Lechosław Tuz

The Assessment of Selected Properties of Welded Joints in High-Strength Steels

Abstract: The use of technologically advanced structural materials entails the necessity of adjusting typical welding processes to special requirements resulting from the limited weldability of certain material groups. Difficulties obtaining high-quality joints may be the consequence of deteriorated mechanical properties and structural changes in materials (beyond requirements of related standards). One of the aforementioned materials is steel characterised by a guaranteed yield point of 1300 MPa, where high strength is obtained through the addition of slight amounts of carbide-forming elements and the application of complex heat treatment processes. A heat input during welding may worsen the aforesaid properties not only in the weld but also in the adjacent material. The tests discussed in the article revealed that the crucial area was that heated below a temperature of 600°C, where the hardness of the material decreased from approximately 520 HV to 330 HV.

Keywords: unalloyed steel, welding, structure, hardness, thermal cycle

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Introduction

On one hand, the use of newly developed engineering materials is connected with unquestionable advantages resulting from the unique properties of the former, yet, on the other, engineers are often confronted with problems not encountered before. One of the major advantages is higher corrosion resistance as well as higher strength or plasticity than those characteristic of previously used materials. In terms of steels, new favourable properties are usually obtained either through the modification of the chemical composition or by changing methods of thermo-mechanical treatment. The controlled processing of steels enables the obtainment of various mechanical properties [1–6]. The aforesaid situation is observed in, e.g. high-strength toughened unalloyed steels, particularly useful as regards the necessary reduction of product weight and the energy consumption of manufacturing processes [1–4]. The replacement of classical structural steels with high yield point steel significantly reduces the weight of a structure and/or may increase its load carrying capacity without increasing the weight (wall thickness) [2,5]. In turn, the use of heat treatment processes during the production of steel entails risks including irreversible changes in the structure and mechanical properties of the steel. Because of the fact that welding heat only locally affects the material adjacent to the weld groove (weld), changes are

dr in. Lechosław Tuz (PhD (DSc) Eng.) – AGH University of Science and Technology in Kraków

limited to a narrow area, i.e. the heat affected zone (HAZ). The HAZ (as is usually revealed by macroscopic tests) tends to include the area heated above Ac1. The aforesaid area may be affected by changes in the structure and mechanical properties. It should also be noted that changes in mechanical properties may also be triggered by temperature lower than Ac1 [1–10].

As a result, the joining process based on welding methods is impeded in view of the fact that the loss of mechanical properties should be avoided but also because of the performance of check tests during welding procedure qualification.

Arc welding-related procedure qualification involves both non-destructive and destructive tests, where the latter include, among other things, macroscopic tests, hardness measurements, static bend and tensile tests as well as impact strength tests. All of the aforesaid tests involve the HAZ, the range of which is determined through macroscopic tests.

The high strength of unalloyed steels is obtained by adding carbide-forming elements (vanadium, titani8um or niobium) or other hardening elements (copper), the formation of the fine-grained structure or through appropriate heat treatment in a metallurgical furnace. Because the above-presented conditions are difficult or impossible to create in welding conditions, welded joints made of high-strength steels are usually characterised by mechanical properties similar or slightly lower than those of the base material [7, 9, 10]. The loss of mechanical properties results from the effect of the welding thermal cycle and thermal processes accompanying welding, .e.g. preheating, excessive interpass temperature, post-weld holding at specific temperature (where only temperature is controlled) or post-weld heat treatment (where both time and temperature are controlled). The research work discussed in the article aimed to assess the effect of a heat input to the HAZ (during welding) on the mechanical properties of steel having a yield point of 1300 MPa and heat input-induced (if any) structural changes. The obtainment of the appropriate temperature of the material involved the use of a simulator of welding cycles.

Test material

Test specimens were made of 4 mm thick steel having a guaranteed yield point of 1300 MPa. Simulations involved V-test specimens (with the V-notch) having dimensions of 3 mm ×10 mm ×55 mm. Table 1 presents the primary mechanical properties provided by the manufacturer in the technical specification. The analysis of chemical composition involved the use of spark spectroscopy and compared with the data provided by the producer (Table 2).

Because of the fact that in spark spectroscopy the contents of such elements as oxygen, nitrogen, carbon or sulphur are determined with unsatisfactory accuracy, the identification of carbon and sulphur was extended, which entailed the use of a LECO device.

The specimens subjected to the spark spectroscopic tests were previously ground mechanically with abrasive paper, the granularity of which was restricted within the range of 100 to 600. After grinding, the specimens were washed. The analysis performed by means of the LECO device involved swarf made using a laboratory drill.

Table 1. Selected mechanical properties of steel S1300QL in the as-received state in accordance with sub-paragraph 3.1 of EN 10204

Yield point $R_{p0.2}$, MPa	Tensile strength R_m, MPa	Elongation A_5, %	Impact energy KV, J at -40°C
1300	1400 ÷ 1700	8	27

Table 2. Chemical composition of steel S1300QL in accordance with the data provided by the manufacturers (maximum % by weight in accordance with heat analysis) [1, 2], in accordance with PN-EN ISO 10025-6 in relation to the product and individual tests

Steel grade	C	Si	Mn	P	S	Cr	Mo	Ni	V[a]	Cu	B	N	Nb[a]	Ti[a]	Zr[a]
XABO 1300	0.25	0.5	1.4	0.015	0.005	0.8	0.7	2.0	0.08	-	-	-	-	-	-
Strenx 1300	0.25	0.5	1.4	0.020	0.005	0.8	0.7	3.0	-	0.3	0.005	-	-	-	-
In accordance with the standard*	0.22	0.86	1.8	0.025	0.012	1.6	0.74	2.1	0.14	0.55	0.006	0.016	0.07	0.07	0.17
Spark spectrometer	0.23	0.23	0.83	0.009	0.004	0.46	0.39	1.23	0.017	0.009	0.0016	-	0.011	<0.002	<0.002
LECO	0.23	-	-	-	0.0012	-	-	-	-	-	-	-	-	-	-

* – indicated values refer to the products designated as QL, i.e. having guaranteed toughness at a temperature of -40°C
a – chemical elements added (at least 0.01%) to refine the grain; it is also possible to add aluminium, where the minimum content of dissolved aluminium should amount to 0.01%, which corresponds to at least 0.013% of total aluminium

Testing methodology

The tests were based on simulations of thermal cycles performed using a Gleeble 3500 simulator. The specimens were heated up to a temperature of 600°C, 800°C, 1000°C and 1100°C for 2 seconds, held at the preset temperature for 0.1 second and next, cooled at a rate of 20°C/sec. to 600°C. From 600°C, cooling was performed using time intervals of approximately 5 seconds. Figure 1 presents the temperature-time courses of simulated thermal cycles. The analysis of the courses revealed the extension of cooling time from 600°C to 150°C along with an increase in the maximum heating temperature from approximately 2 seconds (approximately 225°C/s) to 5 seconds (90°C/s).

After the thermal cycle simulation, the specimens were subjected to grinding with water-resistant abrasive paper, followed by polishing performed using special polishing cloth and the water slurry of aluminium oxide. The microstructure was revealed using a 4% alcoholic solution of nitric acid (the so-called Nital).

The structure was observed using light microscopy, i.e. a stereoscopic microscope (Leica) and a light microscope (Leica DM/LM); magnification was restricted within the range of 10x to 1000x.

Hardness measurements were performed using the Vickers hardness tests and the indenter under a load of 98.1 kG (HV10). The distribution of hardness along the measurement line included both the areas of the material not heated during simulation and the area subjected to heating.

Test results

The macrostructures of the heated areas are presented in Figure 2. Because of the incised reduction of area, the area of the smallest cross-section was characterised by the highest (“preset) temperature. The area adjacent to the (above-named) area of the highest temperature was heated by heat discharged from the area heated as a result of the thermal conductivity of steel, which, depending on temperature, triggered changes in the microstructure (the range of which was the larger the higher

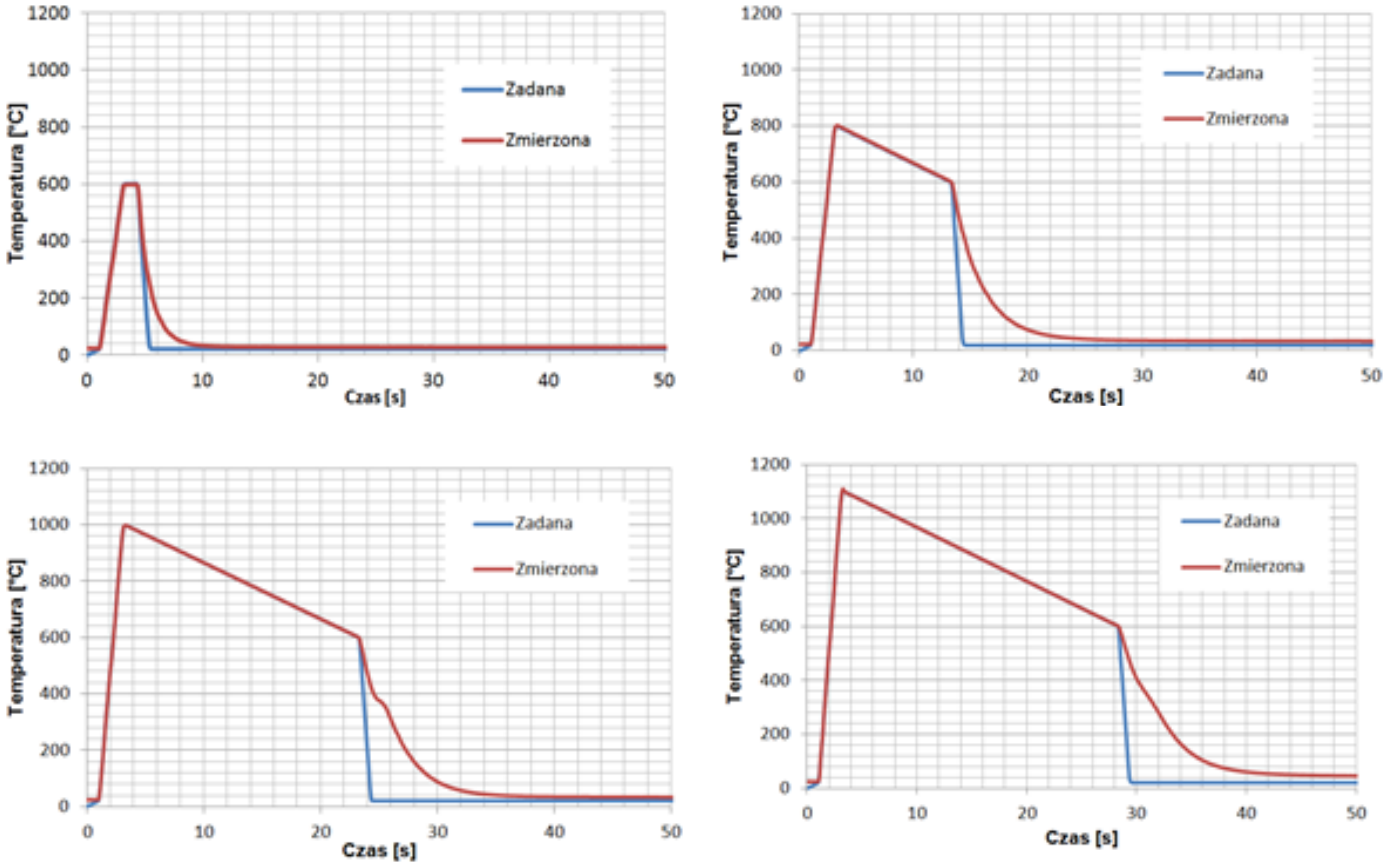

Fig. 1. Course of heating and cooling simulated using the Gleeble simulator; "preset" – preset heating and cooling temperature; "measured" – temperature measured in the specimen axis using an Ni-NiCr thermocouple in the specimen axis

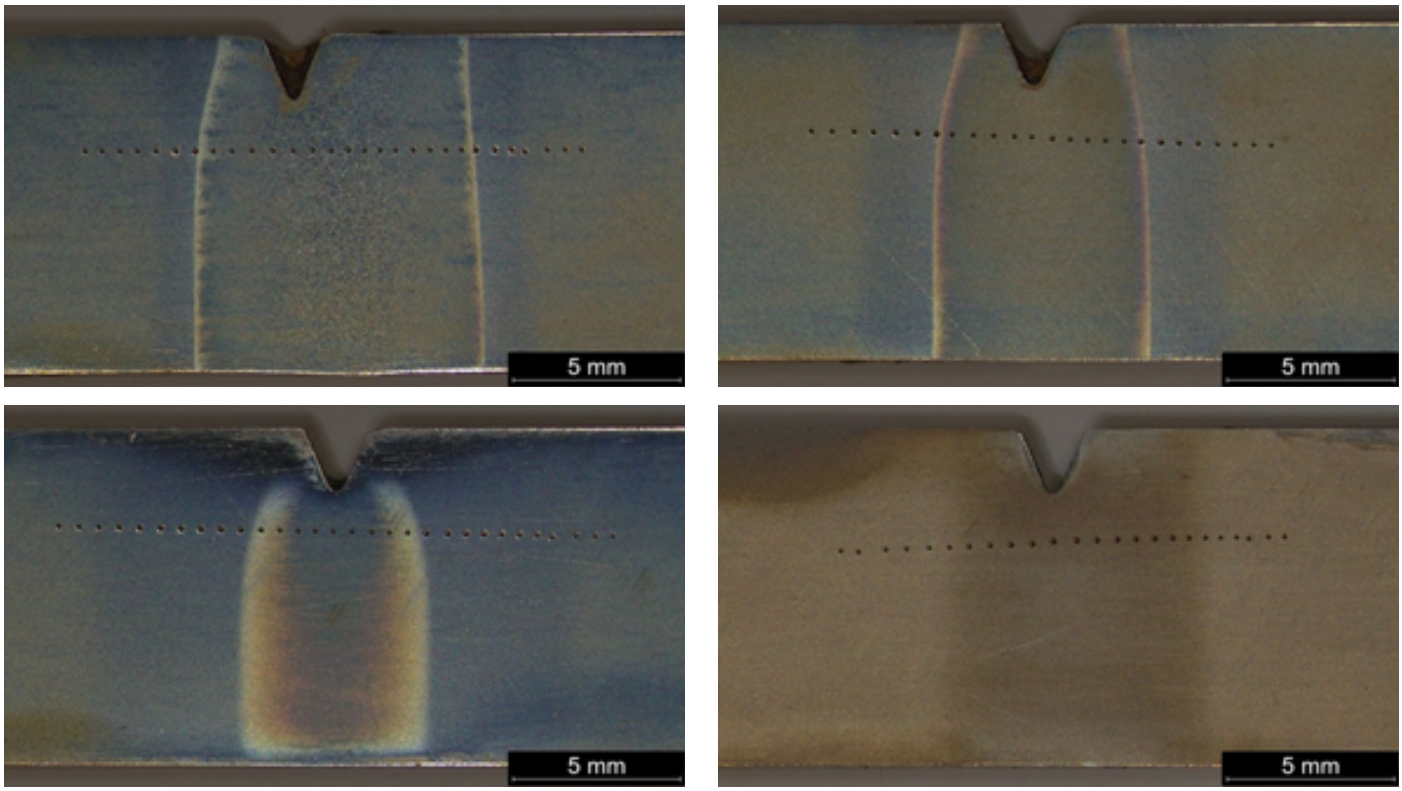

Fig. 2. Macrostructure of the area heated up to a temperature of a) 1100°C, b) 1000°C, c) 800°C and d) 600°C; in the photographs it is possible to notice the areas subjected to hardness measurements (HV10)

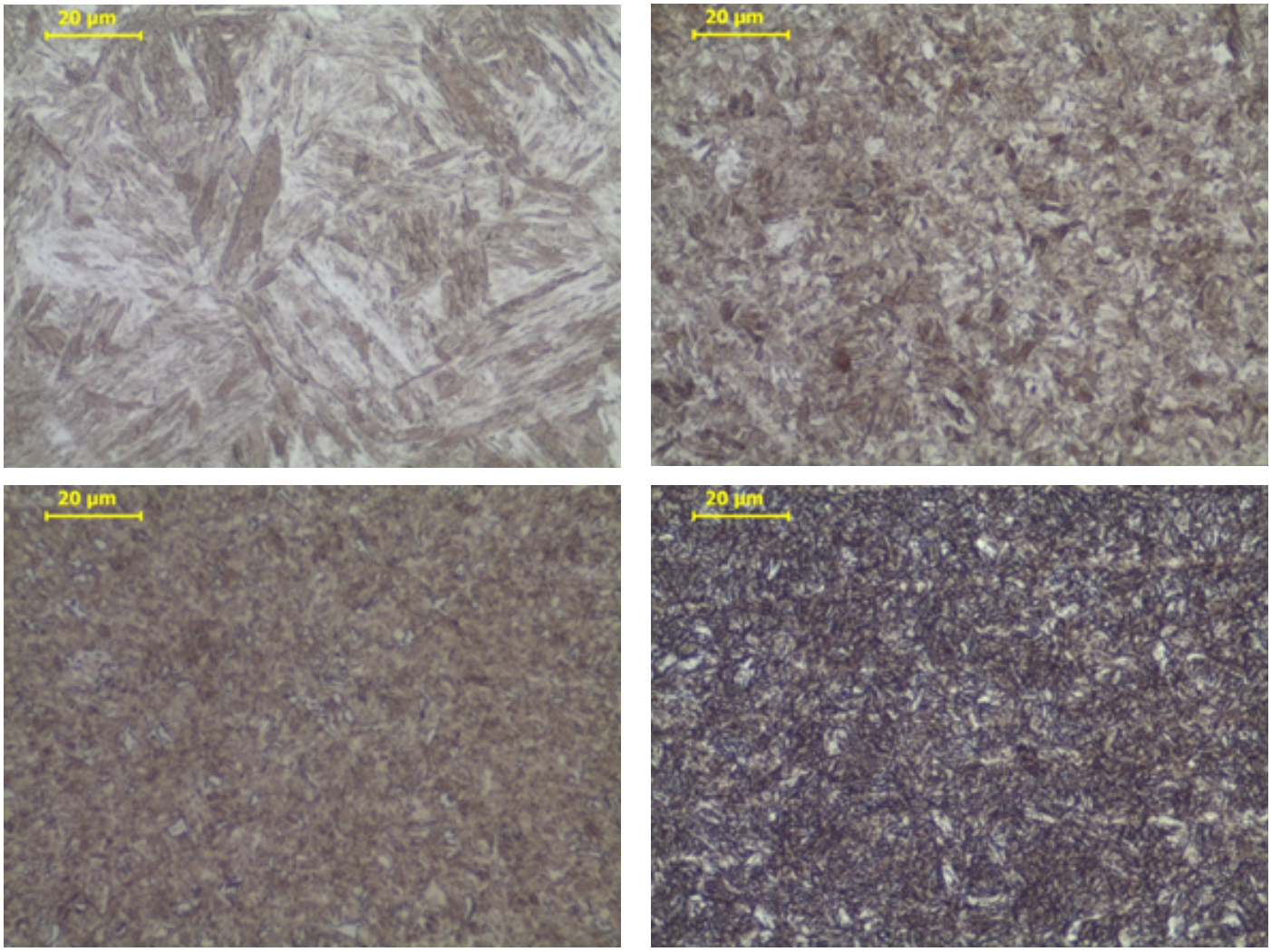

Fig. 3. Microstructure of steel S1300QL after heating up to a temperature of 1100°C (a), 1000°C (c), 800°C (c) and 600°C (d)

the heating temperature). Heating up to a temperature of above 800°C revealed the presence of the clearly visible heat affected zone. Changes resulting from heating up to a temperature of 600°C were slight. In comparison with the specimens heated up to a temperature of 800°C, 1000°C and 1100°C, the area heated up to 600°C was etched otherwise and, as a result, its identification was impeded.

The microstructure of steel S1300QL in the as-received state was the fine-grained microstructure of tempered martensite and bainite. Heating to a temperature of 1100°C resulted in the slight growth of former austenite (Fig. 3a). In cases of the remaining values of temperature (thermal cycles), the grain size did not change and the structure remained fine-grained. Heating up to a temperature above 800°C and cooling at a rate of 20°C/s resulted in the formation of tempered martensite as the dominant structure (Fig. 3a, b c). Heating up to a temperature of 600°C and cooling at a rate of 2 sec. to 150°C resulted in the formation of the martensitic-bainitic structure. Hardness measurements revealed that the area heated up to a temperature above 800°C was not characterised by a significant decrease in hardness (to approximately 480HV10 in relation to 1100°C) if compared with that of the base material. The area heated up to 800°C was even characterised by a slight increase in hardness (up to approximately 550 HV10). The area heated up to 600°C was characterised by a decrease in hardness to approximately 360 HV10. The foregoing indicates that cooling was accompanied by the precipitation of iron carbides (probably ε) resulting from the decomposition of martensite leading, in turn, to a decrease in the content of carbon in martensite. In the above-named temperature, the supersaturation with carbon decreased

and led to the greater amount of slight carbide precipitates (observed in the metallographic specimen). Carbide ε (formed at the first stage) transformed into cementite. It was also possible to observe the first symptoms of the coagulation of carbides.

The analysis of individual hardness distributions in relation to a simulated temperature of 1100°C, 1000°C, 800°C and 600°C revealed the presence of the clearly visible softened zone characterised by a decrease in hardness to approximately 330 HV10 (Fig.4). The location of the zone depended on heating temperature and the width of the heated area; the distribution of hardness itself was similar as in cases of the areas heated up to a temperature of 1100°C, 1000°C and 800°C. As regards a temperature of 600°C, the softened zone was located in the specimen axis.

Summary

The above-presented simulated heating of the high-strength unalloyed steel revealed the possible formation of the martensitic structure resulting from the heating of the steel to high temperature (above Ac1) followed by fast cooling. A decrease in temperature (cooling) was accompanied by the tempering of martensite, accompanied by a slight decrease in hardness (by approximately 30 HV).

The thermal cycles where Ac1 was not exceeded were characterised by the precipitation of carbides ε and the coagulation of cementite. As a result, despite the short thermal cycle, hardness dropped to approximately 330 HV.

A decrease in steel hardness during cooling necessitates the performance of non-destructive tests during welding procedure qualification and take into consideration the material adjacent to the HAZ (revealed during macroscopic tests). The reduction of hardness observed at temperature below Ac1 will be characteristic of welded joints made in high-strength toughened steels, where the heating of the base material by the heat discharged from the weld could lead to the formation of the softened zone.

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