstructure is more equivalent to the material in the homogenous state (see Figure 2), due to a pronounced Lüders strain and a relatively low strain hardening.

Furthermore, as the results of the fatigue tests show, the slopes of the fatigue curves of the materials in the ferritic homogeneous state and with the bimodal microstructure resemble each other. It can be assumed that the crack initiation and propagation mechanism is similar in both microstructures because of their mono-phase nature. In contrast to this, the curve of the DP microstructure is more sloped, and for high cycle numbers, i.e., upwards of \(10^6\) cycles, the maximum stress amplitude is lower than that for the bimodal microstructure. This fatigue curve can be explained by a pronounced crack initiation, predominantly on the boundaries of hard martensite, with a microhardness of 240 ± 8 HV0.01, and ductile ferrite phases, which have a microhardness of 173 ± 17 HV0.01, even at lower stress levels (see Figure 5).\cite{19} Contrary to the DP material, the crack initiation in mono-phase materials, both with normal and bimodal microstructures, seems to be delayed due to the more ductile nature of ferrite. Simultaneously, the fine grains, i.e., a high amount of grain boundaries, in the material with the bimodal microstructure pose an additional obstacle to crack propagation and thus improves the material’s fatigue strength compared to the initial state, in which only coarse grains are present.\cite{20}

Annealing at 650 °C results in a higher energy input and thus in a higher driving force of the recrystallization and in coalescence and coarsening of the precipitated carbides, thereby reducing their pinning effect on the grain growth. Therefore, the coarsening of ferrite grains that were formed from prior martensite islands leads to a less pronounced bimodality of the

![Figure 6. Bright-field images of the sample annealed at 600 °C: a) fine grains, b) coarse grains.](image)
microstructure, reduced strength, and slight improvement of the ductility as compared to the material annealed at 600 °C.

Although the A<sub>C1</sub>-temperature for the investigated steel is 750 °C,[2] the microstructure of the material annealed at 700 °C contains martensite (Figure 4d), and correspondingly, has a stress–strain curve that is typical for DP microstructures, i.e., no Lüders strain and a relatively high strain hardening. The CR microstructure has a higher amount of energy stored and more defects than the homogeneous ferritic microstructure with an A<sub>C1</sub>-temperature of 750 °C. According to numerous previous studies,[13,14] such a distorted microstructure facilitates the α→γ transformation and shifts the A<sub>C1</sub>-temperature to lower values, resulting in a formation of austenite even below the expected A<sub>C1</sub>-temperature of 750 °C.

Based on these results, the appropriate parameters for a thermo-mechanical treatment, i.e., intercritical annealing at 790 °C, cold rolling with a rolling reduction of 70%, and subsequent annealing at 600 °C, create a bimodal grain-size distribution in the mono-phase ferritic microstructure. Air-hardening steel with the bimodal microstructure exhibits superior strength properties and nearly the same ductility (E<sub>r</sub>-values) in comparison with a conventionally CR and annealed microstructure. With regard to high cycle fatigue, which is of high importance for automotive components, the sheet with a bimodal microstructure has the best fatigue strength as compared to both the traditional DP microstructure and the CR and annealed microstructure.

Varying the martensite fraction through the intercritical annealing temperature, and thus, changing the fraction of fine grains after a subsequent cold rolling and annealing process, presents a relatively simple way to adjust the ratio between the two grain fractions and therefore the strength and ductility properties of a sheet with the bimodal microstructure.[18] Hence, the aim of the ongoing research is to investigate the influence of a martensite fraction in the DP state on the resulting properties of materials with the bimodal microstructure.

5. Conclusion

In summary, low-alloy steel with the bimodal microstructure presents a new class of materials with high strength and ductility properties. The results of this study confirm both the feasibility and the high controllability of the investigated thermo-mechanical process, which enables the formation of a bimodal grain-size distribution in air-hardening steel. The most important processing steps are intercritical annealing between the A<sub>C1</sub>- and A<sub>C3</sub>-temperatures, e.g., 790 °C; cold rolling with a high rolling reduction, i.e., 70%; and subsequent annealing of the material below the A<sub>C1</sub>-temperature. The temperature for the last annealing step strongly influences the recrystallization process of the severely deformed ferrite grains and the martensite islands, and thus the microstructure and resulting mechanical properties. The microstructure with the most pronounced bimodal grain-size distribution, obtained by annealing at 600 °C, exhibits a superior strength (UTS value of 567 MPa vs 485 MPa) and nearly the same uniform elongation (E<sub>r</sub>-value of 23% vs 22%) compared to the material with the homogeneous ferritic microstructure. While the fatigue curve of the material with the bimodal microstructure, i.e., the behavior at fatigue loading, is similar to that of the material with the HM, the corresponding maximum stress amplitudes at a given number of cycles are higher at approximately 48 MPa. In addition, the maximum stress amplitude of the bimodal mono-phase microstructure in a high cycle regime (at 10<sup>7</sup> cycles) is even better (245 MPa vs 239 MPa) than that of the material with the DP microstructure.

The influence of the martensite fraction before cold rolling and annealing, i.e., the influence of the temperature during intercritical annealing, is another factor to consider in order to achieve the desired combination of strength and ductility in an air-hardening steel with the bimodal microstructure and will the topic of further research.

Conflict of Interest

The authors declare no conflict of interest.

Keywords

bimodal microstructure, dual-phase microstructure, ductility, low-alloy air-hardening steel, tensile and fatigue strength, thin sheet

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