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Potential Fatigue Strength Change of Welds Due to Repair Welding

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Coarse grain
Crack initiation
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Fatigue limit
Fatigue strength
Hardness
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Re-fined grain
Repair welding
Residual stresses - RS
Small defects
Weld
Weld metal- WM
Weld toe

Ključne riječi

Čvrstoća zamora
Grubo zrno
Inicijacija pukotine
Male greške
Metal zavara - WM
Osnovni materijal - BM
Prijelaz od osnovnog materijala na var
Reparaturno zavarivanje
Trajna čvrstoća
Tvrdoća
Umjetna greška
Usitnjeno zrno
Zamorna pukotina
Zaostala naprezanja - RS
Zavar
Zona utjecaja topline - HAZ

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The fatigue strength of welds in polycrystalline metals depends on the hardness, the size of defects and grains in the areas of the highest stress and on the level of welding residual stresses. Changes of residual stresses and hardness were experimentally determined on a weld which was repaired. Besides, steel samples with the microstructure of martensite were used with the intention to assess effects of residual stresses and grain size to the fatigue strength of welds. The samples were prepared using either thermal cycle simulator or laboratory furnace and water quenching. The used methods of microstructure preparation result in different surface residual stresses. Specimens for the fatigue limit measurement were prepared either defect-free or with artificial defects. Artificial surface defects were small in comparison with the size of grains. The defects were made by drilling small holes and indenting with the Vickers pyramid. The results show that tensile residual stresses and coarser grains lower the fatigue limit of steel and vice versa. Enhancement of welding residual stresses caused by repair welding leads to fatigue strength reduction of welds while hardening of weld materials leads to its increase.

Potencijalna promjena čvrstoće zamora zavara uslijed reparaturnog zavarivanja

Izvornoznanstveni članak

Čvrstoća zavara na polikristalnim metalima prilikom umaranja ovisi o tvrdoći, veličini grešaka i veličini kristalnog zrna u području najvećih naprezanja, te o razini zaostalih naprezanja uslijed zavarivanja. Promjena zaostalih naprezanja i tvrdoće eksperimentalno je određena na zavaru koji je bio repariran. Pored toga, korišteni su uzorci s martenzitnom mikrostrukturom s namjerom da se procjene efekti utjecaja zaostalih naprezanja i veličine kristalnih zrna na čvrstoću na umaranje kod zavara. Uzorci za takvo ispitivanje bili su pripremljeni ili korištenjem simulatora toplinskog ciklusa zavarivanja ili korištenjem laboratorijske peći uz hlađenje vodom. Korištene metode pripreme mikrostrukture uzoraka rezultirale su različitim površinskim naprezanjima. Uzorci za mjerenje trajne čvrstoće bili su pripremljeni bez grešaka i s umjetnim greškama. Umjetne greške bile su male u usporedbi s veličinom kristalnih zrna. Greške su napravljene bušenjem malih provrta i utiskivanjem Vickersove piramide. Rezultati pokazuju da vlačna zaostala naprezanja i grublja zrna smanjuju trajnu čvrstoću čelika i obrnuto. Povećanje zaostalih naprezanja uslijed zavarivanja uzrokovano reparaturnim zavarivanjem vodi prema smanjenju čvrstoće na umaranje zavara, dok povećanje tvrdoće zavara vodi prema njezinom povećanju.

1. Introduction

Much smaller cracks than the most important microstructural units of metals have no influence on fatigue limit. The effect of other defects of the same size is similar. Actually, cracks can start from microstructurally small defects. They spread unexpectedly fast at the beginning. The crack extension decelerates gradually when its tip is approaching micro structural obstacles such as grain boundaries and therefore can stop. Such cracks are non-extending. Kitagawa-Takahashy diagram describes the fatigue limit of metals with non-propagating cracks from the smallest ones (short cracks), over medium (physical small cracks) to the biggest (long cracks) [1]. However, the fatigue limit of metals with defects in size close to or bigger than the microstructural units is reduced.

In the vicinity of welds, large thermal stress gradients are results of local material heating and cooling. Thermal contractions can cause weld cracking, distortion or shorter service life. The main reason is welding residual stress (RS) appearance. As a consequence, welded structures can become susceptible to hydrogen embrittlement and other detrimental phenomena. The type and level of welding RS are crucial for the fatigue strength of welds in the as-welded condition. RS are treated as a local static pre-stress of welds.

Formation of the heat-affected zone (HAZ) is an important consequence of applied thermal cycle when welding. The base metal (BM) adjacent to the fusion line is heated for some time almost to the melting point. The result is an extensive growth of grains. For that very reason the microstructure of the coarse-grain heat-affected zone (CGHAZ) emerges at the fusion line. The effect of thermal cycles diminishes with the distance from the fusion line.

Butt-welds are the strongest welds. The fatigue strength of butt-welds is affected by their shape. Stress concentration in butt-welds is the highest at the weld toes. As shown in Figure 1 toes coincide with the location of CGHAZ. Mechanical properties of metals are microstructure dependent. Hardness of CGHAZ is the result of chemical composition, initial state of BM and applied welding parameters.

Defects are always present in metals due to the way of processing. When an insufficient weld quality is proven by use of suitable non-destructive examination (NDE) method the weld should be repaired. Weld defects in the weld metal (WM) and HAZ appear mainly due to the nature of the welding process. Defects that are smaller than the threshold sensitivity of the NDE method cannot be detected.

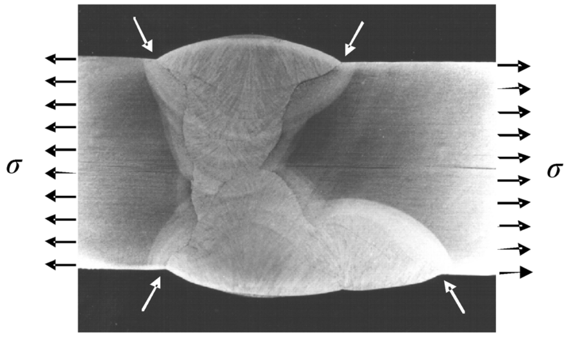


Figure 1. Macrograph of a multi-pass K-butt weld with remote homogeneous stress (arrows indicate weld toes, the position of concentrated stress fields)

Slika 1. Makro snimak sučeonog K-spoja zavarenog u više prolaza u homogenom nominalnom polju naprezanja (strjelice pokazuju prijelaze od osnovnog materijala na zavar gdje je polje naprezanja koncentrirano)

Due to much heavier welding conditions by mending repair welding should be executed extremely carefully. The most important effect of repair welding to the weld is an enhancement of restraint that results in higher level of welding RS. High preheating is often needed by repair welding because of lower heat input usage. All these influence the hardness of the weld, the level welding RS and the weld shape in the area where cracks appear if stress amplitude or stress concentration is sufficiently high during exploitation.

Stress-ratio has a significant role in the fatigue limit of metals: the higher the stress-ratio the easier the fatigue crack growth. Eventual RS change affects stress-ratio at the weld toes.

Change of hardness and RS level caused by repair welding was experimentally assessed in this article. The results are discussed in light of fatigue behaviour of specimens with stress concentration such as at the weld toe. Specimen surface was prepared, either smooth or with surface defect. Two types of small artificial surface defects were used in the experimental work; drilled holes and Vickers indentations. Specimens were made from steel with different average grain size. The result of microstructure preparation on a thermal-cycle simulator and using water quenching is different at RS level.

2. Fatigue limit of metals with small defects

An increased possibility of crack initiation at the weld toe is of the greatest importance for the fatigue strength of welds and the load-carrying capacity of cyclic loaded welded structures. Small weld defects as different inclusions, scratches or cracks are often found at the weld toes. Even a sharp transition from the weld reinforcement to the BM should be treated as a small defect. Anyway, fatigue cracks in butt-welds start in the coarse-grain steel, i.e. in the CGHAZ. Small surface defects are either present or not present there [2].

Defects decrease fatigue limit of metals because of easier crack initiation. In the past, S-N curves were the only available way to predict fatigue lives of workshop-quality welds. Now, LEFM concepts are often applied to the cracks in metals.

Murakami and co-workers treated the influence of various shaped small defects in the same way as cracks, i.e. using LEFM [3–5]. The parameter reflecting the effect of small defects on the fatigue limit of metallic materials is the square root of the defect projection onto the plane perpendicular to the cyclic stress ($\sqrt{\text{area}}$). However, LEFM underestimates growth rates of short cracks within the local plastic zones that can develop as a result of stress concentrations in welds [6, 7].

According to Murakami [3, 4], the fatigue limit of polycrystalline metals with small defects, σ_w, is predicted using only two parameters; defect size and material hardness.

$$\sigma_{\mathrm{w}} = \frac{1{,}43\,(HV + 120)}{\left(\sqrt{\text{area}}\right)^{\frac{1}{6}}}, \tag{1}$$

σ_w is expressed in MPa, HV in Vickers hardness number and $\sqrt{\text{area}}$ in μm.

Natural small weld defects in metals can be artificially modelled. As small surface defects drilled small holes were used in the past with success when endurance limit of real quality metals was studied. Vickers indentation is also a promising small artificial surface defect. Indenting with the Vickers pyramid is easy to plan and execute. The load on the pyramid has to be adjusted according to the material hardness and the expected indentation size [8, 9].

The problem with use of artificial weld defects is the local RS appearance. Those stresses are the result of irreversibility of material plastic deformation with defect manufacturing.

In order to evaluate only the effects of small defect to the fatigue limit, RS should be omitted. Electro-etching is usually used to remove surface stratum with the highest RS without any plastic deformation. In such a way, local and global surface RS are lowered. Unfortunately, local RS lowering is closely linked with the defect's shape and size change also, i.e. with the actual change of parameter size. When global RS are present also, it is not possible to annul only local RS.

Some effects of local RS caused by defect manufacturing were already discussed elsewhere [10, 11]. It was also shown how to carry out a two step heat treatment in order to obtain small artificial defects on the specimens with microstructure of martensite without the presence of local RS.

3. Experimental work and results

In the first part of the experimental work where the effect of global RS on the fatigue limit of steel with and without small defects was studied, nickel-molybdenum steel was used. Chemical composition of the steel is shown in Table 1.

Table 1. Chemical composition of the nickel-molybdenum steel

Tablica 1. Kemijski sastav čelika legiranog nikljem i molibdenom

C	Si	Mn	P	S	Cr	Ni	Mo	Cu	Al
mass %									
0,18	0,22	0,43	0,012	0,028	1,56	1,48	0,28	0,15	0,023

HAZ microstructure was prepared on pieces of steel using thermal cycle simulator and two different heat treatments in furnace. The thermal regime during welding single-cycle simulation on a thermal cycle simulator is shown in Figure 2. Peak temperature T_p is on the level 1300 °C while cooling time $\Delta t_{8/5}$ round 5 s. The result is CGHAZ microstructure formation of martensite. The same microstructure was our aim in the modelling single-step heat treatment in furnace. The procedure is shown in Figure 3a. Two-step heat treatment in furnace used for double-cycle HAZ preparation is shown in Figure 3b. The result is refined HAZ microstructure formation of martensite.

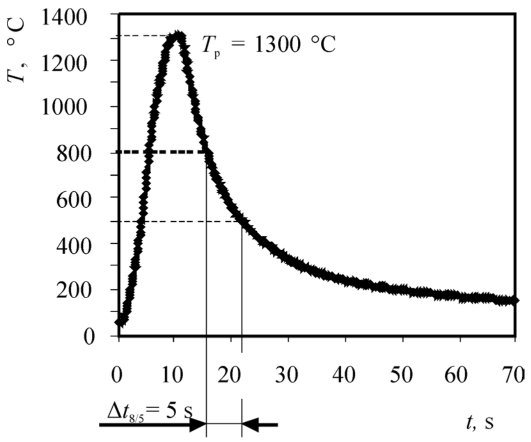


Figure 2. Temperature lapse on pieces of steel characterised by peak temperature (T_p) and cooling time ($\Delta t_{8/5}$) that result is in CGHAZ formation with the microstructure of martensite

Slika 2. Promjena temperature na uzorcima čelika koju karakterizira maksimalna temperatura (T_p) i vrijeme hlađenja ($\Delta t_{8/5}$) a čiji rezultat je oblikovanje grubo zrnate zone utjecaja topline s martenzitnom mikrostrukturom

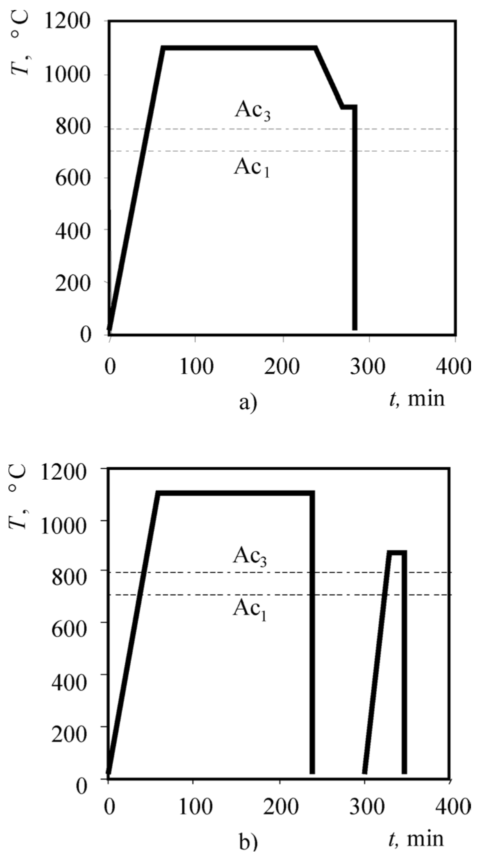


Figure 3. Heat treatments in furnace: simulation of CGHAZ microstructure of martensite (a), simulation of re-fined HAZ microstructure of martensite (b)

Slika 3. Toplinski tretman u peći: simulacija martenzitne mikrostrukture grubo zrnate zone utjecaja topline (a), simulacija martenzitne mikrostrukture prvotno grubo zrnate zone utjecaja topline sa usitnjenim zrnom (b)

At the beginning of both heat treatments, a 3-hour-coarse-grain annealing at 1100 °C was performed on pieces of steel in the furnace:

- In order to prepare CGHAZ microstructure temperature in the furnace was lowered to 870 °C. Pieces of steel cooled to that temperature were then water quenched. The result is a coarse-grain microstructure of martensite very like that formed at the weld toe in the case of single pass "cold welding".
- In order to prepare refined HAZ microstructure pieces of steel were cooled in water after annealing. In the next step of heat treatment re-austenitised steel was water quenched fom 870 °C. The result is refined microstructure of martensite very like that formed at the weld toe in the case of double-pass "cold welding" when tempering weld-passes are applied to the top layer.

Compressive surface RS is the result of cylindrical steel pieces water quenched. They were assessed by the hole-drilling method to be round 300 MPa [12]. In contrast, heat conveying from steel pieces by means of water cooled grips during welding simulation on the weld thermal-cycle simulator results in tensile surface RS. It was not possible to quantify the level of RS at the bottom of the circumferential notched specimens using hole-drilling method, but for sure, RS are tensile.

Microstructures of martensite prepared by temperature lapses shown in Figures 2 and 3 are presented in Figure 4. They are named steels with microstructure MS-a, MS-b and MS-c.

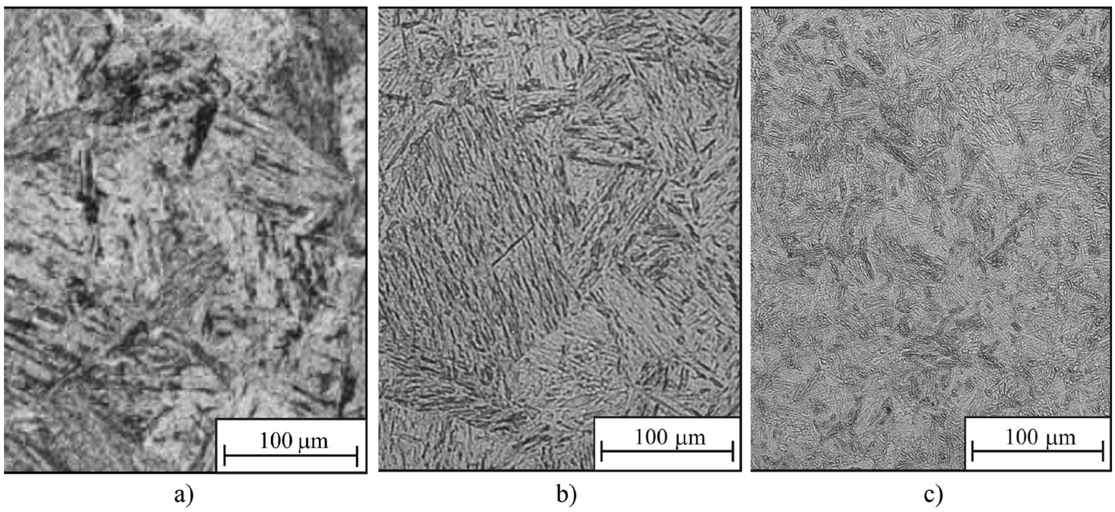


Figure 4. Coarse-grain microstructure of martensite, MS-a, prepared by weld thermal cycle (a); coarse-grain microstructure of martensite, MS-b, prepared by single-step thermal treatment in furnace (b); refined microstructure of martensite, MS-c, prepared by two-step thermal treatment in furnace (c)

Slika 4. Grubo zrnata martenzitna mikrostruktura MS-a, pripremljena toplinskim ciklusom zavarivanja (a); grubo zrnata martenzitna mikrostruktura MS-b, pripremljena toplinskim tretmanom u peći u jednom stupnju (b); usitnjena martenzitna mikrostruktura MS-c, pripremljena toplinskim tretmanom u peći u dva stupnja (c)

Bend-specimens were machined from samples of those steels. They were circularly notched at mid-length. The shape and size of those specimens are shown in Figure 5. Stress concentration factor in bending is 1,74 [13].

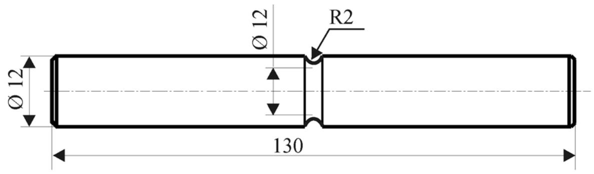


Figure 5. Geometry of rotary bend-specimen

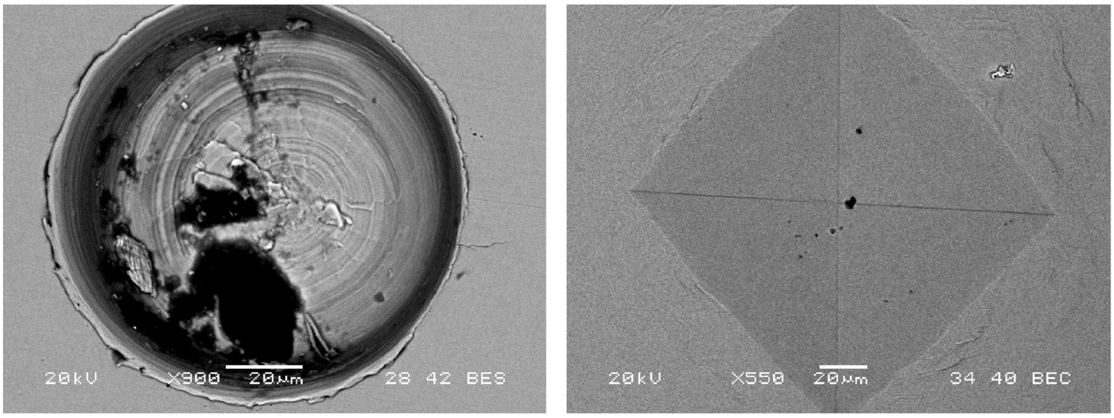
Slika 5. Geometrija probe za rotacijsko savijanje

A small artificial surface defect was prepared at the bottom of the notch, i.e. either drilled hole or Vickers indentation. They were made after simulations. The appearance of defects is shown in Figure 6.

Final result of the present investigation is bending fatigue strength of specimens manufactured from steels MS-a, MS-b and MS-c. The fatigue strength is determined on specimens with and without small defect on the surface. The average grain size of studied steels is not the same. The fatigue strength of specimens is determined in two different surface RS conditions. The results are shown in Table 2.

Already at the stress level a little bit lower than the fatigue limit of steel small cracks appear. They start either at the surface of smooth specimen (Figure 7) or from artificial defects (Figure 8). They begin to spread, but cracks expansion in the course of time stops. Those cracks that have been first initiated and then spread to certain size belong to non-expansing cracks.

The effect of local RS caused by manufacturing of defects, i.e. drilling holes and indenting with the Vickers pyramid with equal parameter $\sqrt{\text{area}} \cong 54$ μm is up to 13 % [10-12].



a) b)

Figure 6. Artificial surface defects: drilled hole with diameter 90 μm and depth 45 μm (a) and indentation made by Vickers pyramid with the diagonal 200 μm (b)

Slika 6. Umjetne površinske grješke: mali provrt promjera 90 μm i dubine 45 μm (a) i utisak Vickersove piramide dijagonale 200 μm (b)

Smooth and surface defected specimens were bend-loaded on a rotary bending machine at room temperature in the moment-control mode. Frequency of the loading and the stress rate were $f \cong 100$ Hz and $R = -1$, respectively. The defect size parameter $\sqrt{\text{area}}$ was the same for both types of used artificial defects, i.e. round 54 μm.

Hardness of steels MS-a, MS-b and MS-c is almost the same, i.e. 440 HV [12]. They have microstructure of martensite. Average grain size of the steels MS-a and MS-b is 200 μm while of the steel MS-c 20-30 μm [12]. Surface RS in the samples of steels MS-a are tensile. Exact level of those stresses is not known. In contrast, in the samples of steel MS-b and MS-c surface RS are experimentally determined. They are on the level 300 MPa [12].

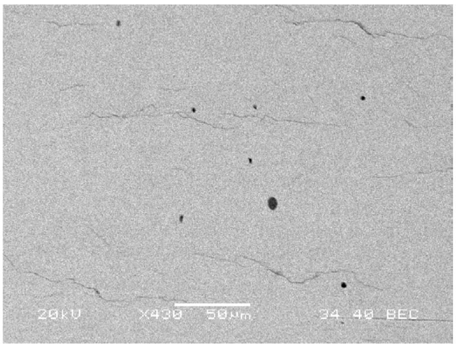


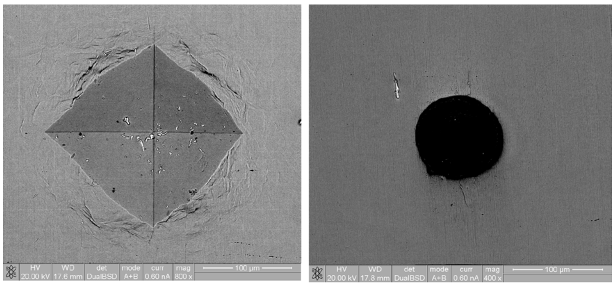
Figure 7. Stable cracks initiated on the specimen without surface defects

Slika 7. Stabilne pukotine inicirane na probi bez površinskih grješaka

Table 2. The bending fatigue strength of steels MS-a, MS-b and MS-c in different conditions

Tablica 2. Trajna čvrstoća čelika MS-a, MS-b i MS-c u različitim uvjetima

Steel / Čelik	MS-a	MS-b	MS-c	MS-a	MS-b	MS-c
Grain size / Veličina zrna, µm	200	200	20-30	200	200	20-30
RS, MPa	tensile / zatezanje	≅ -300	≅ -300	tensile / zatezanje	≅ -300	≅ -300
Surface / Površina	smooth / glatko	smooth / glatko	smooth / glatko	defected / s grješkom	defected / s grješkom	defected / s grješkom
Fatigue strength, MPa / Trajna čvrstoća, MPa	≅ 330	≅ 537	≅ 760	≅ 330	-	664-678



a) b)

Figure 8. Stable cracks initiated on the defected specimen. Small artificial surface defects are Vickers indentation (a) and drilled hole (b)

Slika 8. Stabilne pukotine inicirane na probama s grješkom. Male umjetne površinske grješke su Vickersovi otisci (a) i bušeni provrti (b)

In the second part of the experimental work, a 200 mm long, 200 mm wide and 12 mm thick but-welded coupon in the structural steel plate grade HT 50 was prepared. Chemical composition and basic mechanical properties of the steel are shown in Table 3. Steel was in the quenched and tempered condition.

A 3-weld-pass butt-weld was made by a submerged arc welding (SAW) using flux-cored wire Filtub 128 and flux FB TT. Net heat-input was 17-18 KJ/cm. Preheating was not applied. Inter-pass temperature was kept 70 °C. The chemical composition and the basic mechanical properties of the consumable are shown in Table 3.

A photograph of the weld cross-section and its hardness across the last weld-pass in the as-welded condition is shown in Figure 9.

RS in the primary weld were measured in the as-welded condition. The hole-drilling method was used [14]. Weld reinforcement was carefully removed and strain-gauge rosettes attached. Surface welding RS determined in the transverse direction along the weld face will be shown together with the RS results of the repaired weld in Figure 12a.

The weld was grooved to a depth round 7 mm in length 100 mm and repaired by shielded metal arc welding (SMAW) using stick electrode EVB NiMo with basic coating without preheating. Net heat-input by welding was 9 KJ/cm. Inter-pass temperature was kept 70 °C. The chemical composition and the basic mechanical properties of consumable are shown in Table 3.

Cross-section of the repaired weld is sketched in Figure 10a. The repair is made by seven weld-passes. Hardness and RS were measured in the as-repaired condition. Hardness on the weld face was measured along 14 lines as shown in Figure 10b. The first and the last line cross the primary weld which is affected by heat of repair welding while the rest cross the weld.

Hardness of repaired weld measured along six representative lines perpendicular to the weld axis is shown in Figure 11. We can compare those results with the results of the primary weld shown in Figure 9b.

Table 3. Chemical composition and guaranteed properties of the structural steel and both welding consumables

Tablica 3. Hemijski sastav i garantirana svojstva konstrukcijskog čelika i oba dodatna materijala

	C	Si	Mn	P	S	Cr	Ni	Mo	$R_{p0.2}$	R_m	CVN l/t
Material / Materijal	mass %								MPa	MPa	J, -40 °C
BM	0,06	0,34	0,41	0,010	0,004	0,73	0,27	0,036	>490	520-670	118/78
Wire / Žica	0,05	0,20	1,40	-	-	-	1,20	0,40	>550	630-730	100
Electrode / Elektroda	0,06	0,40	0,90	-	-	-	1,10	0,35	>510	580-710	47

Weld reinforcement of repaired weld was removed. Strain-gauge rosettes were attached. The results of surface welding RS measurement in the middle of the weld in the transverse direction are shown in Figure 12a (WM-repaired weld). The hole-drilling method was used [14]. Residues of drilled holes are clearly seen in Figure 10b (three in the mid-length of repair weld and two at both ends).

The results of RS measurement are shown together with RS in the middle of the weld as well as in the HAZ which were determined in the primary weld before repair welding (WM-primary weld, HAZ-primary weld). The level of RS in the repaired weld is much higher than that measured in the primary weld. Stresses in the middle of the primary weld are tensile (between +10 and +60 MPa). Stresses in the HAZ are compressive (between –120 and 140 MPa). Stresses in the middle of the weld after repair are tensile (between +20 and +180 MPa).

If we compare results of hardness measurement in Figure 11 we will notice that hardness at the end of the repair weld (Lines 2, 3 and 4) has increased in comparison with those in the mid-length of the weld (Line 8). Hardness at the beginning of the repair weld is also higher than in the mid-length of the weld (Lines 13 and 14).

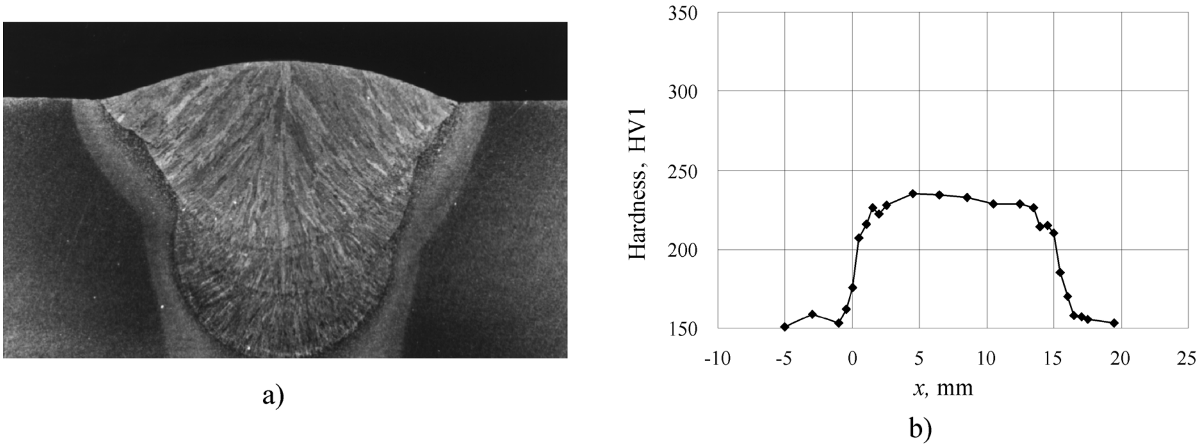


Figure 9. Cross-section of the primary weld (a) and its hardness (b)

Slike 9. Presjek primarnog zavara (a) i njegova tvrdoća (b)

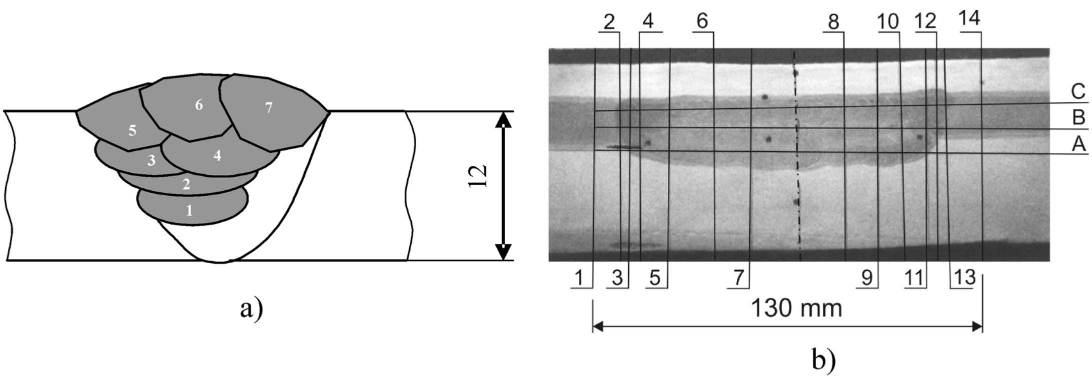


Figure 10. Repaired weld: cross-section of the weld (a), photograph of the weld with lines of hardness measurement (1-14) and three lines for hardness evaluation (b)

Slika 10. Repariran zavar: presjek zavara (a), fotografija zavara s linijama mjerenja tvrdoće (1-14) i tri linije vrednovanja tvrdoće (b)

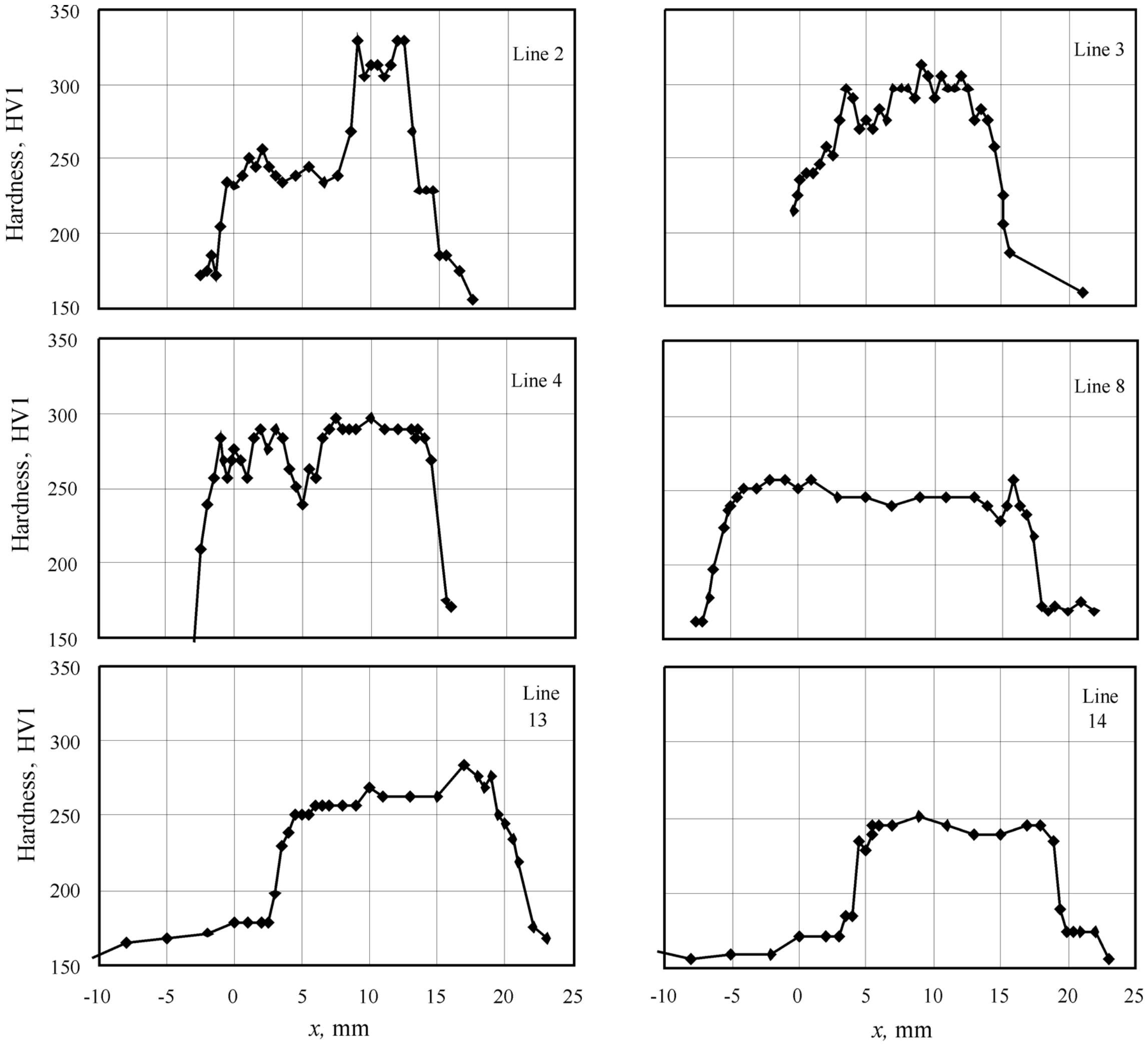


Figure 11. Repaired weld hardness across the weld

Slika 11. Tvrdoća preko repariranog zavara

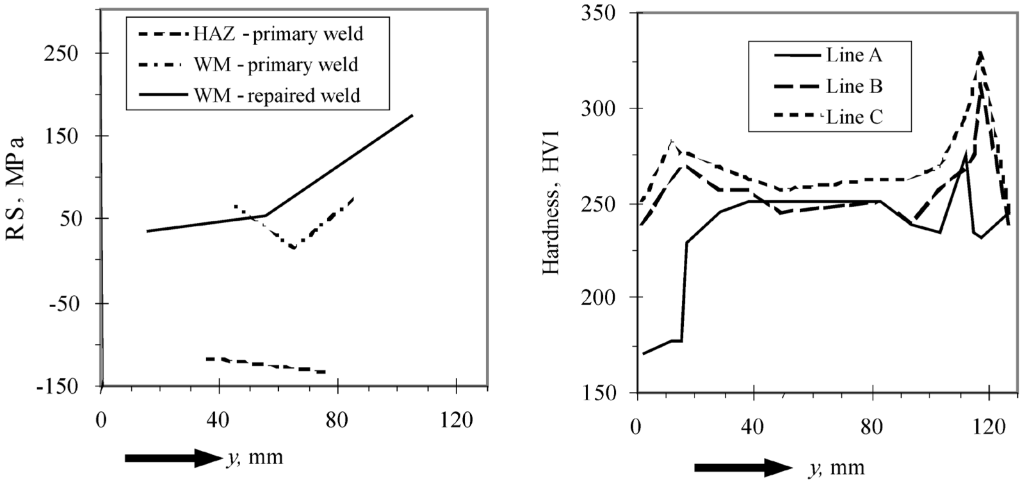


Figure 12. Transverse welding RS along the weld in the primary weld and repaired weld (a), hardness along lines A, B and C (b); both with the welding direction

Slika 12. Zaostala naprezanja uslijed zavarivanja koja su poprečno na zavar u smjeru osi zavara u primarnom zavaru i repariranom zavaru (a), tvrdoća uzduž linija A, B i C (b). I jedno i drugo je prikazano u smjeru zavarivanja

4. Discussion with conclusions

It was proven in the past by experiments that the fatigue limit of polycrystalline metallic materials as well as the threshold stress intensity factor that defines condition for the development of macroscopic crack from small defects depends upon only two variables; defect size and material hardness [5]. Miller [6] pointed out the responsibility of microstructural obstacles for existence of non-expansing cracks at the stress level equal to fatigue limit. The most important obstacles link with the biggest microstructural units of metals, i.e. grains, colonies of pearlite in steels with microstructure of ferrite and pearlite, domains with harder microstructural constituent in steels with duplex microstructure, etc. If we take into account only grains, the distances between grain boundaries should be considered. They are more sparsely distributed in space in the coarse grain microstructure than in the fine grain microstructure.

The fatigue limit of polycrystalline metals like steel at the weld toe on welds in the carbon steels as CGHAZ and re-fined HAZ is not only two parameters dependent. Defect size and material hardness in the present article were namely constant, but nevertheless we registered differences in the experimentally determined fatigue limit of the studied steels. It seems that grain size and global RS level in real welds are by far not less important than size of defects and hardness of welds. This is of great importance for load carrying capacity of welds under fatigue loading.

So, the experimental results listed in Table 2 are surface quality, grain size and global RS level dependent. Let's look deeper at those results in relation to the single variable when other two variables are constant. The results are three very simple diagrams. They express the fatigue limit dependence upon the surface quality (SQ), the grain size (GS) and the RS level (RSL). The diagrams are shown in Figure 13. Such an interpretation of the results actually explains some very important facts in relation to the fatigue strength of the repaired weld in which we assessed change of welding RS and weld hardness.

Figure 13a:

As expected, the fatigue limit of steel measured on the surface of defected specimens does not exceed the fatigue limit that is determined on smooth specimens. The size of the defect parameter $\sqrt{\text{area}}$ in the steel MS-c (54 μm) is comparable with the average grain size of steel MS-c (20-30 μm) while the same size of the defect parameter $\sqrt{\text{area}}$ in the steel MS-a is much smaller than the average grain size of steel MS-a (200 μm).

If a direct consequence of repair welding were new inclusions, bigger scratches or cracks or sharper transitions from the repair weld to the base (BM or WM) the fatigue strength of repaired weld would be lower in relation to the primary weld.

Figure 13b:

In the case of smooth specimens manufactured from steels MS-b and MS-c RS level is the same (-300 MPa) but the fatigue limit of the steel MS-c is higher than the fatigue limit of the steel MS-b. The reason is re-fined grain in the steel MS-c (20-30 μm) in comparison with coarse grain in the steel MS-b (200 μm).

Weld repairing can enlarge grains. Deposition of a new weld-pass is linked with the additional heat input to the material. The consequence is some grain growth.

Figure 13c:

In the case of smooth specimens manufactured from steels MS-a and MS-b grain size is the same (200 μm) but the fatigue limit of the steel MS-a is much lower than the fatigue limit of the steel MS-b. The reason is tensile RS in the former steel in comparison with compressive ones in the last steel (-300 MPa).

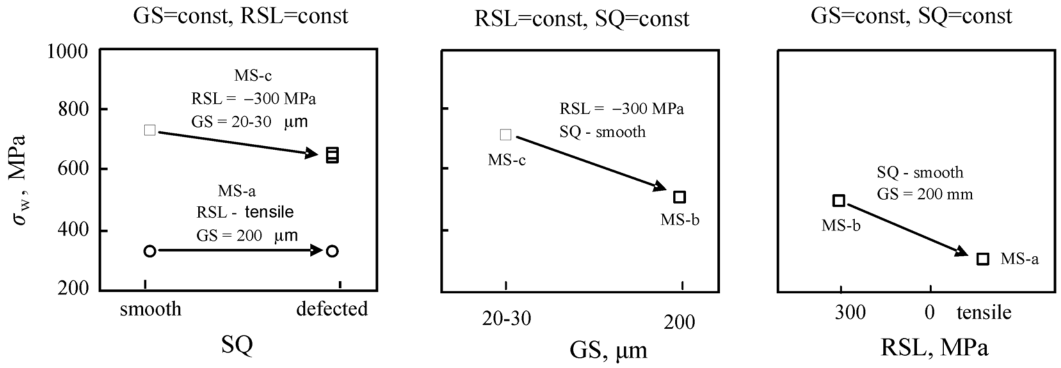


Figure 13. Influential variables to the fatigue limit of studied steels: surface quality-SQ (a), grain size-GS (b), residual stress level-RSL (c)

Slika 13. Utjecaji na trajnu čvrstoću čelika u studiji: kvaliteta površine SQ (a), veličina zrna GS (b), razina zaostalih naprezanja RSL (c)

Welding RS change due to repair welding. As shown in Figure 12a RS in the primary weld measured in the WM were lower than RS in the repaired weld measured also in the WM. The main reason for an enhancement of RS is higher restraint. The highest RS in the repair weld were registered at the end of weld repair made by welding.

As seen in Figure 12b weld hardness can be drastically changed due to repair welding. The main reason can be a lower heat input demanded because of less space in the grooved part of weld, use of a harder electrode, too low preheating used etc. It is worth noting that hardness is the highest at both ends of repair. The reason is non-stationary thermal conditions at the beginning of the repair welding and at the end.

Hardening of weld materials caused by repair welding has an effect upon the fatigue strength of welds. It leads to its increase. In contrast, higher welding RS caused by repair welding lead to drastic fatigue strength reduction of welds. Grain growth during repair welding makes material of weld domains less sensitive to small defects.

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