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Surface Welding as a Way of Railway Maintenance

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1. Introduction

Since its early days the development of railway systems has been an important driving force for technological progress. From the 1840s onward a dense railroad network was spread all over the world. Within a few decades railway became the predominant traffic system carrying a steadily increasing volume of goods and number of passengers. This rapid development was accompanied by substantial developments in many areas such as steel production, engine construction, civil engineering, communication, etc (Zerbst et al., 2005). The railway industry worldwide is introducing heavier axle loads, higher vehicle speeds, and larger traffic volumes for economic transportation of goods and passengers. Increasing demands for high-speed services and higher axle loads at the turn of the 21st century account for quite new challenges with respect of material and technology as well as safety issues. The main factors controlling rail degradation are wear and fatigue, which cause rails to become unfit for service due to unacceptable rail profiles, cracking, spalling and rail breaks. Degradation of rail is microstructure and macrostructure sensitive and there is a complicated interaction between wear mechanisms, wear rates, fatigue crack initiation and growth rates, which affect rail life (Eden et al., 2005; Kapoor et al., 2002). Defects such as squats and wheelburns occur even in the most modern and well maintained railway networks and, as a broad general rule, every network develops one such defect each year, every two kilometers. At least one European railway network suffers almost 4000 rail fractures every year. Although such fractures are rarely dangerous when actively managed, they entail a high replacement cost and can be disruptive to the network (Bhadeshia, 2002). The replacement of such defects with a short rail section is expensive and not always desirable as it introduces two new discontinuities in the track in the form of two aluminothermic weld that destroy the advantages obtained with long hot-rolled rail.

Given that an average cost per repair or short replacement rail can run into several thousands of euros and that the occurrence of wheel rail interface defects is likely to increase with the evident increase in levels of traffic on most railways, the importance of the surface welding is easy to understand. Growing need for reparation due to large financial demands, have imposed research in this field.

Based on up-to date theoretical grounds and referencial facts, the aim of this paper is to show the possibilities of surface welding of the pearlitic high-carbon steel and the properties of the obtained joint. Discussion of the aquired results and conclusions indicate superior

properties of reparation welded layers in comparison to base steel. In repaired rail, maximal stresses are induced in newly deposited layer, i.e. new layer becomes area of future crack initiation, that in turn will delay its initiation and provide secure and reliable exploitation. This results open further possibilities for cheaper and reliable rail maintenance in future. Finally, this work shows clearly that repaired rails, due to improved microstructure and crack initiation resistance, have dominant mechanical properties in comparison to the original rails.

2. Rail degradation

There are many kinds of loadings which can adversely affect the life of rails; amongst these, wear and plastic deformation induced by contact stresses can combine to cause unacceptable changes in the rail head profile. Rails are subjected to complex stress state. There are many stresses that operate in a rail and can influence rail defects and rail failure. As bending and shear stresses arise principally from the gross vehicle load, the rail is also subjected to contact stresses, thermal stresses and residual stresses. Residual stresses in rails are introduced by different mechanisms. Primarily they stem from the manufacturing process, namely from heat treatment and roller straightening (Schleinzer & Fischer, 2000; Schleinzer & Fischer, 2001). A special case of residual stresses is welding residual stresses at rail joints. Since the loading conditions at the tread of a wheel and at the running surface of a rail have a number of features in common the appearance of cracks will also be similar. Cracks may be induced at or below the surface. Surface cracks are initiated due to high traction forces at high speed rails and they will propagate under the influence of a lubricant in an inclined angle in the direction of the motion of the applied load for rails operated in one direction. Transverse branching may then lead to the complete fracture of the rail. Sub-surface cracks are reported to initiate beneath the gauge corner 10–15 mm below the running surface and 6–10 mm from the gauge face (Clayton, 1994). They seem to propagate towards the rail surface and to behave like original surface cracks after penetration.

Note, that cracks close to or at the surface are a rather new problem connected with high speed operating. In former times rails experienced enough wear to permanently remove the surface layer containing the new emerging cracks. In order to fulfil the increasing demands for higher axial and dynamic loads modern rail steels tend to exhibit much higher resistance to wearing with the disadvantage that the surface layer removed is not any more large enough to prevent small cracks from extending into the rail (with respect to the development of rail steels (Muders & Rotthausen, 2000)).

A typical development of a rail crack is illustrated schematically in Fig. 1 (Ishida & Abe, 1996). Originating from a small surface or sub-surface crack, a dark spot is developing at the surface accompanied by crack growth in an inclined angle below the surface. At a certain point this crack branches into a horizontal and a transverse crack. The transverse crack will extend down into the rail and finally cause its fracture.

Today's rail failures can be divided into three broad groups as follows: those originating from rail manufacturing defects; those originating from defects or damage caused by inappropriate handling, installation and use and those caused by the exhaustion of the rail steel's inherent resistance to fatigue damage (Cannon et al., 2003).

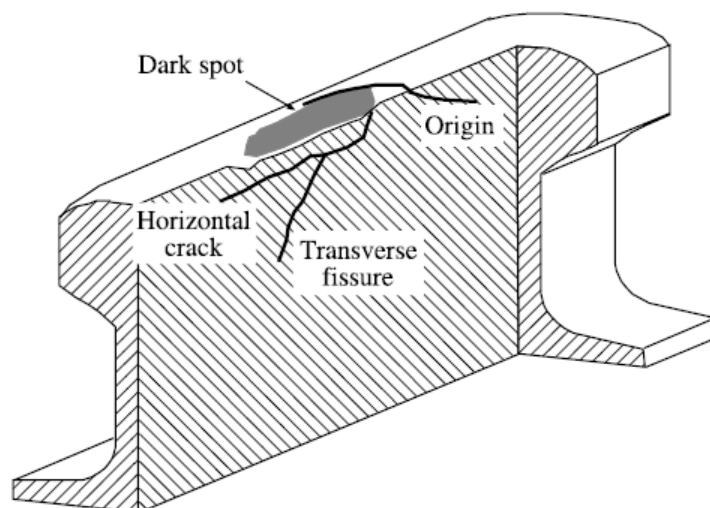


Fig. 1. Typical development of a rail crack (schematically) (Ishida & Abe, 1996).

Rolling contact fatigue (RCF) is likely to be a major future concern as business demands for higher speed, higher axle loads, higher traffic density and higher tractive forces increase. Head checks, gauge-corner cracks and squats are all names for surface-initiated RCF defects. They are caused by a combination of high normal and tangential stresses between the wheel and rail, which cause severe shearing of the surface layer of the rail and either fatigue or exhaustion of ductility of the material. The microscopic crack produced propagates through the heavily deformed (and orthotropic) surface layers of steel at a shallow angle to the rail running surface (about 10°) until it reaches a depth where the steel retains its original isotropic properties. At this stage the crack is a few millimetres deep into the rail head. At this point the crack may simply lead to spalling of material from the rail surface. However, for reasons still not clearly understood, isolated cracks can turn down into the rail, and, if not detected, cause the rail to break. These events appear to be rare, but are highly dangerous since RCF cracks tend to form almost continuously at a given site. Fracture at one crack increases stress in the nearby rail, increasing the risk of further breaks and disintegration of the rail (Cannon et al., 2003).

RCF initiation is not normally associated with any specific metallurgical, mechanical or thermal fault; it is simply a result of the steel's inability to sustain the imposed operating conditions. The problem is known to occur in most of the rail-steel types in common use today.

While wear has been reduced, rolling contact fatigue defects have become more prominent on busy routes where the rails are highly stressed. Although its wear reserve may not be used up, rail may have to be replaced because such defects quickly become critical for safety (Pointner & Frank, 1999). The relationship between RCF and mechanical wear is not well understood, as for example zero (or minimal) mechanical wear leads to significant microcrack propagation and thus RCF failure. On the other hand, excessive mechanical wear eliminates RCF but leads to unrealistically short rail life (Kapoor et al., 2001).

The rate of rail degradation depends also on the location; rail head erosion is at a maximum in regions where the track curves. In Fig. 2 is shown damage of the inner edge of rail head, caused by centrifugal force which tends to expel vehicle towards the outside of the track. Such damage can be repaired by surface welding, Fig 2b.

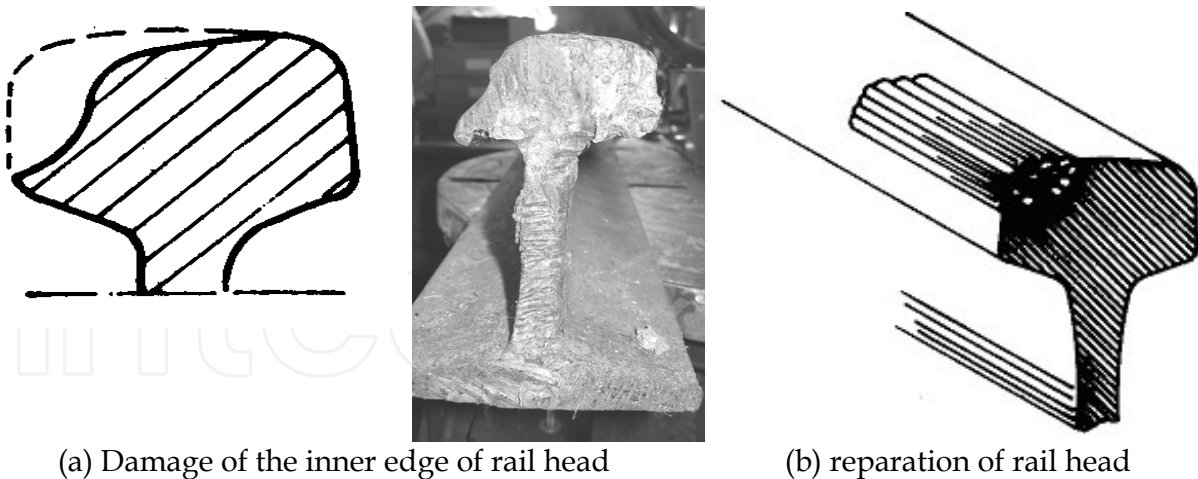


Fig. 2. Rail head degradation (Popovic et al., 2006).

2.1 Fracture control concepts

A few different fracture control concepts are applied in railway systems, and one of them is damage tolerance concept (Zerbst et al., 2005). Within the frame of this concept, the possibility of fatigue crack growth is basically accepted. The aim is to prevent the crack to grow to its critical size during the lifetime of the component, i.e. to estimate number of cycles to critical crack size. In fatigue, crack extension is expressed as a function of stress intensity range ΔK and the crack extension rate, da/dN , whereby da denotes an infinitesimal crack extension due to an infinitesimal number of loading cycles dN . The basic idea is that the largest crack that could escape detection is presupposed as existent. After that, the initial crack can extend due to various mechanisms such as fatigue, stress corrosion cracking, high temperature creep, or combinations of these mechanisms. Such a failure process is visible, and catastrophic rail failure can be prevented by regular examination of the top surface of the railhead. Maintenance methods (lubrication and grinding) help combat the wear and rolling contact fatigue phenomena referred to in local parameters. By applying these methods appropriately, maintenance costs can be reduced (Vitez et al., 2005). Rail grinding prolongs rail service life by preventing the emergence of defects or by delaying their development, preventive grinding to improve the quality of the running surface of newly-laid rails and corrective grinding to remove rail defects that have already developed by reprofiling the rail to optimize wheel/rail contact.

3. Rail steels

Choice of material for rail steels is of fundamental importance. This is because the rail's behaviour in service depends critically on the properties of the metal. Much effort and a considerable amount of research has already been undertaken in the search for the ideal rail steel (Pointner & Frank, 1999). In recent years rail steel production has improved as manufacturers have developed steels with increased hardness and better wear resistance.

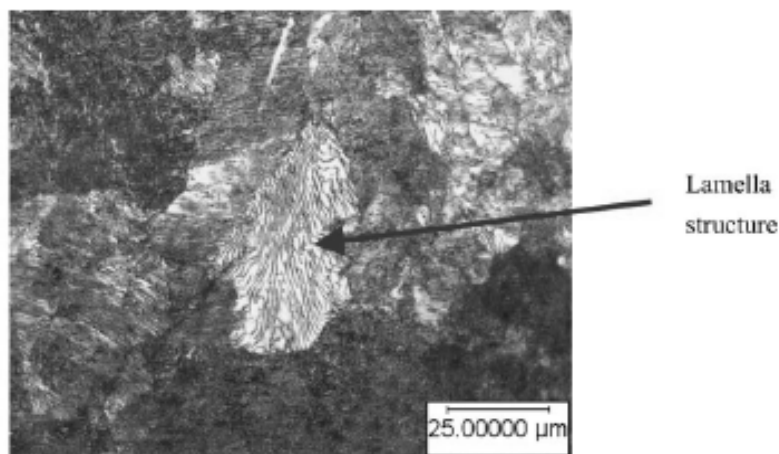
There are many criteria which determine the suitability of a steel for rail track applications. The primary requirement is structural integrity, which can be compromised by a variety of fatigue mechanisms, by a lack of resistance to brittle failure, by localised plasticity and by

excessive wear. All of these depend on interactions between engineering parameters, material properties and the environment. The track material must obviously be capable of being manufactured into rails with a high standard of straightness and flatness in order to avoid surface and internal defects which may cause failure. Track installation requires that the steel should be weldable and that procedures be developed to enable its maintenance and repair. Commercial success depends also on material and life time costs.

Since steel has one of the highest values of elastic modulus and shows superb strength, ductility and wear resistance, most modern rails have pearlitic microstructures and carbon-manganese chemistries similar to those produced in rails in 1900. Ordinary rail steels contain about 0.7 wt% of carbon and are pearlitic. Pearlite consists of a mixture of soft ferrite and a hard, relatively brittle iron carbide called cementite, Fig. 3a. Pearlite presumably achieves a high resistance to wear because of the hard cementite and its containment by the more plastic ferrite, but pearlitic steels are not therefore tough. In pearlite, altering lamellae of iron and iron carbide are arranged, and lamella spacing has a large effect on hardness. Naturally cooled standard rails have coarse lamella spacing and relatively low values of about 300 Brinell hardness (HB). Control-cooled premium rails have finer lamella spacing and thus higher hardness of 340-390 HB (Lee & Polycarpou, 2005).

Raising carbon content and refining pearlite spacing increases the hardness of pearlitic steel, and this has been shown to lead to improved wear resistance. Hence rail manufacturers have worked to produce pearlitic steels with higher carbon contents (now achieving approximately 1 wt%) and finer structure (using head-hardening processes). Even though hardness generally has a positive effect on rail wear, there is a limit to the hardness that can be reached with pearlitic steels, and this hardness has been reached in modern rails (Lee & Polycarpou, 2005).

There has been considerable effort devoted to finding alternatives to the pearlitic rails, but with alterable results. In an attempt to develop rail steels with higher hardness and alternative microstructures, several types of bainitic steel were developed. While pearlitic steels obtain their strength from the fine grains of pearlite, bainitic steels (Fig. 3b) derive their strength from ultra-fine structures with a lot dislocations which are harmless but confer high strength (Aglan et al., 2004). Bainitic steel is easy to be cast, welded and inspected by ultrasonic methods. The new generation of bainitic steels achieved higher tensile and fatigue strengths and performed well in service.



(a)

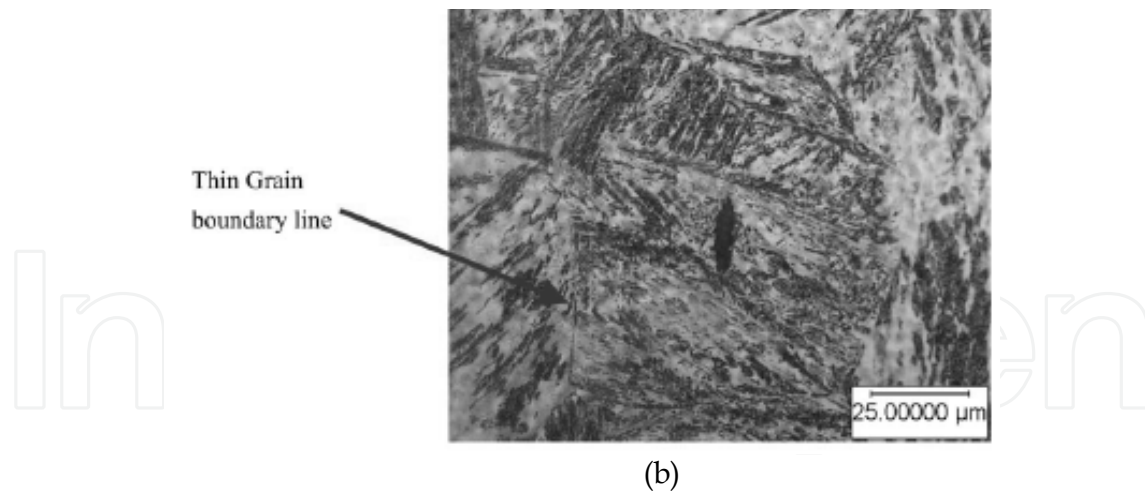


Fig. 3. Optical microstructures of rail steels: (a) pearlite; (b) bainite (Aglan et al.,2004).

4. Weldability of rails and types of filler materials

Main problem in welding of pearlitic steels is their poor weldability, i.e. susceptibility to welding defects, due to its high carbon equivalent. Since the rail is produced from this type of steel and subjected to complex strain state, leading to its degradation, surface welding is presently the dominant maintenance way to prolong exploitation life. Damaged parts produced from pearlitic high-carbon steel can be surface welded, in spite of their poor weldability, and by properly choice of welding technology, it is possible to get improved structure with dominant properties comparing to the original part (Popovic et al.,2010). To achieve that, it is necessary that obtained morphology corresponds to the new steel generation, i.e. bainitic microstructure.

For surface welding are mostly in use semi-automatic arc welding processes, with flux-cored and self-shielded wires. Basic difference between them is the first requires an external shielding gas, and the second does not. In both cases, core material acts as a deoxidizer, helping to purify the weld metal, generate slag formers and by adding alloying elements to the core, it is possible to increase the strength and provide other desirable weld metal properties (Lee,2001; Sadler,1997). These processes have replaced slowly MMA process and they almost ideal for outdoors in heavy winds. The key strength of these processes lies in the replacement of those aspects of the conventional MMA process that often results in variability in the quality of the repair with automatic and more controlled operations. Although the MMA process is used many industries, it is heavily reliant on the competence of the welder, is time consuming, and is prone to internal defects such as porosity that subsequently grow through fatigue, and if not detected by ultrasonic inspection, result in rail breaks.

The result of flux-cored wire application is higher quality welds, faster welding and maximizing a certain area of welding performance (Popovic et al.,2010). The number of layers in surface welded joint depends of the damage degree, most frequently it's three, sometimes with buffer layer. The buffer layer is applied at the crack sensitive materials, what high carbon steel certainly is (high CE). The function of buffer layer is to slow down the growth of initiated crack with its own plasticity. Constructions, like railways, are exposed to cyclic load and wear in exploitation life, so crack must be initiated. Sometimes in these cases it is necessary to use buffer layer, what besides the good affects, has some

drawbacks. Namely, the use of buffer layer significantly slows down surface welding process, due to replacement of wires and settings of other welding parameters. Since, as already noted, for surface welding are mainly in use semi-automatic and automatic processes, it significantly extends the working time. The new classes of flux-cored and self-shielded wires are recently developed, and it is possible to achieve the requested properties of welded joints without buffer layer (Popovic et al.,2011).

5. Experimental procedure

The material used in present work is pearlitic steel, received in the form of rails, type UIC 860 S49, what is the most common rail type on domestic railroads. It's chemical composition and mechanical properties are given in Table 1.

Chemical composition, %							Tensile strength R _m (N/mm ²)	Elongation A _c (%)
C	Si	Mn	P	S	Cu	Al		
0.52	0.39	1.06	0.042	0.038	0.011	0.006	680-830	≥14

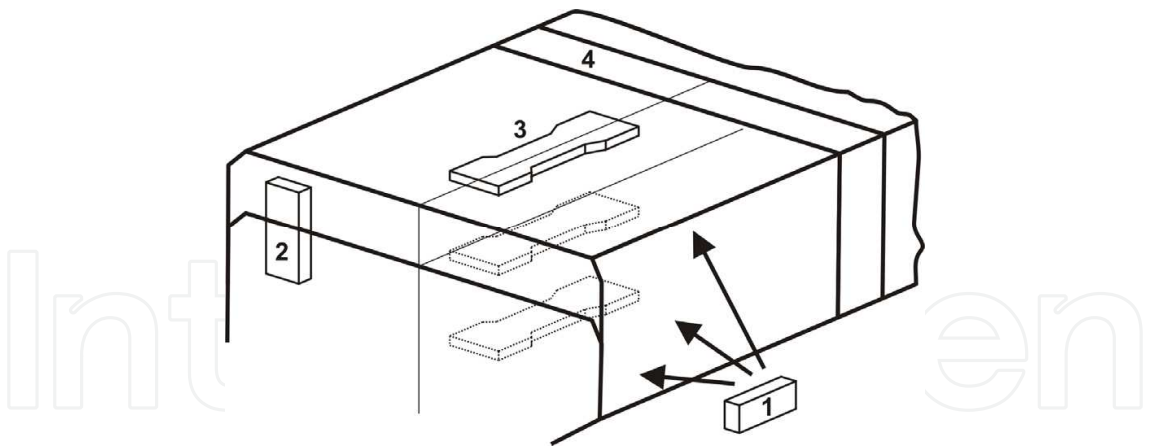
Table 1. Chemical composition and mechanical properties of base metal.

The surface welding of the testing plates was performed by semi-automatic process. As the filler material, the self-shielded wire (FCAW-S) and flux-cored wires (FCAW) were used, whose chemical compositions and mechanical properties are given in Table 2. The plates were surface welded in three layers; sample 1 with FCAW-S without buffer layer; sample 2 with FCAW with buffer layer (according to Table 2).

Sample No.	Wire designation		Wire diam. mm	Chemical composition						Hardness, HRC	
				C	Si	Mn	Cr	Mo	Ni		Al
Sample 1	OK Tubrodur 15.43 (self-shielded wire)		1.6	0.15	<0.5	1.1	1.0	0.5	2.3	1.6	30-40
Sample 2	1.layer (buffer layer)	Filtub 12B (flux-cored wire)	1.2	0.05	0.35	1.4	-	-	-	-	-
	2. and 3. layer	Filtub dur 12 (flux-cored wire)	1.6	0.12	0.6	1.5	5.5	1.0	-	-	37-42

Table 2. Chemical composition of filler materials.

Heat input during welding was 10 kJ/cm and preheating temperature was 230°C, since the CE equivalent was CE=0.64 (Popovic et al.,2010). Controlled interpass temperature was 250°C. Sample 1 is surfaced with one type of filler material (self-shielded wire), while for surfacing of sample 2 were used two types of wires, but both flux-cored: one for buffer layer and the second one for last two layers. As shielded gas for welding of sample 2, CO₂ was used. To evaluate the mechanical properties, specimens for further investigation were cut from surface welded rail head, according to Fig.4.



1- specimen for toughness and crack growth resistance estimation
2- specimen for microstructural analysis
3- tensile specimens
4- specimen for hardness measurements

Fig. 4. Specimens from surface welded rail head (Popovic et al.,2010).

5.1 Hardness

Hardness measurements were performed using a load of 100Pa. Hardness profiles of surface welded joints are shown in Fig. 5. The lowest hardness is in the base metal (250-300 HV), being the hardness of naturally cooled standard rails(Lee & Polycarpou, 2005; Singh et al., 2001). In HAZ hardness increase is noticeable in both samples, due to complex heat treatment and grain refinement (Popovic et al.,2010). In sample 2 comes to a sharp decrease of hardness in first surfaced layer, i.e. in buffer layer. The function of buffer layer is to stop the growth of initiated crack with its own plasticity and reduced hardness. The hardness of II and III welded layers of both samples are the highest and similar, due to influence of alloying elements in filler materials, which shift transformation points to bainitic region⁴. Maximum hardness level of 350-390 HV is reached in surface welded layers and it provides improvement of mechanical properties and wear resistance.

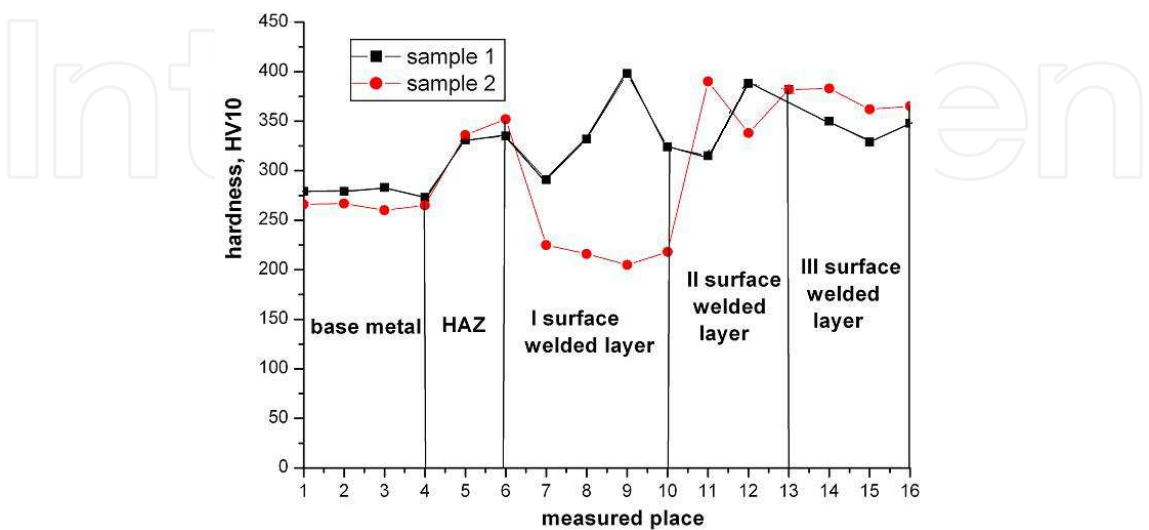
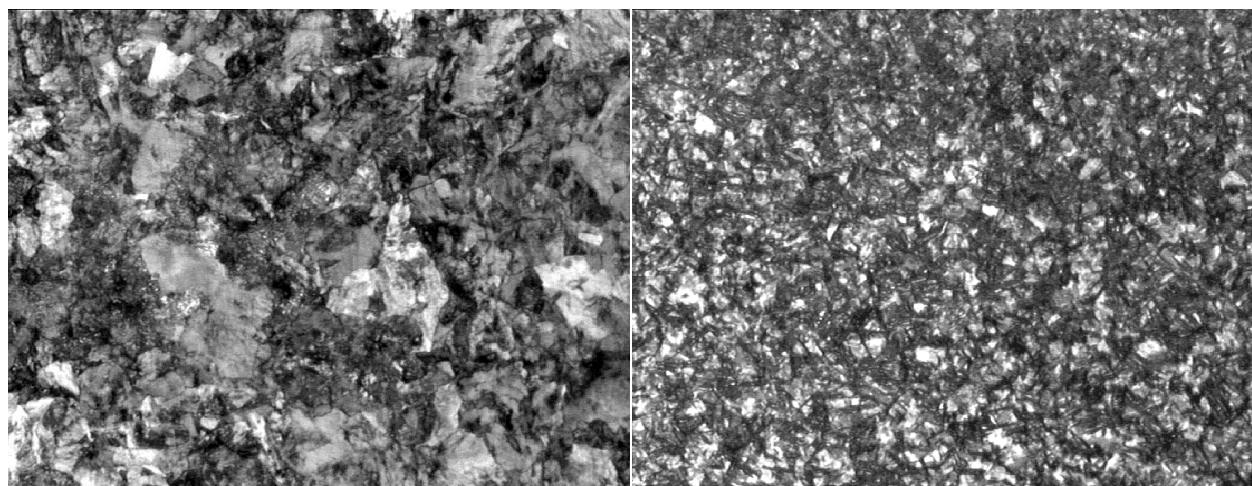


Fig. 5. Hardness profiles along the joint cross-section of samples (Popovic et al.,2011).

5.2 Microstructure

Microstructural analysis of all characteristic zones of welded layer has been done. Heat affected zone (HAZ) also has pearlitic microstructure, but with finer grain, than base metal (Figure 6), so its structure is improved and it is not a critical place in weldment. That is result of thermomechanical treatment of HAZ which is re-heated three times. Structural compatibility between deposit metal and base metal was achieved and martensitic layer wasn't formed.

The greatest differences appear in first layer microstructure, Fig.7. First layer microstructure of sample 1 consists of ferrite, pearlite and bainite, what is result of mixing of low-alloyed filler material with high-carbon base metal. For first layer deposition of sample 2 is used low-carbon wire alloyed with Mn, as a function of buffer layer, so characteristic structure consist of great fraction of ferrite with relatively large primary grains. Beside proeutectoid ferrite, microstructure contains Widmanstätten and acicular ferrite (Popovic et al.,2007).



(a) 200x

(b) 200x

Fig. 6. Microstructure of a) base metal and b) HAZ of both samples (Popovic et al.,2007).

The second layer microstructure is the most important in surface welded joint, because it has the greatest influence on mechanical and technological properties and exploitation behavior of repaired parts. For this structure is characteristic larger fraction of bainite, consequence to the less mixing with base metal. In second layer of sample 2 occurs fine grain ferritic structure with low content of bainite. This structure has finer grain compare to first layer, what is result of heat treatment and chemical composition (presence of Mo in filler material).

The third layer of sample 1 has some coarser grain structure, with higher content of bainite, compare to previous layer, what is consequence of re-heating absence. For third layer of sample 2 is characteristic bainitic microstructure with small amount of martensite and locally zones of proeutectoid ferrite.

Though used filler materials are different type, alloying concepts, sort of protection, buffer layer, as final result is obtained desirable bainitic microstructure with superior properties compare to base metal (Popovic et al.,2007). Except metallography examination, this is confirmed by other detail tests (Popovic,2006).

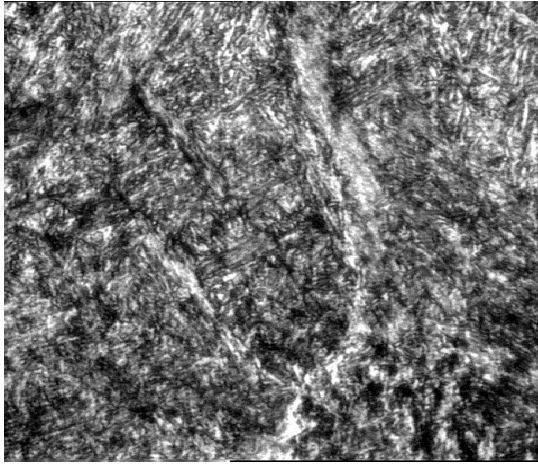
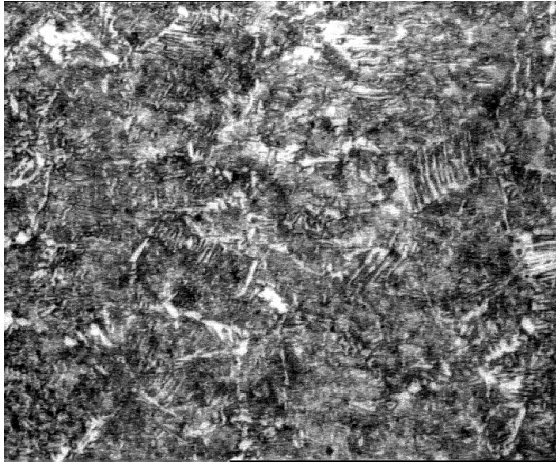
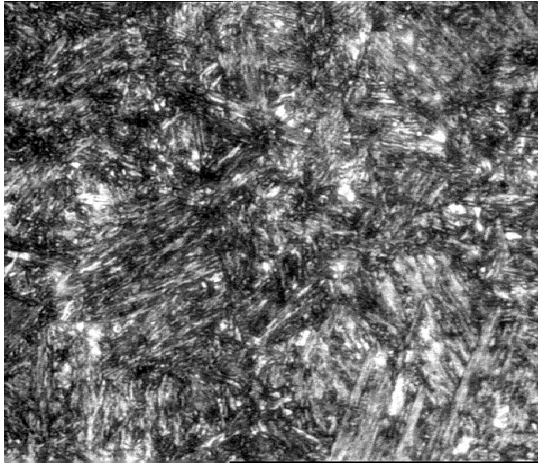
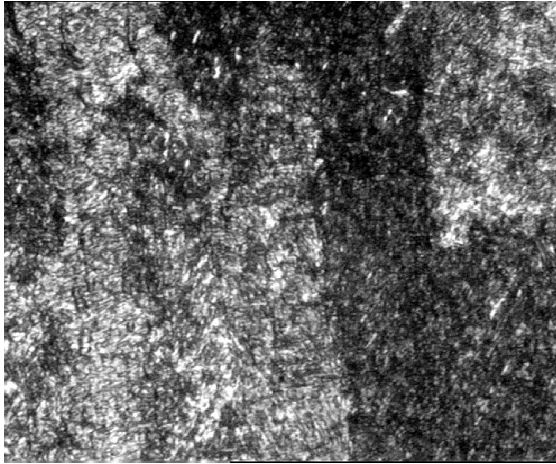
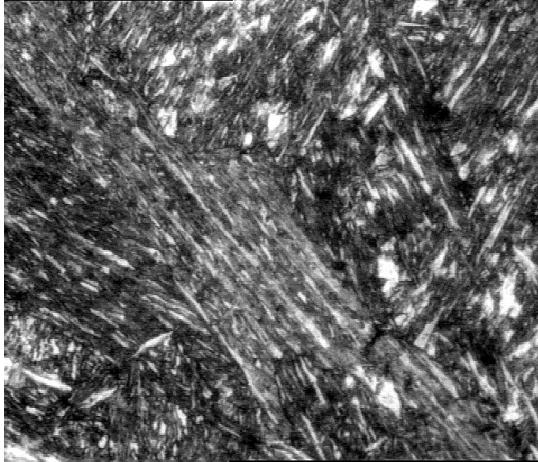
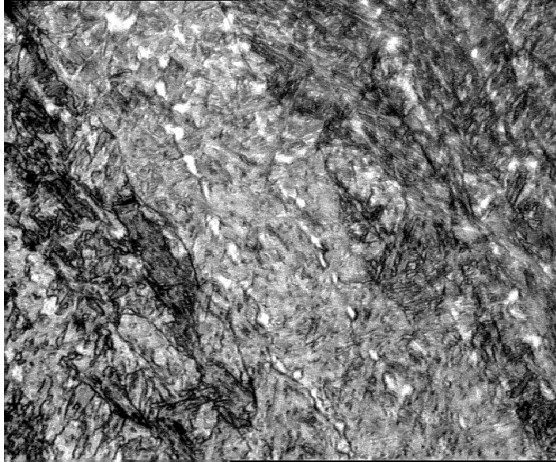
	Sample 1	Sample 2
1. layer	 500 x	 500 x
2. layer	 500 x	 500 x
3. layer	 500 x	 500 x

Fig. 7. Microstructure of all surface welded layers (Popovic et al.,2007).

5.3 Tensile tests

The tensile tests were conducted on a 2 mm thick specimens. The room temperature mechanical properties (ultimate tensile strength, UTS) of the surface welded joint are shown in Figure 8. The basic requirement in welded structures design is to assure the required strength. In most welded structures this is achieved with superior strength of WM compared to BM (overmatching effect), and in tested case this is achieved (Burzic & Adamovic,2008; Manjgo et al.,2010). The highest UTS is in weld metal of sample 2 (1210 MPa), due to solid state strengthening by alloying elements.

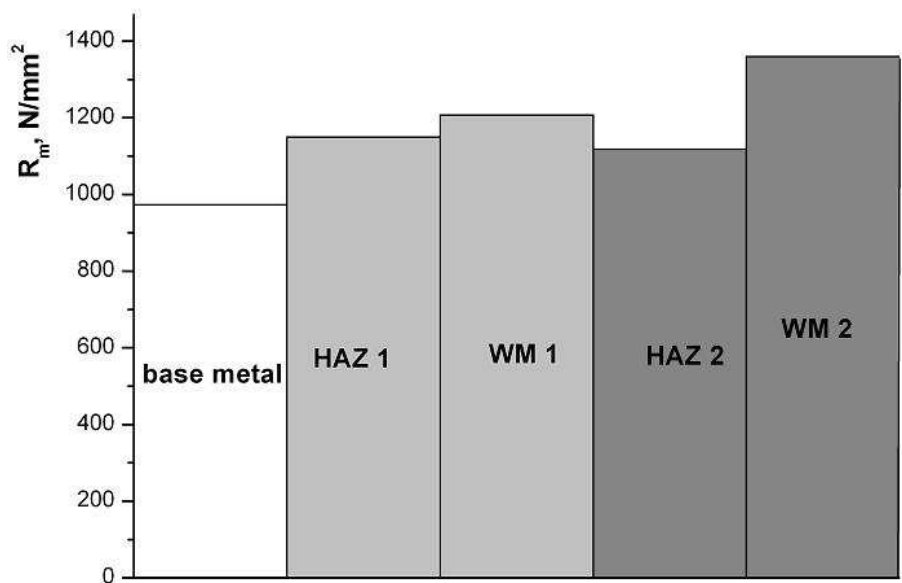


Fig. 8. Ultimate tensile strength of the surface welded joints (Popovic et al.,2011).

5.4 Impact testing

Impact testing is performed according to EN 10045-1, i.e ASTM E23-95, with Charpy V notched specimens, on the instrumented machine SCHENCK TREBEL 150 J. Impact testing results are given in Table 3 for base metal and HAZ at all testing temperatures. Total impact energy, as well as crack initiation and crack propagation energies, for weld metal of both samples at all testing temperatures (20°C, -20°C and -40°C) are presented in Table 4 and in Figure 9.

The total energy of base metal is very low (5 J), due to very hard and very brittle cementite lamellae in pearlite microstructure (Popovic et al.,2011), while the toughness of HAZ is higher (11-12 J) and is similar for both samples at all testing temperatures.

	Total impact energy, E_u , J		
	20°C	-20°C	-40°C
base metal	5	3	3
sample 1-HAZ	12	11	10
sample 2-HAZ	11	10	9

Table 3. Instrumented impact testing results of Charpy V specimens for base metal and HAZ at all testing temperatures.

	sample 1-WM			sample 2- WM		
	20°C	-20°C	-40°C	20°C	-20°C	-40°C
Total impact energy, E_u , J	29	23	17	34	14	11
Crack initiation energy, E_{in} , J	20	16	15	12	10	10
Crack propagation energy, E_{pr} , J	9	7	2	22	4	1

Table 4. Instrumented impact testing results of Charpy V surface weld metal specimens at all testing temperatures.

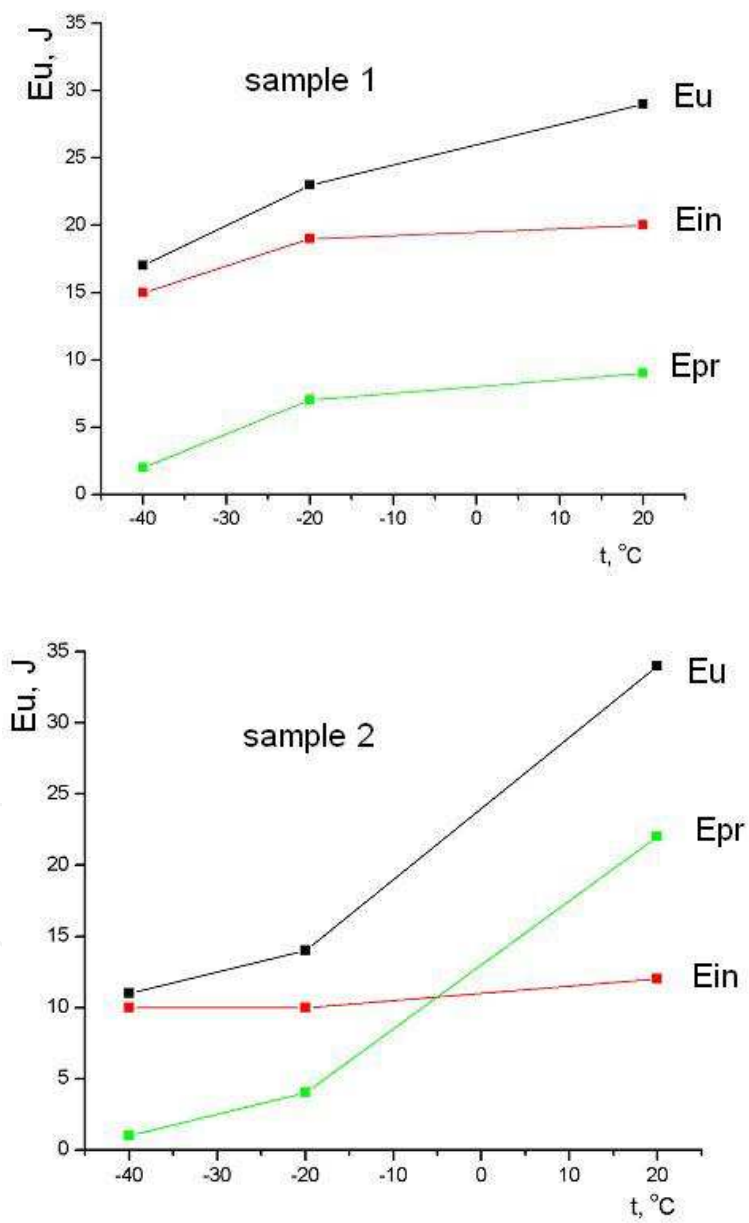
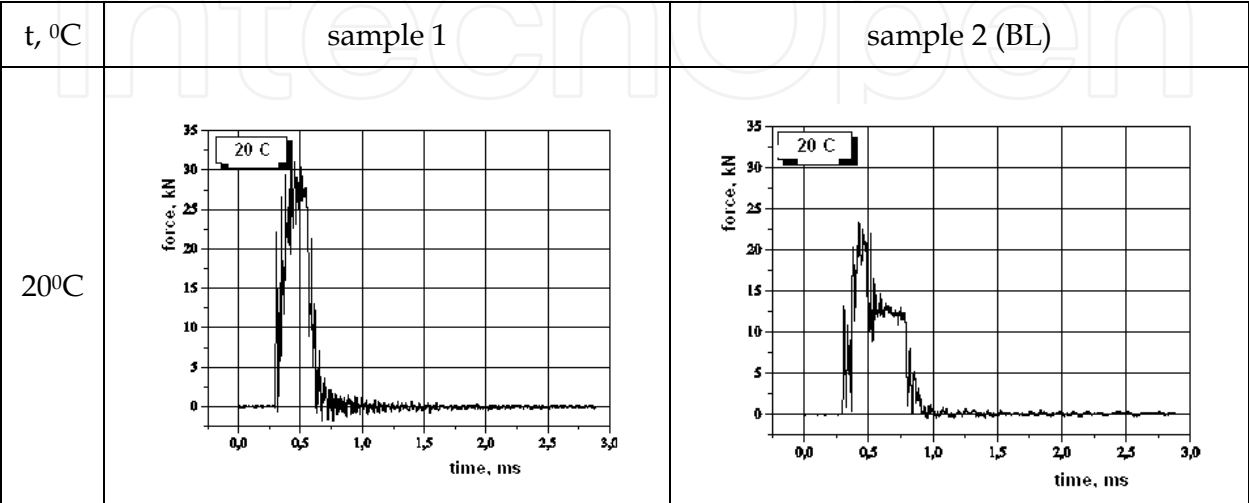


Fig. 9. Dependence total impact energy, crack initiation and crack propagation energy vs.temperature for (a) weld metal of sample 1 and (b)weld metal of sample 2 (Popovic et al.,2011).

Values of total impact energy of samples 1 and 2 at room temperature are significantly higher (29 J and 34 J) than in base metal (5 J), as a consequence of appropriate choice of alloying elements in the filler material. The presence of Ni, Mn and Mo promotes the formation of needled bainitic microstructure and grain refinements, and increases the strength and toughness also(Popovic, 2006). By analyzing the impact energy values of sample 1, a change of toughness in continuity is observed, with no marked drop of toughness, and for all tested temperatures, crack initiation energy is higher than crack propagation energy. This is the reason for the absence of significant decrease of toughness. The highest value of total impact energy was found for the sample 2 at room temperaure (34 J), which is the only case when the initiation energy is lower than propagation energy (12 J and 22 J, respectively). This shown practically the buffer layer function. Namely, the initiated crack during propagation comes to plastic buffer layer, which slows down crack further growth. For this reason, the crack propagation energy is the largest part of total impact energy. However, at -20°C, significant drop of total impact energy is noticable (14 J) due to losing of buffer layer plastic properties at lower temperatures. The low-carbon wire (0.05%C i 1.4%Mn) has excellent toughness, but and marked rapid drop on S-curve (dependence toughness vs. temperature). Transition temperature of this material above -20°C is confirmed by the obtained impact toughness results. The use of buffer layer is reasonable if the exploatation temperature is above -5°C; on the contrary, at lower temperatures, buffer layer is losing its function and the toughness is decreased (Popovic et al.,2011).

Diagrams force-time, obtained by instrumented Charpy pendulum, are given in Figure 10. As can be seen, for the sample 1 the character of diagrams force-time changed little by lower temperature. Namely, this material at room temperature has diagram with marked rapid drop, as consequence of unstable crack growth. After the maximum load, a very fast crack growth is started, and it is confirmed by the low value of crack propagation energy(Grabulov et al.,2008). On the contrary, on the sample 2 diagram at room temperature, the presence of buffer layer is clearly shown. The initiated crack, during its growth, comes to buffer layer which temporary stops the further crack growth and changes crack growth rate. The obtained experimental diagram doesn't belong to any type, according to standard EN 10045-1. This leads to toughness increase, primarily crack propagation energy, and it is also here the only case when the crack initiation energy is lower than crack propagation energy.



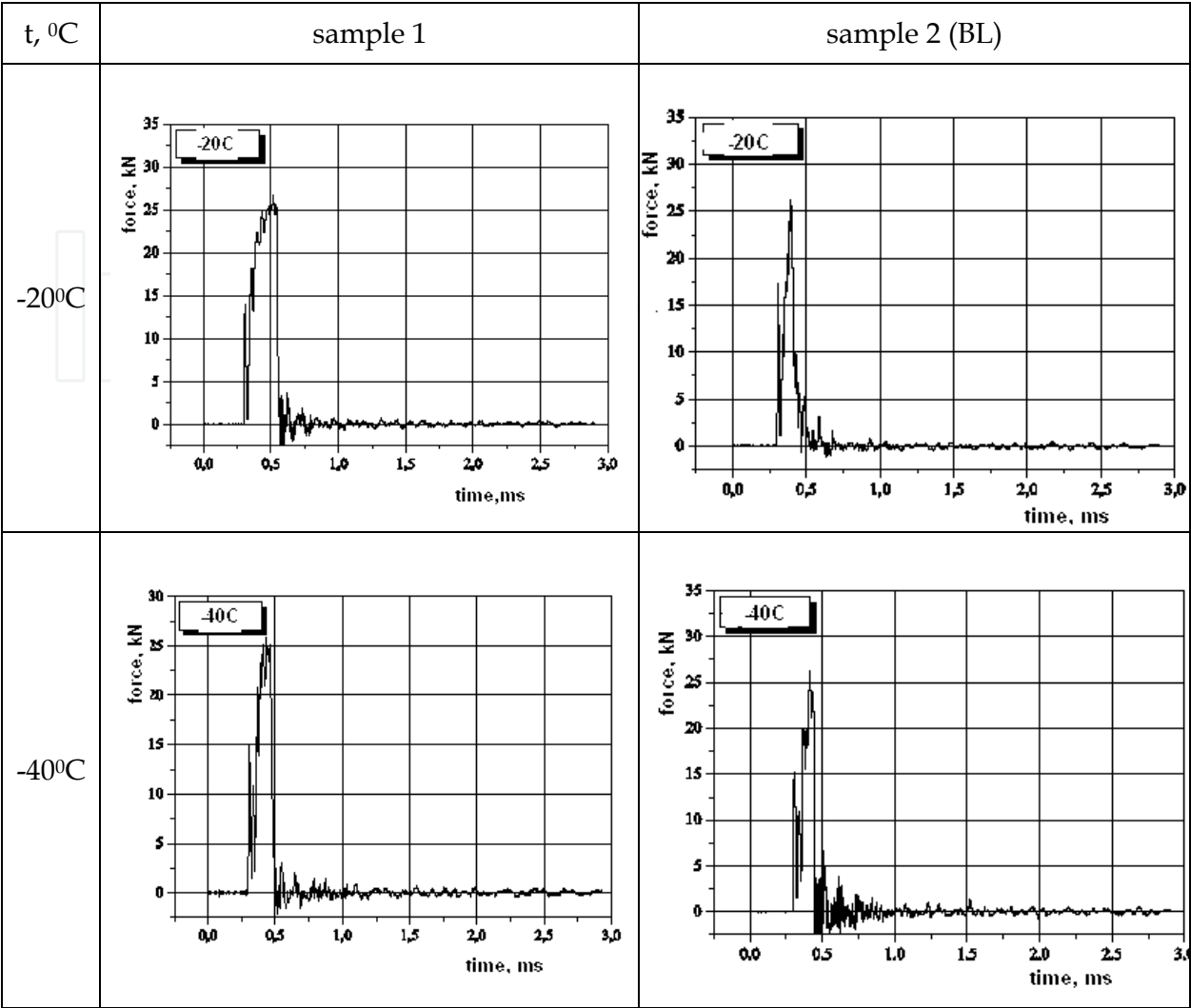


Fig. 10. Diagrams force-time, obtained by instrumented Charpy pendulum for sample 1 and sample 2 (Popovic et al.,2011).

5.5 Crack growth rate

A basic contribution of fracture mechanics in fatigue analysis is the division of fracture process to crack initiation period and the growth period to critical size for fast fracture (Burzic & Adamovic,2008). Fatigue crack growth tests had been performed on the CRACKTRONIC dynamic testing device in FRACTOMAT system, with standard Charpy size specimens, at room temperature, and the ratio $R=0.1$. A standard 2 mm V notch was located in base metal and in third layer of WM, for the estimation of parameters for BM, WM and HAZ, since initiated crack will propagate through those zones. Crack was initiated from surface (WM) and propagated into HAZ, enabling calculation of crack growth rate da/dN and fatigue treshold ΔK_{th} . The results of crack growth resistance parameters, i.e., obtained relationship da/dN vs. ΔK for base metal, sample 1 and for sample 2 are given in Figure 11 and 12. Parameters C and m in Paris law, fatigue threshold ΔK_{th} and crack growth rate values are given in Table 5 for all samples as obtained from relationships given in Figures 11 and 12, for corresponding ΔK values.

The behaviour of welded joint and its constituents should affect the change of curve slope in validity part of Paris law. Materials of lower fatigue-crack growth rate have lower slope in the diagram da/dN vs. ΔK . For comparison of the properties of surface welded joint constituents the crack growth rates are calculated for different values of stress-intensity factor range ΔK .

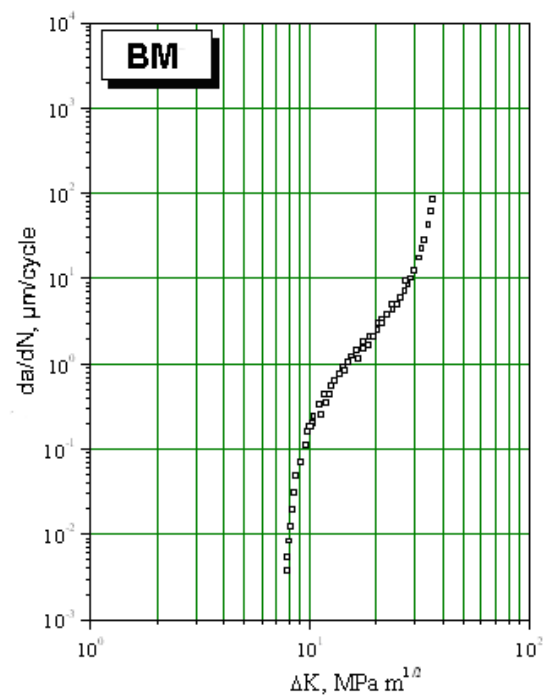


Fig. 11. Diagram da/dN vs. ΔK for base metal.

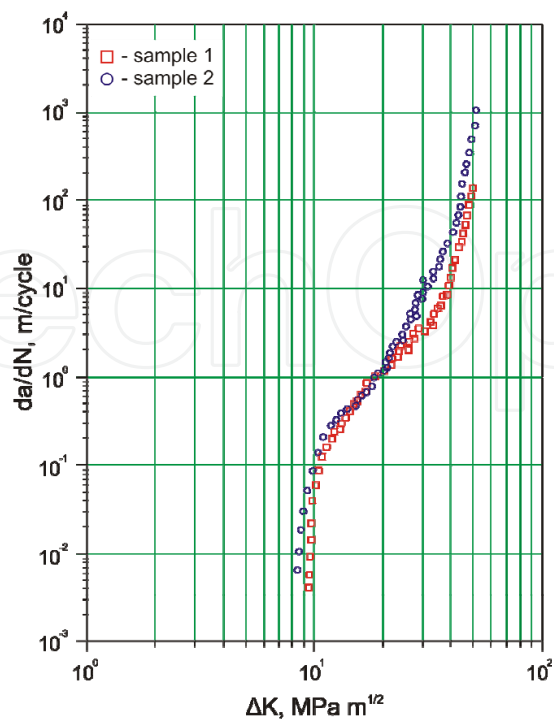


Fig. 12. Diagram da/dN vs. ΔK for sample 1 and sample 2.

	Zone of surface welded joint	Fatigue threshold ΔK_{th} , MPa m ^{1/2}	Parameter C	Parameter m	Crack growth rate da/dN, m/cycle			
					$\Delta K=10$ MPam ^{1/2}	$\Delta K=15$ MPam ^{1/2}	$\Delta K=20$ MPam ^{1/2}	$\Delta K=30$ MPam ^{1/2}
Base metal	BM	8.0	$3.31 \cdot 10^{-11}$	3.28	$6.31 \cdot 10^{-08}$			
sample 1	WM 1	9,5	$4.45 \cdot 10^{-13}$	3.74		$1.11 \cdot 10^{-08}$	-	-
	WM 2		$3.78 \cdot 10^{-13}$	3.61		-	$1.88 \cdot 10^{-08}$	-
	HAZ		$4.07 \cdot 10^{-13}$	3.79		-	-	$1,61 \cdot 10^{-07}$
sample 2	WM 1	8,9	$4.63 \cdot 10^{-13}$	3.87		$1.65 \cdot 10^{-08}$	-	-
	WM 2		$3.85 \cdot 10^{-13}$	3.88		-	$2.07 \cdot 10^{-07}$	-
	HAZ		$3.76 \cdot 10^{-13}$	3.93		-	-	$1.18 \cdot 10^{-06}$

Table 5. Parameters C, m, ΔK_{th} and crack growth rate values for all zones of surface welded joints.

Bearing in mind that weld metal consists of two layers (third layer is used for V notch), as referent values of ΔK were taken: $\Delta K = 10$ MPa m^{1/2} for BM, $\Delta K = 15$ MPa m^{1/2} for WM1, $\Delta K = 20$ MPa m^{1/2} for WM2, and $\Delta K = 30$ MPa m^{1/2} for HAZ. It's important that all selected values are within a middle part of the diagram, where Paris law is applied. The crack growth rate in base metal is 3-4 times higher than in both weld metal layers, i.e. the growth of initiated crack will be slower in weld metal layers. This means that for the same value of stress intensity factor rang ΔK , base metal specimen needs less number of cycles of variable amplitude than weld metal specimen, for the same crack increment.

In all three zones of surface weleded joint (WM2, WM1 and HAZ), sample 2 with buffer layer has higher crack growth rate than sample 1, i.e. the growth of initiated crack will be slower in sample 1. This means that for the same value of stress intensity factor rang ΔK , specimen of sample 2 needs less number of cycles of variable amplitude than specimen of sample 1, for the same crack increment. The maximum fatigue crack growth rate is achieved in HAZ for both samples, when stress intensity factor range approaches to plane strain fracture toughness.

If a structural component is continuously exposed to variable loads, fatigue crack may initiate and propagate from severe stress raisers if the stress intensity factor range at fatigue threshold, ΔK_{th} , is exceeded(Burzic & Adamovic,2008). Fatigue treshold value ΔK_{th} in base metal ($\Delta K_{th} = 8$ MPa m^{1/2}) is lower than fatigue treshold value ΔK_{th} in weld metal of both metal. Fatigue treshold value ΔK_{th} for sample 2 ($\Delta K_{th} = 8.9$ MPa m^{1/2}) is lower than that for sample 1 ($\Delta K_{th} = 9.5$ MPa m^{1/2}). This means that crack in sample 2 will be initiated earlier, i.e. after less number of cycles, than in sample 1.

Values of fatigue threshold and crack growth rates corespond to initiation and propagation energies in impact testing, and in this case, good corelation is achieved (Popovic, 2006). Sample 1 has higher crack initiation energy (20 J) and higher ΔK_{th} (($\Delta K_{th} = 9.5$ MPa m^{1/2} for

sample 1 i $\Delta K_{th} = 8.9 \text{ MPa m}^{1/2}$ for sample 2). With comparison of crack propagation energy and crack growth rate, it is hard to establish the precise analogy, as toughness was estimated for the surface weld metal, whereas crack growth rate for each surface welded layer. Generally, buffer layer didn't show slow the initiated crack growth, with aspect of crack growth rate, while this effect is obvious in the case of toughness, i.e. crack propagation energy (Popovic et al, 2011).

6. Conclusion

On the base of obtained experimental results and their analysis, the following is concluded:

1. The experimental investigation of surface welded joints with different weld procedures has shown, as expected, significant differences on their performance in terms of mechanical properties. But, in both cases, it was shown, that in spite of poor weldability of high carbon steel, they can be successfully welded. Structural compatibility between deposite metal and base metal was achieved and martensitic layer wasn't formed. Obtained HAZ has better structure compare to base metal.
2. The filler material is relevant parameter which affects on deposite layer quality. Work with self-shielded wires is more simple, specially for outdoor applications. Both used wires are on high technological level and can be recommended for reparation of high-carbon steel damaged parts. Final microstructure is the result of different influences: type of filler material, heat input, degree of mixture with previous layer and post heat treatment with subsequent surface layer. It is necessary to know all these factors and also to know the way of affect. Though applied wires are with different alloying concepts, result in both cases is that initial pearlitic morphology is replaced by final desirable bainitic microstructure. It was shown that, by selecting corresponding parameters, it is possible to obtain the morphology of the best properties.
3. The maximal hardness level of 350-390 HV is reached in surface welded layers of both samples, with equal hardness of base metal (250-300 HV). The main difference appears in the first deposition layer, where as expected, in sample 2 the hardness is significantly lower (buffer layer). The obtained hardness values ensure simultaneously the improvement of mechanical and wear properties, and in the case of a rail, represents maximal hardness preventing the wheel wear (Popovic et al., 2010). Similar results are obtained by tensile testing. Sample 2 has slightly higher ultimate tensile strength (1360 MPa) than sample 1 (1210 MPa) due to solid solution strengthening by alloying elements.
4. The most improved results are obtained for impact properties. The toughness of base metal is 6-7 times lower than the toughness of weld metal, and more than twice lower than toughness of HAZ. For welding with buffer layer, at -20°C , the drop of total impact energy is significant, due to lowering of buffer layer plastic properties at lower temperatures. The transition temperature of this material is above -20°C , and it was confirmed by obtained impact toughness results. The use of buffer layer is beneficial for exploitation temperature above -5°C . On the contrary, at lower temperatures, buffer layer loses its function and toughness decreases. On the contrary, for sample 1 the change of toughness is continous and without marked drop of toughness. At all tested temperatures, the crack initiation energy is higher than crack propagation energy. This

- may be the reason for the absence of significant decrease of toughness and that should be kept in mind during design and exploitation.
5. The results show that base metal is characterized by a lower fatigue threshold than weld metal, i.e. a 3 to 4 times higher crack growth rate. This means that crack will initiate more likely in base metal, and that it requires fewer cycles to reach the critical size. Contrary to a typical welded joint, a surface welded layer is the safest place for crack initiation.
 6. Values of fatigue threshold and crack growth rates correspond to initiation and propagation energies in impact testing. In the case of fatigue threshold and crack initiation energy, good correlation was achieved. Sample 1 has higher crack initiation energy (20 J) and higher ΔK_{th} (9.5 MPa $m^{1/2}$) than sample 2 (12 J and $\Delta K_{th} = 8.9$ MPa $m^{1/2}$). On the contrary, buffer layer didn't show decrease of initiated crack growth rate, as this effect is obvious in the case of toughness, i.e. crack propagation energy. Since the constructions from high-carbon steel are used at low temperature, and bearing in mind the extended working time, in modern surface welding technologies, the use of buffer layer is not recommended.
 7. Testing results of base metal and surface welded layer represent typical behavior of two steel microstructure-pearlitic and bainitic, what is confirmed through microstructural investigation. It has been showned that, thanks to appropriate choice of filler material and welding technology, surface welding of damaged parts is not only a way of reparation, but and a way of improvement of starting properties

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