Investigations on metallurgical and mechanical properties of resistant spot welded dissimilar joints between a high manganese steel and a low alloyed steel grade

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Abstract. The present investigation gives a brief overview into the correlation of the amount of molten ferrite and metallurgical as well as mechanical properties of resistance spot welded austenitic-ferritic joints. The fusion rate describes the ferritic quantity in the weld metal and provides a reliable reference to the integrity of dissimilar joints. Metallurgical effects due to varied fusion rates are shown by element distribution, micro hardness measurement and phase analysis. In addition, this work states specific fusion rates which lead to a degradation of the mechanical properties and can be used as a basis for application guidelines. This can be seen in certain tensile tests with single and multiple spot specimens.

1 Introduction
The use of new lightweight materials for the body in white structure is an important aspect regarding the weight reduction targets of the automotive industry. In order to reduce emissions and manufacturing costs the development of high strength steels is an important factor. The implementation of these materials in future electric mobility concepts proves to be even more significant. High manganese steels with their full austenitic structure offer a special property profile better than conventional steel grades. The TWIP Effect (Twinning Induced Plasticity) allows for extraordinary ductility combined with tensile strength higher than 1.000 MPa [1].

Contrary to conventional austenitic FeCrNi based steels FeMn alloys contain high amounts of carbon, silicon, aluminum and of course manganese for stabilizing the austenitic phase [2]. The weldability of these grades has become an important aspect for the processing sector. Especially heterogeneous spot welded joints between austenitic and ferritic sheets represent a great challenge for the automotive production. It is well known that the chemical composition of the base material determines the stacking fault energy \( \Gamma \) (SFE) and thus ultimately the hardening mechanism [3]. This aspect should therefore also be transferable to the common weld metal. Previous investigations have shown the presence of brittle martensitic phases which often lead to an early failure [4]. Consequently, it is necessary to estimate the influence of this embrittlement on the spot integrity of dissimilar joints.

As illustrated in figure 1, dissimilar spot welds of austenitic and ferritic steels show a deviating behavior in terms of nugget formation. Due to the higher thermal conductivity of low alloyed grades with their body centered cubic lattice (bcc) nuggets grow predominantly in direction of the austenitic
steel and form asymmetrical weld metals as well as heat affected zones. This goes also along with unequal crystallization rates within the nugget.

![Figure 1. Cross section of an austenitic-ferritic spot welded joint.](image)

### 2 Materials and Experimental Setup

The used austenitic base material is an uncoated innovative FeMn steel with a tensile strength of $R_m \geq 950$ MPa and an elongation of $A_{80} \geq 41\%$. The ferritic joining partner is a widespread zinc coated micro alloyed steel. The tensile strength is $R_m \geq 410$ MPa and the elongation is $A_{80} \geq 23\%$. Both sheets are used in a thickness of 1.5 mm. The chemical composition is presented in table 1.

| Steel    | C   | Mn  | Al  | Si  | Cr  | Ni  | Ti  | Nb  | V    | Fe & other |
|----------|-----|-----|-----|-----|-----|-----|-----|-----|------|------------|
| FeMn     | 0.64| 17.0| 1.8 | 0.05| 0.01| 0.007| 0.075| 0.0002| 0.002| bal.       |
| Ferritic | 0.06| 0.53| 0.037| 0.222| 0.088| 0.046| 0.003| 0.014| 0.002| bal.       |

To characterize the metallurgical properties specimens according to SEP 1220-2 [5] are used. For the welding current range and the element distribution the two coupons of dimension 45 mm x 45 mm with 40 mm overlap are joined by one single spot without shunt. Micro hardness measurement, phase analysis and destructive testing of mechanical properties is performed under quasi-static load with shear tensile and cross tensile specimens according to SEP 1220-2 [5] as well. In addition a multiple spot specimen is designed as illustrated in figure 2. This certain geometry should provide an impression over the load transfer capacity between the spot welds. In all cases the external load is applied by a static material testing machine Zwick Z100 with a testing speed of 1/10 mm/s.

![Figure 2. Multiple spot specimen (dimensions in mm).](image)

For the element distribution the Electron Probe Microanalyzer (EPMA) JEOL JXA 880 L is used. The analysis is performed in a scan field of 900 x 900 µm with a step interval of 3 µm. The scan field is located at the notch tip of the spot weld between the sheets. This area is usually exposed to a distinctive stress under shear load [6].
The micro hardness measurement (HV 0.1) by UCI method (Ultra Sonic Impedance) is performed in the same area as the element distribution and done by the automated hardness scanner BAQ UT 200. The step interval is 100 µm.

For the phase analysis by EBSD (Electron Backscatter Diffraction) the DualBeam FEI Scios equipped with AMETEK-EDAX analytic is applied. The acceleration voltage is 20 kV and for the step size a distance of 200 nm is chosen.

3 Definition and adjustment of fusion rates

This investigation into dissimilar γ-α-joints defines the fusion rate Φ as the quantity of molten ferrite referring to the entire weld metal. According to equation (1), the determination of Φ is made in a polished cross section by forming the area ratios of molten ferrite and austenite as illustrated below in figure 3.

\[ \Phi = \frac{A_{\text{Ferrite}}}{A_{\text{Ferrite}} + A_{\text{Austenite}}} \]  

Figure 3. Fusion rate determination by area ratios in a cross section of a dissimilar γ-α-joint.

For the following investigations two fusion rates are considered. The adjustment is made by applying commercial CrNi process tapes in a thickness of 0.15 mm between the electrode cap and the sheet metal as shown in figure 4. In the first case one tape is applied on the austenitic side of the joint and in the second case two tapes are applied on the ferritic side. With this procedure two certain fusion rates of \( \Phi = 13 \pm 3 \% \) and \( \Phi = 50 \pm 3 \% \) are achieved. The welding parameters are shown in table 2.

Figure 4. Adjustment of fusion rates with process tapes (left: \( \Phi = 13 \% \), right: \( \Phi = 50 \% \)).
In order to evaluate an influence of the fusion rate on the mechanical properties of resistance spot welds the welding parameters are set in a certain way to obtain a uniform nugget diameter of \(5 \pm 0.1\) mm. This ensures a comparability and provides more information about the joint properties.

| Table 2. Welding parameters according to fusion rate. |
|---------------------------------|---|---|---|---|---|---|---|
| \(\Phi\) [%] | Electrode Force [kN] | Squeeze Time [ms] | Prepulse Time [ms] | Prepulse Current [kA] | Pause Time [ms] | Welding Time [ms] | Weld Current [kA] | Hold Time [ms] |
| 13 | 4 | 500 | 200 | 4 | 200 | 400 | 5.8 | 300 |
| 50 | 4 | 500 | 200 | 4 | 200 | 400 | 5.0 | 300 |

4 Metallurgical properties

During welding the chemical composition of dissimilar metal welds in the common fusion zone must be expected to differ from that of the base materials. In the present study two certain fusion rates of \(\Phi = 13\%\) and \(\Phi = 50\%\) are considered. Therefore, it is necessary to first characterize the distribution of the main alloying elements and then measure the resulting micro hardness level in areas which determine the integrity of the joint. For a better understanding of microstructural mechanism which lead to an early failure of the joint a phase analysis is carried out. The hardness measurement as well as the phase analysis is done under as-welded conditions and in certain prestrained states.

4.1 Element distribution

Figure 5 shows a quantitative course of Fe, Mn and Al in sheet thickness direction in the center of the nugget at a maximum fusion rate of \(\Phi = 50\%\). At the boundaries \(\gamma\)-Fe (I) \(\rightarrow\) weld metal (II) and weld metal (II) \(\rightarrow\) \(\alpha\)-Fe (III) increases in Fe concentration occur as expected. The Mn content of the weld metal is significantly reduced due to the fusion process. Within the weld metal (II), however, no significant gradient in sheet thickness direction can be detected on a macroscopic level which is representative for an even course. Only the detected cavity in the center influences the course. For Al there is no difference in concentration at the selected resolution, neither at the transitions nor within the weld metal.

![Figure 5. Cross section with EDX line scan in the center of the nugget \((\Phi = 50\%)\).](image)
Figure 6. Distribution of alloying elements Fe \((1 + 2)\), Mn \((3 + 4)\) and Al \((5 + 6)\) at the notch tip of the spot weld (Micro analysis; left: \(\Phi = 13\\%\); right: \(\Phi = 50\\%\)).

The analysis is carried out in the notch tip of the joint and shows parts of the base materials, the heat affected zones and the weld metal. According to [6], the area shown in the figure represents the most stressed area for resistance spot welds and is therefore of particular interest.
It is obvious that the chemical compositions of the weld metals differ significantly from those of the base materials, which were displayed in table 1. Part 1 and 2 of figure 6 show the Fe distribution for the areas of heat affected zone (HAZ) and weld metal (WM) of the γ-α-joint separated by the solidification line (SL). While in part 1 at Φ = 13 % the Fe content in the HAZ of the FeMn steel (HAZ γ-Fe) is about 80 % on average, values of 87 % can be detected in the weld metal. In part 2 at Φ = 50 % the Fe content in the weld metal is even above 90 %. This is accompanied by a change in the proportions of the main alloying elements Mn, Al and C. With an increasing fusion rate the concentration of the main alloying elements is reduced in the weld metal.

In the analysis of Mn clear differences between the two fusion rates can be seen. The average Mn content of the austenite stabilizing element in the base material is about 17 wt.%. When comparing the Mn contents of the weld metals (WM) from part 3 (Φ = 13 %) and part 4 (Φ = 50 %) a significant difference in concentrations become apparent as a result of the increasing fusion rate. While large areas in the weld metal (WM) are between 11 - 18 wt.% of Mn at a low fusion rate the Mn content decreases to values between 7 - 11 wt.% in some cases even below this at a high fusion rate.

According to the segregation behavior of Mn figure 7 shows certain areas of Mn concentration. At a high fusion rate areas of Mn enrichment and depletion lie directly next to each other. This observation is confirmed by the investigations of [7] and [8] in which Mn follows the solidification structure. In interdendritic spaces Mn is enriched while in the dendrite itself Mn depletion occurs.

These differences in the concentration of Mn on microstructural level can lead to different hardening mechanisms in the respective areas. An important aspect in this context is the SFE which is dependent on the alloying elements. In the literature are different values for the SFE in which a change between hardening mechanisms take place [9-12]. According to Jin et al. [13] the base material used in the tests has a SFE of $\Gamma = 33$ mJ/m$^2$ and is therefore a full austenitic TWIP steel. Taking into account the research results of Song et al. [14] the limit between the formation of mechanical twins and strain induced martensite is $\Gamma = 20$ mJ/m$^2$. Accordingly, following the mechanism maps of Song et al. [14] a fusion rate of $\Phi = 13$ % would result in a weld metal which is still in range of the TWIP effect. At the higher fusion rate of $\Phi = 50$ % the SFE is below $\Gamma = 20$ mJ/m$^2$. The resulting hardening mechanism in the weld metal would therefore be the TRIP effect.

The Al concentration of the base material is about 1.5 wt.% (table 1). At $\Phi = 13$ % the increasing ferrite content in part 5 shows that the Al content in the weld metal (WM) decreases to values between 0.7 - 1.1 wt.%. At a fusion rate of $\Phi = 50$ % Al is further reduced to values between 0.3 - 0.9 wt.% as shown in field 6. The investigations of Senk et al. [15] show that the element is subject to a certain solidification kinetics. As a result of this, Al is deposited anticyclically in the microstructure during phase formation. In areas of Mn depletion, therefore, Al enrichment takes place and vice versa. In combination with the concentration differences of Mn the behavior shown schematically in figure 8
occurs. As a result, an increased Al content in the dendrite can primarily lead to a formation of ferrite or martensite.

Figure 8. Anticyclically solidification behavior of Mn and Al in the weld metal.

4.2 Micro hardness
The previous observation has shown that depending on the fusion rate an increased ferrite content in the weld metal consequently leads to a chemical composition which differs from the one of the base materials. According to this result, the hardening behavior of the weld metal and the surrounding areas is to be further investigated. For this purpose, conventional shear tensile test specimens according to [5] are subjected to different external loads after welding so that discrete strain states are created. The samples are pretrained in the quasi-static range to certain force levels. Table 3 shows the unstrained initial state (F₀) and three further values at 4 kN (F₁), 8 kN (F₂) and F₃. The state F₃ is just below the maximum force of the welded specimen. It was determined in preliminary tests and specified as 0.9 · Fₘₐₓ. It can be seen that different forces are achieved at this state. While the rupture of the specimens with a large quantity of ferrite (Φ = 50 %) already starts at Fₘₐₓ = 9.8 kN, the material separation at a lower fusion rate (Φ = 13 %) begins at about Fₘₐₓ = 12.6 kN. This behavior according to the mechanical properties is discussed later in section 5. Subsequently, an UCI hardness measurement is performed on both unstrained and pretrained welded joints as illustrated in figure 9.

Table 3. Prestrained conditions of UCI-specimens for hardness measurements.

| State | Fusion rate | Fusion rate |
|-------|-------------|-------------|
|       | Φ = 13 %    | Φ = 50 %    |
| F₀    | 0 kN        | 0 kN        |
| F₁    | 4 kN        | 4 kN        |
| F₂    | 8 kN        | 8 kN        |
| F₃ = 0.9 · Fₘₐₓ | 11.3 kN    | 8.8 kN      |
Figure 9. Micro hardness (UCI) at the notch tip of the joint (left: $\Phi = 13\%$; right: $\Phi = 50\%$).
The comparison of the UCI measurements of both fusion rates show clear differences in the hardness distribution in the weld metal with increasing prestraining states. While the hardness of the weld metal with a low fusion rate of $\Phi = 13\%$ for the state $F_1$ is approximately at the base material level of the FeMn steel, a brittle, seam-like area forms along the austenitic-ferritic fusion line at $\Phi = 50\%$. In addition, hardness differences can be seen directly in the weld metal along the solidification structure.

In the prestrained state $F_1$, a homogeneous hardness level is still evident at $\Phi = 13\%$. No hardness peaks can be detected at the solidification line, nor are pronounced gradients formed along the dendritic solidification structure. For this condition it can be stated that no deformation induced phase transformation in the microstructure of the weld metal has occurred, which can be verified by means of hardness measurement. This is due to the fact that the high proportion of austenite stabilizing elements in the weld metal, in addition to the usual gliding of dislocations, primarily supports the formation of mechanical twins. However, before the TWIP effect is initiated in the material, a critical dislocation density must be exceeded. According to Fonstein [16] and Remy [17], this value depends on the SFE and other factors such as temperature and should therefore be considered for each alloying concept. In the state $F_1$ the first signs of hardening can already be seen in the weld metal at $\Phi = 50\%$. No change can be documented in the seam along the fusion line which is present from the beginning. On the other hand, an increase in hardness is observed in the weld metal in the stressed areas around the notch tip of the joining plane.

The state $F_2$ for $\Phi = 13\%$ differs only slightly from case $F_1$. A slight increase in hardness can be measured in the notch tip of the joining plane. A hardness peak can be detected in one measuring point. In contrast, the measurement with a high fusion rate of $\Phi = 50\%$ shows a significant increase in hardness in the weld metal. Large areas are between 340 HV and values of 460 HV. Directly in the notch tip even values between 520 HV and 700 HV are achieved. The evidence of such brittle microstructure suggests at this point that, in addition to the increase in dislocation density, deformation induced martensite has occurred due to the high stress. According to the high quantity of ferrite in the weld metal, the SFE decreases and the originally full austenitic phase is no longer sufficiently stabilized. Metastable austenite is formed which transforms under stress. To what extent this transformation is $\varepsilon$ or $\alpha'$ martensite cannot be further determined by the UCI measurement. If the results of Schumann [18] are taken into account, the area in which the indirect phase transformation of $\gamma$ (fcc) $\rightarrow$ $\varepsilon$ (hcp) $\rightarrow$ $\alpha'$ (bcc) takes place is in a concentration range between 10 and 15 wt.% Mn.

In the last state considered ($F_3$) the hardness measurement is carried out after a load immediately before failure of the specimens. At $\Phi = 50\%$ a hardness between 500 and 700 HV can be determined in the area in front of the notch tip at a shear tensile force of $F_3 = 8.8$ kN. This area shows clear signs of embrittlement. Also at 0.9 · $F_{\text{max}}$ an increase in hardness at a similar point in the weld metal can be demonstrated at $\Phi = 13\%$. In absolute terms this condition is reached at $F_3 = 11.3$ kN. This delayed hardening can be explained by the dynamic Hall-Petch effect. In which the twin density in the austenite grains increases with increasing elongation, thus reducing the free path between the twins. The distances available for dislocation glide become shorter until finally a blockage occurs [16], [17].

### 4.3 Phase analysis

In the investigations carried out so far, differences in the concentrations of the main alloying elements as well as in the mechanical properties on a microstructural level in the highly stressed areas of the nugget could be demonstrated. The question of whether hardening is due to an increase in dislocation or twin density or martensite formation will be investigated by EBSD analysis. A decisive influencing factor is the chemical composition of the weld metal. Based on this, the two fusion rates of the states $F_0 = 0$ kN and $F_3 = 0.9 · F_{\text{max}}$ shall be considered in accordance with table 3. These four specimens are used to investigate how the weld metal behaves during mechanical loading. Furthermore, the question is to be answered, which microstructure modifications support an early failure and whether a change of the hardening mechanisms occurs.

EBSD analysis can be used to determine the existing phases and their orientation. In addition to the thermally induced phase formation during welding, alloys with a high Mn content can also undergo a
phase transformation induced by deformation. Thus, lattice shear during martensite formation can be facilitated by an externally applied load. As a consequence it is difficult to characterize the resulting microstructure and its mechanical properties, since aspects such as dislocation and twin density as well as SFE play a role. These mechanisms can influence each other during the loading process.

Figure 10 shows the phase analysis for the state $F_0$ of the two fusion rates $\Phi = 13\%$ and $\Phi = 50\%$. In this case, the scan is carried out on a part of the fusion line near the notch tip. For $\Phi = 13\%$, the left part of the figure shows a clear separation of both phases from each other along the solidification line. From the ferritic HAZ only a small amount of ferrite has grown into the weld metal. In the right part of figure 10, however, it is clearly visible that the proportion of ferrite increases at a higher fusion rate. In this area the distribution is island-like within the austenite phase.

![Figure 10. Phase analysis at the notch tip of the joining plane at $F_0 = 0$ kN (left: $\Phi = 13\%$; right: $\Phi = 50\%$).](image)

Figure 11 shows the scan field in the center of the nugget. The phase analysis of the two fusion rates in reference condition $F_0$ show clear differences in the proportion of ferrite in the weld metal. The left part of the figure shows a full austenitic microstructure for $\Phi = 13\%$. No modification of the $\alpha$-phase can be detected at this point. In contrast, in the right part of the figure at $\Phi = 50\%$, it can be seen that the proportion of ferrite increases significantly. The formation of ferrite follows the solidification structure and takes place for the most part in the dendrites.

![Figure 11. Phase analysis in the center of the weld metal at $F_0 = 0$ kN (left: $\Phi = 13\%$; right: $\Phi = 50\%$).](image)
Due to the high crystallographic similarity of ferrite and martensite, the detection and difference of both phases is not trivial, even with electron diffraction based methods. As in the present case, the conclusion that martensite is necessarily involved cannot be drawn directly from the detection of $\alpha$-Fe in the weld metal. Furthermore, the question about the origin (deformation or thermally induced) and the type ($\alpha'$- or $\epsilon$-martensite) remains open.

The misorientation relationships between the austenitic and martensitic modifications are subject to discrete laws that describe a parallelism between planes and directions of the face-centered cubic austenitic (fcc) and tetragonal body-centered martensitic (bct) cell as shown in table 4. On the basis of the misorientation distributions between the different phases measured by EBSD analysis, a pronounced maximum between 42° and 46° can be seen in figure 12, which essentially corresponds to the range described by the misorientation relationships listed in table 4. This supports the assumption that the occurring phase is martensite.

Table 4. Misorientation relationships between austenite and martensite [19].

| Misorientation relationship   | Misorientation angle | Reference |
|------------------------------|----------------------|-----------|
| Kurdjumow-Sachs              | 42,85°               | [20]      |
| Greninger-Troiano            | 44,23°               | [21]      |
| Nishiyama-Wassermann         | 45,98°               | [22]      |

Furthermore, the statement can be derived that the transformation into a cubic cell has taken place. The detected phase is therefore $\alpha'$-martensite, because $\epsilon$-martensite has a hexagonal cell and could not be detected in the weld metal.

![Figure 12. Misorientation distribution between the fcc and bct phases.](image)

In the following investigation two metallurgical states are subjected to an external load. Before the phase analysis, the samples are prestrained to $F_3 = 0.9 \cdot F_{\text{max}}$ as described in the previous section. For $\Phi = 13$ % the EBSD analysis is shown in figure 13. The image quality (IQ) map shows that deformation twins in the austenitic phase have occurred due to stress. Only in a few spots martensitic formation can be detected along the twin structures. The weld metal consists almost entirely of austenite.
Figure 13. IQ map (left) and phase analysis (right) in the weld metal of a prestrained specimen ($\phi = 13\%$; $F = 0.9 \cdot F_{\text{max}}$).

A similar observation can be made in figure 14 along the joining plane between ferrite and austenite. The highly stressed area in the weld metal shows also a significant increase in twin density. This is an indicator that the SFE of the microstructure in the fusion zone is still in range of the TWIP effect as the primary hardening mechanism.

Figure 14. IQ map (left) and phase analysis (right) at the solidification line of a prestrained specimen ($\phi = 13\%$; $F = 0.9 \cdot F_{\text{max}}$).

For the higher fusion rate figure 15 shows that several effects are superimposed during loading. In addition to the thermally formed $\alpha'$-martensite, there is also deformation induced martensite. Due to the energetically better conditions, this deformation induced martensite is deposited primarily along the deformation twins within the grains [23]. According to figure 16, both types of this microstructure can be detected at the crack tip of the prestrained sample. From this it can be deduced that this effect supports the failure of the joint. The thermal martensite resulting from the welding process, in combination with the deformation induced martensite and twins resulting from the loading, can cause a ductility loss of the structure that crack formation occurs along the microstructural interface between areas of different mechanical properties.
**Figure 15.** IQ map (left) and phase analysis (right) in the weld metal of a prestrained specimen ($\Phi = 50\%$; $F = 0.9 \cdot F_{\text{max}}$).

**Figure 16.** IQ map (left) as well as SEM illustration (above) and phase analysis (right) at the notch tip of a crack of a prestrained specimen ($\Phi = 50\%$; $F = 0.9 \cdot F_{\text{max}}$).
5 Mechanical properties

In the previous section, metallurgical aspects were examined with regard to different fusion rates. It was demonstrated what resulting weld metal can be expected with an increasing quantity of ferrite. In this part, the effects of material degradation on the load capacity are examined. In order to determine the mechanical properties of the resistance spot welded joints shear and cross tensile tests under quasi-static load are carried out.

5.1 Shear and cross tensile properties

According to the joint properties, figure 17 represents a comparison of different fusion rates under shear and cross tensile load. For both cases the results show a reduced load capacity with an increasing fusion rate. Although the shear tensile force decreases from 12.6 kN at $\Phi = 13\%$ to 9.8 kN at $\Phi = 50\%$ it only provides limited information about the load capacity. Compared to the determined energy under shear tensile stress, the force is only influenced on a lower level. In contrast, the difference of the energy absorption between both fusion rates is about 50\% and thus represents a more significant degradation of the investigated spot welds according to the field of application.

The cross tensile test shows a similar behavior. With an increasing fusion rate the cross tensile force decreases from 4.6 kN at $\Phi = 13\%$ to 1 kN at $\Phi = 50\%$. As mentioned above the energy absorption capacity represents more reliable parameter to evaluate the mechanical properties. In the present case this means a decrease of 95\% from an average energy of 44.4 J at $\Phi = 13\%$ to 2.1 J at $\Phi = 50\%$.

Furthermore, this investigation shows that the ductility of a spot weld is subjected to degradation. Figure 18 illustrates cross sections of failed spot welds after mechanical testing. Comparing the two fusion rates it can be concluded that the failure behavior of the specimen with a fusion rate of $\Phi = 13\%$ is much more ductile due to higher plastic deformation of the weld metal and the surrounding areas. The other specimen with a fusion rate of $\Phi = 50\%$ shows in contrast a brittle failure through the weld metal along the joining plane.

![Figure 17. Shear tensile test (left) and cross tensile test (right) in dependence of $\Phi$.](image-url)
Figure 18. Cross sections of failed spot welds after shear tensile testing.

A similar behavior can be observed in the cross sections of figure 19. For $\Phi = 13\%$ the failed spot weld shows a pull-out failure with a high degree of deformation due to the detected necking. For the higher fusion rate of $\Phi = 50\%$ a brittle separation of the two joining partners through the weld metal takes place. Neither in the nugget nor in the HAZ or the BM an essential plastic deformation can be determined.

Figure 19. Cross sections of failed spot welds after cross tensile testing.

5.2 Multiple spot geometry

In order to reduce the degree of abstraction between a standardized shear test specimen and a resistance spot welded assembly, a multiple spot geometry as presented in figure 2 was used in the following experiment. The focus of the analysis is primarily on the influence of the fusion rate on the load transfer behavior between the individual spot welds. The results of the shear tensile tests are shown in figure 20. Each curve is a representative case for $\Phi = 13\%$ and $\Phi = 50\%$ out of $n = 5$ experiments.
As shown in the previous section, due to a low fusion rate a high ductility of the weld metal enables a sufficiently large relative movement between the parts to be joined under shear tensile stress. This aspect leads to an even loading of all weld spots. For $\Phi = 13\%$ assembly failure does not occur in the form of individual, successive failure, but occurs collectively after the tensile strength is exceeded. In contrast, the increased ferrite content in the weld metal of $\Phi = 50\%$ shows that the failure passes through seven discrete stages as follows:

1. reaching $F_{\text{max}}$ and crack initiation in spot I,
2. failure of spot I,
3. load transfer and crack initiation in spot II,
4. failure of spot II,
5. load transfer and crack initiation in spot III,
6. failure of spot III and
7. complete failure.

This results in the so-called "zipper effect" at higher fusion rates. Although a load transfer to adjacent spot welds takes place, but only to a small extent into the base material, the first spot weld fails early. As a consequence of that the subsequent spot weld is also subjected to maximum stress until a brittle fracture occurs. This behavior continues until the components are finally separated. As illustrated in the graph of figure 20 the energy absorption capacity is.

**Figure 20.** Shear tensile test of multiple spot geometry in dependence of $\Phi$.

6 Conclusion

The experimental results confirmed that the ferrite amount in the weld nugget influences the mechanical properties. Furthermore, the here defined fusion rate $\Phi$ offers an appropriate possibility to characterize the joint integrity of dissimilar spot welds of FeMn and low alloyed steels. With an increasing fusion rate respectively quantity of ferrite in the weld metal the SFE shifts to values which allow other hardening mechanism such as TRIP-Effect to take place. As a result brittle martensitic phases occur besides dislocation glide and the formation of mechanical twins under external load. In addition to the deformation induced martensite after prestraining, the specimens martensite formed during welding can
be detected. The consequence is a degradation of the mechanical properties and a brittle failure of the joint with no essential plastic deformation. This behavior was demonstrated on standardized as well as modified multiple spot specimens and must be further investigated by SEM of fracture surface topography. From these results it can be concluded that the fusion rate $\Phi$ must be adequately taken into account when processing these materials.

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