

OPTIMIZATION OF SINGLE PASS WELDING OF HIGH CARBON BAINITIC STEEL

Peter Kirbiš, Tomaž Vuherer, Tomaž Irgolič, Ivan Anžel

Key words: *high carbon bainitic steel, heat input, microstructure, welding*

Abstract:

Single pass welds have been made on a newly developed high carbon bainitic steel with the process of A-Tig welding. Despite the high carbon content, welding could be performed without the requirement of preheating, as a fully carbide free bainitic microstructure forms within the weld from the initially spheroidized condition. Therefore the welds are not susceptible to cold cracking upon cooling to room temperature. The single pass welds were performed by varying the heat input through changes in the welding speed, all the welds produced show a greatly uniform hardness of about 640HV1. This strongly suggests that the precipitates required for bainite formation in the new steel grade are only slightly influenced by the spheroidization annealing.

1. INTRODUCTION

Low temperature carbide free bainitic steels are known for their superior combination of mechanical properties, they are capable of achieving tensile strengths up to 2500MPa, while maintaining reasonable values of ductility and fracture toughness [1], [2]. Their mechanical properties are derived from the very fine scale of bainitic ferrite plates embedded within a matrix of retained austenite. The latter undergoes a stress/strain induced transformation into martensite and gives rise to work hardening, known as the TRIP (transformation induced plasticity) effect [3]. In order to obtain fine bainitic ferrite subunits with a thickness of only 20-50 nm requires low transformation temperatures between 200 and 300°C. These unfortunately coincide with prolonged isothermal holding times, which depending on the steels composition may last from 8 hours or as long as 10 days [4]. The bainite reaction is thought to be of a displacive character, therefore the rate limiting factor is the formation of bainite nuclei [5]. This requires the paraequilibrium partitioning of carbon [6] and consequently the kinetics are slowed down exponentially with decreasing temperatures [7]. To ensure the low transformation temperatures and a carbide free bainitic microstructure substantial quantities of carbon and about 1.5% Si are present in the steels compositions [8]. The amount of carbon is rarely below 0.8 w%, resulting in high carbon equivalents.

When such steels are welded they are susceptible to cold cracking [9] due to the formation brittle martensite within the weld and reaustenitized heat affected zone (HAZ). Therefore they need to be preheated, to a temperature which usually is close to the designated temperature of bainite formation. Prolonged isothermal holding times are then applied to regenerate the bainitic microstructure within the weld and reaustenitized region [10]. As this treatments can last from several hours to days, they are performed at a slightly higher temperature, then the initial austempering. The application of rotary impacted trailed welding is also known to accelerate the regeneration [11]. It has been shown that the strength of the weld can be comparable to that of the base material as welds with tensile strengths of up to 2100 MPa (92% of the strength of base materials), have been obtained [12]. On the other hand the reaustenitized region is somewhat more prone to failure due to grain growth. It is known that larger grains form smaller fractions of bainite [13], thus requiring the heat inputs to be tightly controlled [14].

A new grade of low temperature bainitic steels with exceptionally rapid transformation kinetics at temperatures below 200°C, has been recently developed. The mechanism through which this has been achieved is by the introduction of numerous Nano scale precipitates. This precipitates develop carbon depleted zones, which upon cooling act as potential nucleation sites for the bainitic

ferrite subunits. With such an approach kinetically suitable conditions for the bainite reaction are obtained at a high temperature and it thus proceeds rapidly upon cooling below B_s . By virtue of their specific trait this newly developed alloys have hence been named kinetically activated bainite (KAB) steels. In these steels fully bainitic microstructures can be easily obtained during air cooling, without the need for any isothermal holding procedures. The weldability of the new steel was subject to previous investigations, and it has been observed that the bainitic microstructure regenerates within the weld.

For certain applications the steel needs to be soft annealed and machined if the final parts are to be then welded it is preferable to perform welding prior to final austempering. The whole welded part can then be heat treated and a uniform hardness obtained throughout the whole section. The aim of this paper is to optimize the welding parameters during single pass welding of soft annealed steel plates, and to investigate their effect upon the microstructure. The latter is of interest as the precipitates which promote the formation of bainite might have coarsened during the soft annealing process.

2. MATERIALS AND METHODS

The alloy was produced using master alloys and pure components, induction melted in vacuum and mold cast under a protective atmosphere of pure argon. The alloys chemical composition can be seen in table 1.

Table 1. Steel composition (in w %).

	i	n	o	r		u	l	g		i	i	r		
.8	SUM 8.2%													

Certain elements have been introduced from the master alloys used or are present as unavoidable impurities. The ingot was then homogenized at 1200°C for 2 days, followed by hot rolling at 1050°C. During the latter the thickness was reduced from the initial 18 mm to a final thickness of 6mm, followed by air cooling. Upon cooling to room temperature the newly developed steel forms a fully bainitic microstructure, comprised of fine bainitic ferrite sheaves, separated by fine blocks of retained austenite, as shown in Fig. 1.

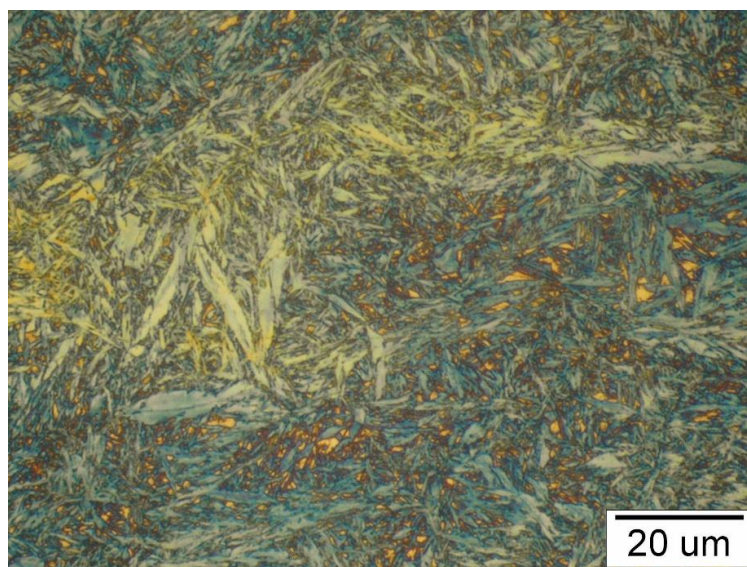


Fig. 1. Initial carbide free bainitic microstructure obtained after air cooling (etched with 7% $\text{Na}_2\text{S}_2\text{O}_5$).

The microstructure is revealed by tint etching with 7% aqueous sodium-metabysulfite ($\text{Na}_2\text{S}_2\text{O}_5$). This etchant is known to respond by coloring bainitic regions blue and martensite brown, whereas retained austenite etches white or in a slight purple [18]. In the hot rolled and air cooled condition, the steel has a high hardness of 660 HV1. And exhibits a yield strength of 1800 MPa. The retained austenite within the microstructure introduces a TRIP effect resulting in about 6% of homogeneous elongation, the steel continuously work hardens during straining to an ultimate tensile strength of 2800 MPa.

Prior to welding the steel was soft annealed by applying a rapid spheroidization treatment [15], with the use of the divorced pearlite reaction. The latter is very easily obtained in the new steel due to the high coherency between ferrite and cementite, which is known to promote this perlite morphology. After spheroidizing, we obtain a low hardness of 225HV1. This enabled the steel to be then milled and the final thickness reduced to 3mm. The microstructure in the soft annealed state is shown in Fig. 2.

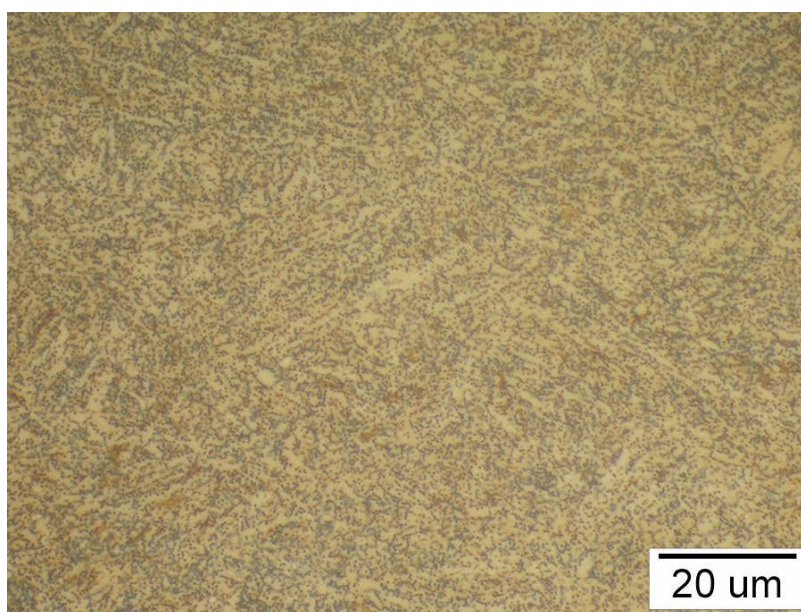


Fig. 2. Microstructure after spheroidization annealing, visible are numerous spherical cementite particles and prior bainite lath boundaries. (Etched with Viella).

Critical thermodynamic temperatures such as B_s , M_s , as well as the temperatures of onset and completion of austenite formation A_{c1} and A_{c3} respectively, are summarized in table 2. The methods used in the calculations can be found in [16], [17].

Table 2. Calculated critical thermodynamic temperatures of the newly developed KAB steel (in $^{\circ}\text{C}$).

A_{c1}	A_{c3}	B_s	M_s
703	879	240	51

Welding was performed with single pass welds of two adjacent plates on a steel backing, without the use of preheating. Prior to welding an active flux was applied to increase the penetration depth. Different welding speeds were tested and the results are summarized in table 3. The heat input was consequently varied as described by equation 1 in accordance with the standard EN 1011-1:

$$Q = \frac{U \cdot I \cdot 60}{V \cdot 1000} \cdot k \left[\frac{\text{kJ}}{\text{mm}} \right]$$

Where U , I and V are the welding voltage [A], current [V] and speed [mm/min] respectively. The efficiency of the welding process needs to be accounted for as well and in the case of TIG welding the above standard recommends the value of k as 0.6.

Table 3. Welding parameters

eld	W	Heat input [J/cm]	A mp. [A]	Vol tage [V]	Speed [cm/min]	tion [mm]	Penetra depth
1		13630	42	16	6		3
2		10220	42	16	8		3
3		8172	42	16	10		2,7

3. RESULTS

Samples for metallographic investigation were EDM cut, and hot mounted in resin. Preparation proceeded by grinding using SiC papers from P220 to P4000, followed by polishing using 1 μm diamond suspension, the final step was polishing with 0.05 μm colloidal alumina. Hardness measurements were performed using a Vickers hardness tester with a 1 Kg load and a spacing of 10d (0.6 mm). Metallographic observations of the etched samples were performed with an optical microscope using polarized light. The microstructure of the weld and reaustenitized region produced by a high heat input is shown in Fig. 3. It can be seen that due to the high heat input appreciable grain growth occurs within the reaustenitized region, where the grains have coarsened up to a size of about 200 μm . Grain growth is accompanied with the segregation of alloying elements towards the grain boundaries, and the retention of austenite in the adjacent area upon cooling. The microstructures were revealed by tint etching with 7% aqueous sodium-metabysulfite. Coloration using the latter strongly suggests a predominantly bainitic microstructure with some martensite formed throughout the weld and in the heat affected zone. As this phenomena are thought to originate from excessive heating, we attempt to mitigate them by reducing heat input, by means of a faster welding speed.



Fig. 3. Weld microstructure formed with a high heat input, visible are radially oriented dendrites and severe grain growth (etched with 7% $\text{Na}_2\text{S}_2\text{O}_5$).

An example of a weld produced using a low heat input can be seen in Fig. 4, the dotted line indicates the position of the welded plates.

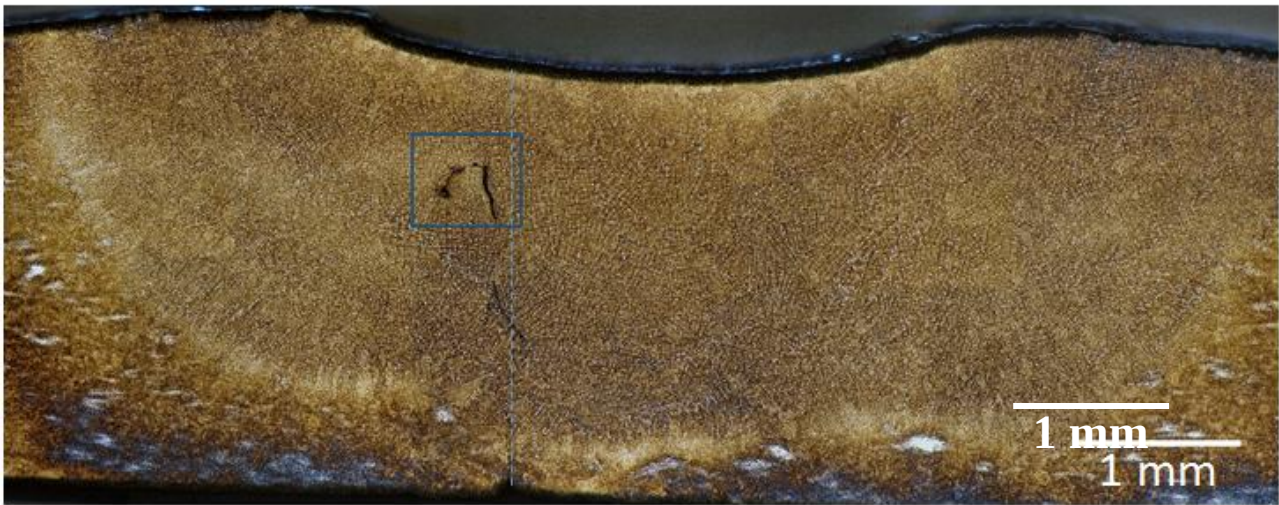


Fig. 4. Microstructure of Weld 3, showing the weld bead, HAZ and centerline porosity due to insufficient fusion (etched with Viella).

A more detailed view of the weld is provided in Fig. 5. In contrast a low heat input will result in a finer weld microstructure and reduce the amount of grain growth in Heat Affected Zone (HAZ), the latter is further reduced by the use of single pass welding, as previous studies have shown the weld intersections to be somewhat heterogeneous with respect to hardness.

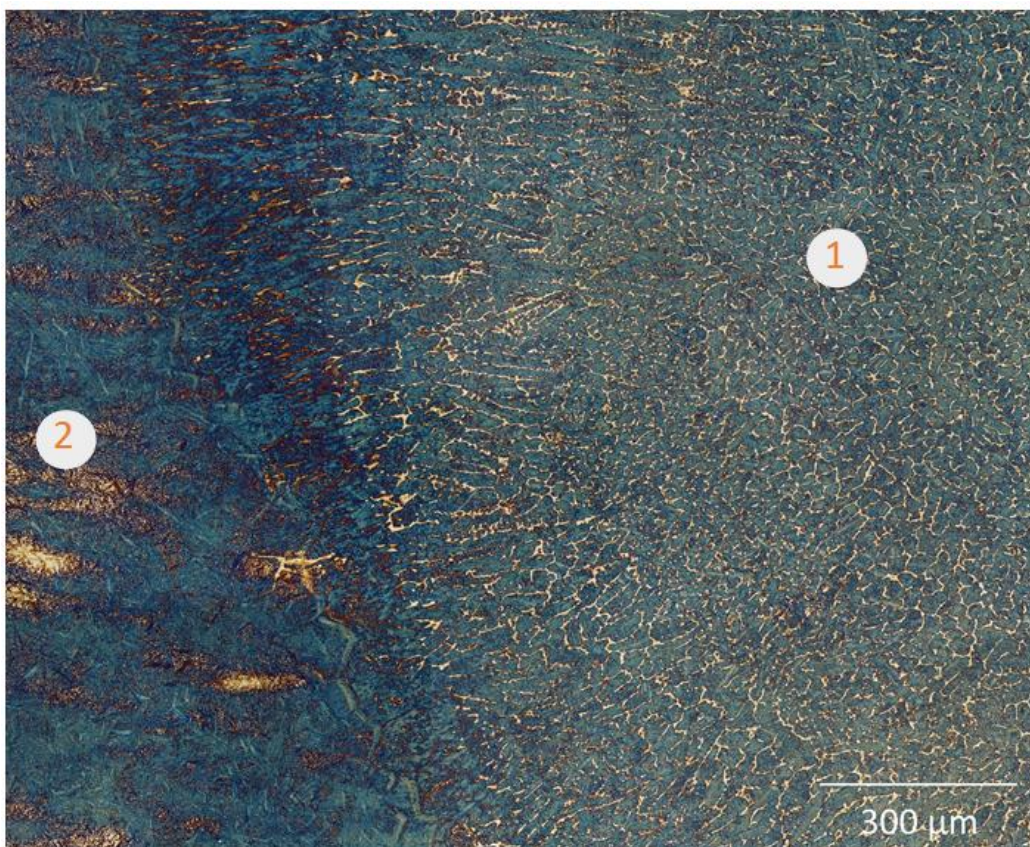


Fig. 5. Weld microstructure of weld 3, due to a lower heat input the solidification morphology changes towards equiaxed dendrites. Grain growth and coarse retained austenite are still present but to a lesser extent (etched with 7% $\text{Na}_2\text{S}_2\text{O}_5$).

The sites of interest are the martensite islands in the HAZ, and the dendritic microstructure of the weld. After single pass welding with the adjusted parameters all of the welds exhibit a uniform hardness of 640 HV1 throughout the entire weld as shown in table 4. From the micrograph in Fig. 6-B it can be seen that appreciable segregations within the heat affected zone are still visible, but to a lesser extent. The morphology of the weld microstructure changes from radially orientated dendrites, towards more equiaxed dendrites with a slight radial character adjacent to the HAZ as can be seen in Fig. 5.

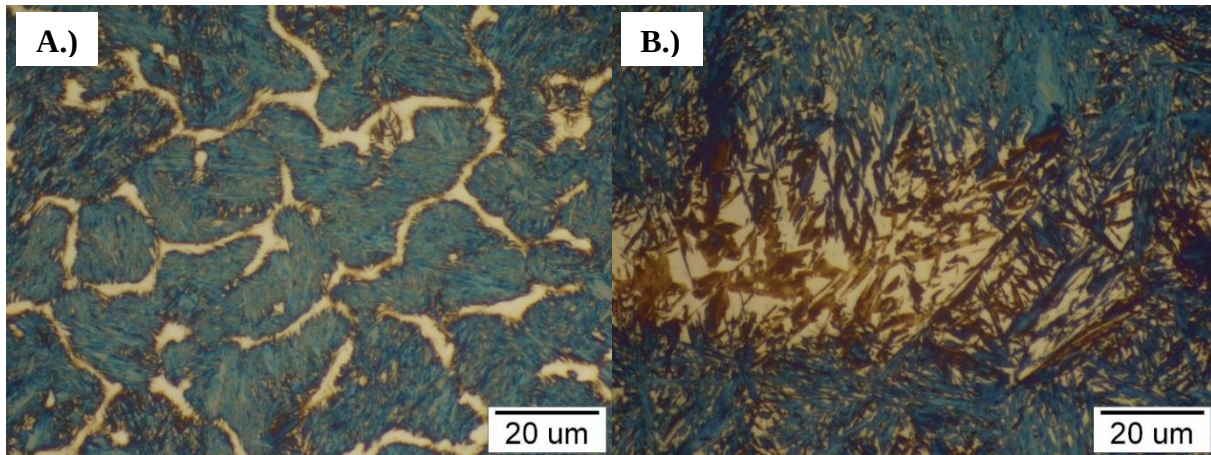


Fig. 6. A.) Detail 1, showing equiaxed dendrites formed within weld 3. B.) Detail 2, showing the localized formation of martensite and coarse retained austenite blocks (etched with 7% $\text{Na}_2\text{S}_2\text{O}_5$).

From the measurements in table 4, it can be seen that the highest hardness is attained within the weld, as the remaining material remains spheroidized, therefore we observe a sharp decline in hardness as we enter the HAZ.

Table 4. Properties of different welds.

Weld	Weld y [V%]	Width of reaustenitized zone [mm]	Hardness [HV1]
1	12	2,1	640
2	10,2	1,6	640
3	9,7	1,2	642

4. DISCUSSION

During single pass welding the hardness of the different welds is uniform throughout the weld, and almost constant between the different welds. The heat input however greatly influences the width of the reaustenitized region, and the penetration depth. The latter has proven too shallow only in the case of weld 3. Weld 3 is also notably different as it solidified as equiaxed dendrites. The solidification morphology also exerts a certain influence on the total amount of retained austenite measured in the normal cross section, as can be seen in table 4. The metallographic analysis conclusively show that the bainitic microstructure was able to regenerate during the cooling of the weld. This offers new possibilities for application of KAB steels, where the as welded properties of the weld are not as important as the final properties after reheating to 1100°C followed by air cooling. The latter does however not eliminate banding, which needs to be mitigated beforehand.

The welds produced at higher welding speeds have a preferable microstructure from the point of view of toughness, however they have shown to be prone to centerline porosity as marked by the square in Fig. 4, which occurs due to the somewhat poor liquidity of the molten metal. The latter

can be traced back to its high aluminum content, which is known to exert this effect. We observe substantial localized formation of martensite or even retention of coarse retained austenite. These have been discussed in previous works, but suffice to say they are the cause of embrittlement in such microstructures [18]. Manganese is known to promote banding and localized formation of martensite in high strength steels [19] and even softer grades like dual phase steels [20], where the latter commonly contain about 2 % of Mn. Its elimination is possible by optimization of the hot rolling treatment. Manganese is commonly present in low temperature bainitic steels, due to the high hardenability requirements to form such microstructures [21], [22], and has been shown as beneficial in improving the toughness [23]. Therefore it is not viable to completely remove it from the steels composition. The hardness of the welds is highly homogeneous and shows no appreciable variations across the three welds studied. This might be explained by the fact that all welds were produced from an initially spheroidized microstructure, as generally speaking the hardness tends to vary as a function of the heat input.

5. CONCLUSIONS

It has been shown that the concept of KAB steels can be applied during welding of the steel from the spheroidized condition. Significant segregations which result in the formation of coarse retained austenite regions within the reaustenitized zone remain present even after appreciable reductions in the heat input. The latter is however favorable as it reduces the total amount of retained austenite. Therefore it seems that the weld integrity can only be further improved by alterations of the chemical composition. In particular the Mn content is thought to be the cause of the observed banding, and will either have to be regulated or the hot rolling treatment optimized. Lower heat inputs result in improved weld solidification microstructures, but provide pure fusion properties. Due to porosity being present at lower heat inputs the use of an additive material is thought to be favorable, and should be subject to further studies. The additive material should be designed with the aim to minimize interdendritic austenite, or to produce it with very high stability to provide sufficient toughness at low temperatures.

6. ACKNOWLEDGEMENTS

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