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SOME FACTORS AFFECTING FATIGUE RESISTANCE OF WELDS

NEKI FAKTORI KOJI UTIČU NA OTPORNOST ZAVARA PREMA ZAMORU

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Abstract

The fatigue strength of polycrystalline metals is hardness and defect size dependent. The fatigue strength of welds depends on the size of grains and actual local mean stress, too. Local mean stress is influenced by the global R-ratio and static preloading. The welds in the as-welded condition are pre-stressed due to the existence of welding residual stresses. The sign and magnitude of residual stresses depend on the welding conditions. Changes of residual stresses, hardness and defect size are experimentally determined in the present work. Samples of studied metals are prepared by using either thermal cycle simulator or laboratory furnace and water quenching. The methods of microstructure preparation and type of load result in actual local R-ratio at the used weld toe models. Specimens are cyclic loaded in the similar way as steel at the weld toe. The bottom of the notch is either defect-free or defected by Vickers indentation. Final results show that coarser grain and higher local R-ratios lower the fatigue strength of welds.

Ključne reči
- CrNiMo čelici
- zavar
- otpornost na zamor
- zona uticaja toplote
- male greške zavarivanja

Izvod

Zamorna čvrstoća polikristalnog metala zavisi od tvrdoće i veličine greške. Zamorna čvrstoća zavara takođe zavisi od veličine zrna i delujućeg lokalnog srednjeg napona. Na lokalni srednji napon utiču ukupni R odnos i statičko preopterećenje. Zavari u zavarenom stanju su prethodno opterećeni zaostalim naponom od zavarivanja. Predznak i veličina zaostalih napona zavise od uslova zavarivanja. Promene zaostalog napona, tvrdoće i veličine greške su eksperimentalno određene u prikazanom radu. Uzorci od proučavanih metala su pripremljeni korišćenjem simulatora termičkih ciklusa ili laboratorijskih peći i kaljenja u vodi. Metoda pripreme mikrostrukture i vrsta opterećenja omogućili su određivanje delujućeg R-odnosa u modelima podnožja zavara. Epruvete su ciklično opterećivane na način sličan opterećenju čelika u podnožju zavara. Dno zareza je bilo bez grešaka ili oštećeno utiskivanjem na Vikersu. Konačni rezultati su pokazali da je u slučaju grubljih zrna i većeg lokalnog R-odnosa zamorna čvrstoća zavara manja.

INTRODUCTION

The fatigue damage limits load carrying capacity of welded structures subjected to variable stress. Material fatigue always results from fluctuating stress. Local fatigue damage is often the principal cause of premature failures. Fatigue failures of huge welded constructions are usually catastrophic, leading to severe property damage, environment pollution, and even loss of human lives, /1-3/.

Actually, the fatigue damage shortens service life of highly loaded structural components. When the fatigue damage accumulates to an appropriate level, the fatigue crack initiates at the surface of pre-existing defects usually at stress concentrators such as notches, holes and welds. Once formed, the propagation rate of fatigue cracks is influenced by the load history, microstructure and environmental factors.

A great number of factors affect the behaviour of welded joints under cyclic loading as are material strength, weld shape, the sign, magnitude and frequency of stressing as well as the temperature. Processing is not often considered although it determines the homogeneity of weld materials and surface quality as well as the sign and distribution of welding residual stresses (RS). Thus, the way of processing, and in particular, its metallurgical effects have an overriding influence on the performance of cyclic loaded welded components.

FACTORS AFFECTING WELD FATIGUE CRACKING

Typical factors affecting the fatigue cracking of welds are summarised in the following sections.

The loading pattern must contain minimum and maximum with large enough variation. Peak stress levels must be of sufficiently high value. If peak stresses are too low, no crack initiation will occur. The material must be subjected to a large number of stress cycles. Several properties of the stress state need to be considered, such as stress amplitude, mean stress, biaxiality, in-phase or out-of-phase shear stress, and load sequence. The fatigue strength of anisotropic materials depends on the principal stress direction.

The geometry of welds needs to be considered as transitions, notches, variation in cross section and all others that lead to stress concentration. Because the stress raiser in the form of inclusion is not so effective, inclusion fatigue is rare. The surface quality is also important. Roughness causes microscopic stress concentrations and lower fatigue strength. The compressive RS can be introduced in the surface of the most sensitive parts of welds by number of techniques. They produce surface RS in compression which increase the fatigue life of welded component in the case of high-cycle fatigue.

Fatigue life varies for different steels, due to the specific crystal lattice, strength and fatigue resistance. High strength steels are more notch-sensitive than mild steels.

Environmental conditions can cause erosion, corrosion, or gas-phase embrittlement, which all affect fatigue life. Corroded parts form pits that act like notches. Corrosion reduces the amount of material and lowers the strength because it increases the actual stress. Corrosion fatigue is a problem encountered in many aggressive environments. The decarburization is also an important factor because it weakens the surface by making it softer. Stresses due to bending and torsion are highest at the surface.

Higher temperatures decrease the fatigue strength.

Cutting, welding and other manufacturing processes involving heat or plastic deformation can produce high level of tensile RS which decreases the fatigue strength. The RS which add to the design stress may easily exceed the limit stress as imposed in the initial design.

The size and distribution of internal defects as gas porosity, non-metallic inclusions and shrinkage voids can significantly reduce fatigue strength.

For most metals, smaller grains yield longer fatigue life; however, presence of surface defects or scratches has a greater influence than in a coarse grained alloy.

Results of the present work contribute to understanding how the defect size (DS), grain size (GS) and RS sign and level could affect the fatigue strength of butt-welds.

DIRECT INFLUENCES OF WELDING

The result of unavoidable heat effect during welding is the heat-affected zone (HAZ) and weld metal (WM) formation. They are weakest links in the welded structure.

Large temperature gradients are the consequence of very fast local material heating when welding and pretty fast cooling, too. The reduced ability of solid materials to contract during cooling and especially the occurrence of low temperature phase transformation give rise to the appearance of RS in welds. For this reason some parts of welds can become susceptible to hydrogen embrittlement and other detrimental phenomena. Final results of RS appearance are various types of weld cracking, distorted joints or simply shorter service life of welded structure.

RS are self-balancing forces. Their character in the most sensitive weld domains to fatigue damage is often tensile. Due to welding, some weld domains have totally new or changed microstructure and properties. So, those domains can be sensitive to material fatigue and statically tensile pre-stressed, earlier fatigue crack initiation can occur, and crack propagation can be faster. The final consequence is shorter life of the welded structure.

Volume weld defects act as stress raisers while planar defects as stress intensifications. The effects of the first are represented in experimentally determined and statistically evaluated S-N curves of different types of welds. Effects of cracks have to be evaluated by using the fracture mechanics approach, linear-elastic or elastic-plastic, depending on the crack size and stress level.

The resistance of steels to fatigue crack initiation and its growth depends on the microstructure. Base metals (BM) are chosen with regard to properties, actually due to the convenient microstructure. The microstructure of both the as-welded condition is the result of chemical composition and applied welding thermal cycle/cycles.

FATIGUE STRENGTH OF BUTT-WELDS WITH SMALL DEFECTS

A butt-weld is a weld with the highest effectiveness. In the case of cyclic loading, butt-weld cracking is found at weld toes where stresses are concentrated. If cracks initiated earlier it would be of the greatest importance for fatigue strength when conditions for crack propagation are fulfilled, too. The load-carrying capacity of cyclic loaded welded structures depends, therefore, on the actual strength of the most demanding welds.

Small weld defects as are different inclusions, scratches or cracks, are often found at weld toes of butt-welds. Even sharp transitions between weld reinforcement and BM can be treated as a small defect. Anyway, the fatigue cracks at the toe of butt-welds initiate often in coarse-grain material, i.e. the coarse grain HAZ.

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

Murakami and co-workers treated the influence of various shaped small defects in the same way as cracks, i.e. using LEFM, /4-6/. The parameter reflecting the effect of small defects on the fatigue strength of metals is found to be the square root of the defect projection onto the plane perpendicular to cyclic stress ($\sqrt{\text{area}}$). LEFM underestimates propagation rates of short cracks within the local plastic zones that can develop as a result of stress concentrations in welds, /7, 8/.

Natural small weld defects in metals can be artificially modelled. As small surface defects, the drilled small holes are used in the past with success when the endurance limit of real quality metals was studied. Vickers indentation is a promising small artificial surface defect because indenting with the Vickers pyramid is easy to perform. Load on the pyramid has to be adjusted according to material hardness and the expected indentation size, /9, 10/.

The problem with use of artificial weld defects is the additional local RS appearance as the result of local material plastic deformation when indenting.

In order to evaluate only the effects of small defect to the fatigue strength, global RS due to processing, i.e. welding, should be omitted. Electro-etching is usually used to move surface stratum with the highest RS without any plastic deformation. In such a way, the local and global surface RS are lowered. Unfortunately, local RS lowering is closely linked to the defect's shape and size change. Both influence the parameter of small defect quantified as $\sqrt{\text{area}}$. Global RS cannot be lowered separately.

EXPERIMENTAL WORK AND RESULTS

Two kinds of CrNiMo steels are used in the experimental work, steels A and B (Table 1).

Coarse grain microstructure is prepared by simulating thermal conditions in the BM close to the weld on pieces of steel A of size 15×8.5×70 mm by cycle simulator, /11/. The peak temperature of the thermal cycle was at least 1350°C while cooling time $\Delta t_{8/5}$ around 5 and 10 s. The result of the simulation was either purely martensitic microstructure or mainly martensitic microstructure with a small portion of bainite. Tensile strength, R_m, of materials found at the weld toe and grain size, GS, are shown in Table 2.

Bending fatigue strength is experimentally determined on notched specimens of size 14.5×8×70 mm supported at 59 mm. They are machined from samples with simulated microstructure HAZ_1, HAZ_2, HAZ_3, HAZ_4 and HAZ_5 (Table 2). The shape and size of the specimens are shown in Fig. 1a. The stress concentration factor caused by notch in bending is 1.74, /9/. Artificial surface defects of different sizes and shapes are prepared at the bottom of the notch by indenting with a Vickers pyramid. As sketched in Fig. 2, they are single indentations of different sizes and a series of five indentations in line. The average sizes of single indentations, d, are 105, 160, and 221 μm while the average length of the series, l, is 386 and 692 μm. Local RS in the surroundings of the defects caused by indenting are present. Global RS is supposed to be zero because notches are machined after the welding thermal cycle simulation.

Cyclic specimen loading is executed at global R-ratio close to zero. All results of fatigue strength tests, $2\sigma_f$, are shown in Table 2. The influence of local RS due to indenting with Vickers pyramid is studied in the next part of the research.

Purely martensitic HAZ_{Simul} microstructure is prepared on pieces of steel B using thermal cycle simulator and two different heat treatments in furnace. Peak temperature of simulated welding thermal cycle was 1300°C while cooling time $\Delta t_{8/5} \cong 5$ s. The result is coarse grain. The same microstructure is our goal in modelling single-step heat treatment in furnace, namely the coarse grain $HAZ_{I\text{-}step}$. Two-step heat treatment in furnace is modelled for the $HAZ_{II\text{-}step}$ simulation. The result of such a simulation is a refined purely martensitic microstructural formation. The hardness is the same, but grains are finer, /10/.

Global compressive RS are the result of water quenching cylindrical steel pieces. Their level is assessed by hole-drilling method around 300 MPa, /13/. In contrast to quenching, heat conveying from steel pieces by means of water cooled grips welding in the thermal-cycle simulator results in global tensile RS. We did not quantify the level of those RS at the bottom of the notch, but they are tensile.

Table 1. Chemical composition of steels A and B (mass %).
Tabela 1. Hemijski sastav čelika A i B (maseni %)

Steel	C	Si	Mn	P	S	Cr	Ni	Mo	V	Al	Ti	Cu
A	0.09	0.27	0.30	0.015	0.010	1.05	2.63	0.27	0.07	0.045	0.026	-
B	0.18	0.22	0.43	0.012	0.028	1.56	1.48	0.28	-	0.023	-	0.15

Table 2. Tensile strength, R_m, grain size, GS, and bending fatigue strength, σ_f, at R = +0.02–0.04 of simulated steel A samples.
Tabela 2. Zatezna čvrstoća, R_m, veličina zrna, GS, i savojna zamorna čvrstoća , σ_f, za R = +0.02–0.04 simuliranih uzoraka od čelika A

Material	R_m [MPa]	GS [μm]	$2\sigma_f$ [MPa]					
			smooth	$d \cong 105$ μm	$d \cong 160$ μm	$d \cong 221$ μm	$l \cong 386$ μm	$l \cong 692$ μm
HAZ_1	1210	130	938	915	905	834	-	769
HAZ_2	1171	140	950	882	-	814	-	769
HAZ_3	1192	180	927	905	-	814	-	724
HAZ_4	1176	140	927	927	882	769	-	724
HAZ_5	1172	180	915	905	-	814	834	724

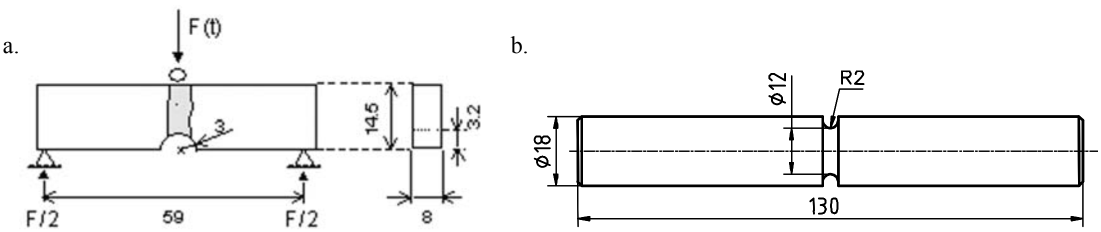

Figure 1. Specimens for fatigue strength determination: the in-plane bending of specimens (a); rotary bending specimens (b).
Slika 1. Epruveta za određivanje zamorne čvrstoće: savijanje epruvete u ravni (a); obrtno savijanje epruvete (b)

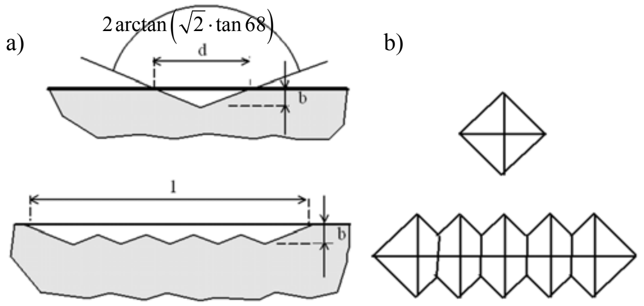

Figure 2. Single (a) and series (b) of Vickers indentations.
Slika 2. Pojedinačni (a) i niz (b) otisaka Vikersa

Single Vickers indentations are used as artificial surface defects (Fig. 2b). Local RS due to indenting are present when the Vickers pyramid is indented after quenching. There are no local RS when Vickers pyramid is indented before local quenching, /10/. This is possible only on specimens with microstructure of the $HAZ_{II\text{-}step}$.

Cyclic loading of specimens with HAZ_{Simul}, $HAZ_{I\text{-}step}$ and $HAZ_{II\text{-}step}$ microstructures are performed at global R-ratio –1. Results of fatigue strength tests, σ_f, are presented in Table 3. Cyclic loading of specimens with $HAZ_{II\text{-}step}$ microstructure are executed at R-ratio +0.1, too. Results of fatigue strength tests, $2\sigma_f$, are also presented in Table 3.

Table 3. Tensile strength, hardness, grain size, and bending fatigue strength of steel B specimens with simulated microstructure.
Tabela 3. Zatezna čvrstoća, veličina zrna i savojna zamorna čvrstoća epruveta od čelika B sa simuliranom mikrostrukturom

Material	R_m MPa	Hardness HV10	GS µm	RS_s MPa	Fatigue strength			
					–		$d \cong 200$ µm, $\lvert RS_l \rvert \geq 0$	$d \cong 200$ µm, $\lvert RS_l \rvert = 0$
						MPa		
HAZ_{Simul}	–	466	200	tensile	σ_f	330	330	–
$HAZ_{I\text{-}step}$	1366	452	200	≅ –300	σ_f	537	–	–
$HAZ_{II\text{-}step}$	1431	455	20–30	≅ –300	σ_f	760	664	678
					$2\sigma_f$	1119	998	953

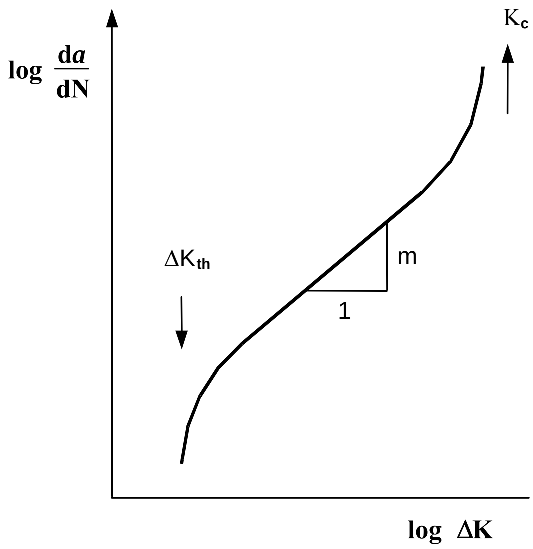

Figure 3. Long fatigue crack propagation rate.
Slika 3. Brzina rasta dugačke zamorne prsline

Bending and tensile fatigue strength is experimentally determined on cylindrical specimens, ∅18×130 mm with a little bit different notches in order to have the same stress concentration factor 1.74, /12/. They are machined from samples with simulated microstructures HAZ_{Simul}, $HAZ_{I\text{-}step}$ and $HAZ_{II\text{-}step}$. The shape and size of specimens are shown in Fig. 1b. Artificial surface defects are made by indenting Vickers pyramid at the bottom of the notch. The average size of indentations, d, was 200 µm.

Crack propagation rate is measured on specimens with simulated microstructures of pure martensitic HAZ. Resonant testing machine Cracktronic with crack length measuring system Fractomat is used. Threshold stress intensity range for long cracks ΔK_{th}, and constants used C and m are experimentally determined (Fig. 3).

Linear part of log-log dependency is the Paris equation

$$\frac{da}{dN} = C(\Delta K)^m$$

Specimens with microstructures HAZ_{Simul}, $HAZ_{I\text{-}step}$ and $HAZ_{II\text{-}step}$ are tested at R-ratio –1, whilst specimens with microstructure $HAZ_{II\text{-}step}$ at R-ratio +0.1, too. The results are shown in Table 4.

Table 4. Fatigue crack propagation parameters.
Tabela 4. Parametri rasta zamorne prsline

Material	R_m MPa	Hardness HV10	GS µm	R	ΔK_{th} MPa·m$^{0.5}$	C	m
HAZ_{Simul}	–	466	200	–1	13	$1.7\ 10^{-13}$	3.6
$HAZ_{I\text{-}step}$	1366	452	200	–1	12	$1.7\ 10^{-15}$	4.6
$HAZ_{II\text{-}step}$	1431	455	20–30	–1	17	$1.1\ 10^{-13}$	3.5
				+0.1	15	$3.6\ 10^{-13}$	3.2

DISCUSSION

Data from Tables 2 and 3 are shown in Figs. 4 and 5. Figure 4a shows fatigue strength, σ_f, over defect size, DS, quantified as $\sqrt{\text{area}}$. We can sketch three simple dependences that correspond to different grain sizes and R-ratios. In general, fatigue strength decreases when defect size increases. The exception are results obtained with very coarse grain and extremely low local R-ratio.

Fatigue strength of polycrystalline metals with small defects is two-parameter dependent, /7-9/:

$$\sigma_f = 1.43(\text{HV}+120)/\left(\sqrt{\text{area}}\right)^{\frac{1}{6}}$$

The fatigue strength, σ_f in MPa, hardness, HV in Vickers hardness number, defect size $\sqrt{\text{area}}$ in μm.

Data from fatigue strength of specimens with indentations in Fig. 4a are presented in Fig. 4b in log-log scale. The regression curve that corresponds to medium grain size HAZ materials and local R-ratio close to zero seems to be linear as predicted in Murakami's formula.

Obviously, the fatigue strength of polycrystalline steel at the weld toe in welds made in carbon steels is not only two-parameter dependent. We see in Fig. 4a much different fatigue strength of steels with almost the same hardness around 200 HV and defect size 53 μm. It seems that grain size and global RS level are significant, too. This is of the greatest importance for understanding load carrying capacity of butt-welds under cyclic loading if processing parameters, actually welding parameters, are considered.

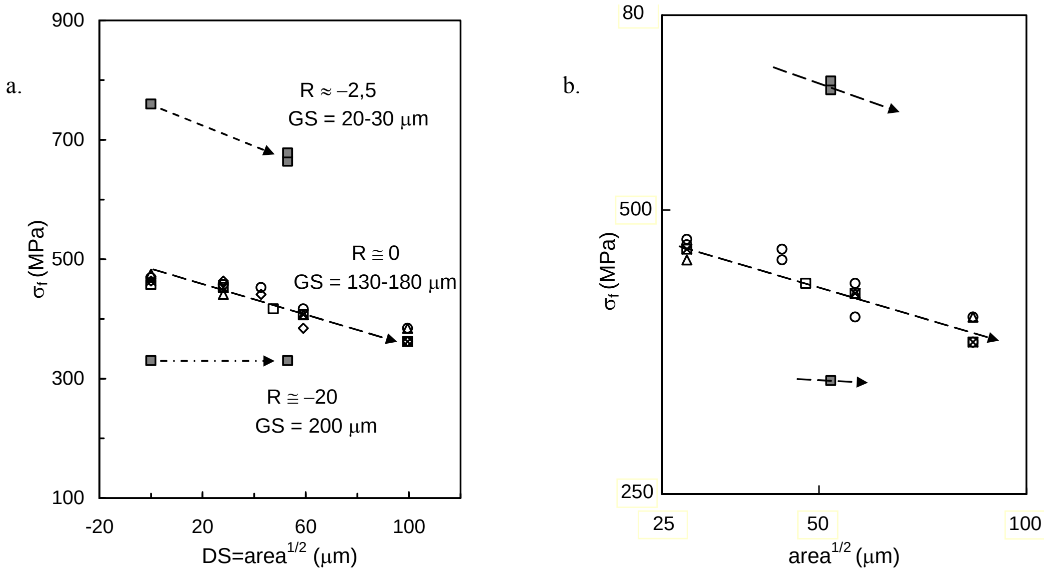

Figure 4. Fatigue strength, σ_f, versus defect size, DS: linear scale (a); log-log scale (b).
Slika 4. Zavisnost zamorne čvrstoće, σ_f, i veličine greške, DS: u linearnoj razmeri (a); u log-log razmeri (b)

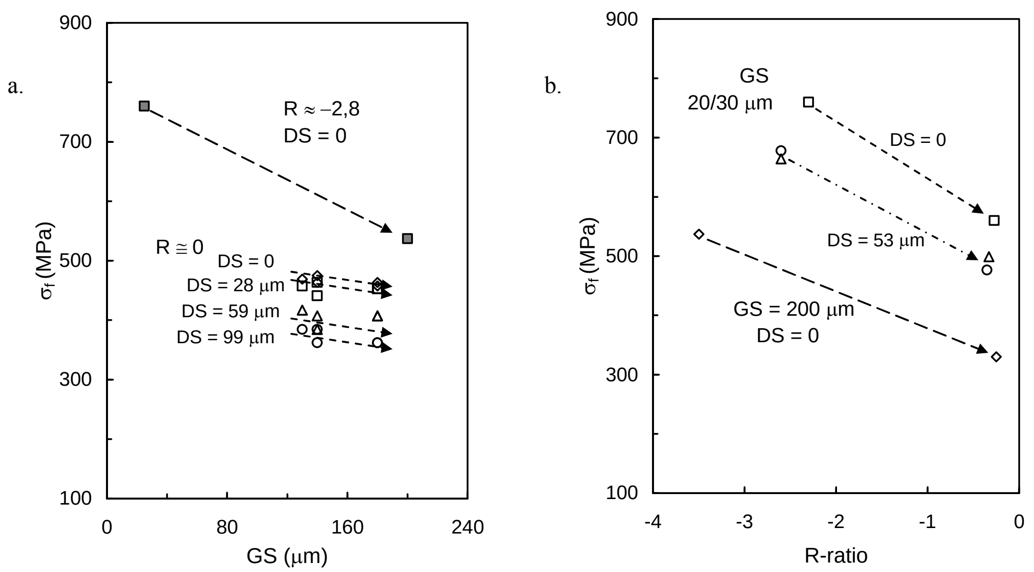

Figure 5. Fatigue strength, σ_f, versus grain size, GS (a) and R-ratio (b), in respect.
Slika 5. Zavisnost zamorne čvrstoće, σ_f, i veličine greške, DS (a) i R-odnosa (b), respektivno

Miller /11/ pointed out the responsibility of microstructural obstacles for existence of non-propagating cracks at stress level equal to fatigue strength. The most important obstacles link with the biggest microstructural units of metals are grains, colonies of pearlite in steels with ferritic/pearlitic microstructure, domains with harder microstructural constituent in steels with duplex microstructure.

Considering only distances between grain boundaries, average grain size is crucial. Grain boundaries are more sparsely distributed in space in coarse grain microstructure than in the fine grain. This is the reason that a once initiated crack that is short can stop to propagate and become a non-propagating crack, /14/.

We can try to explain effects of each variable separately. The dependence of fatigue strength, σ_f, on grain size, GS, at two different actual R-ratios, caused by combination of global R-ratio and effects of RS is shown in Fig. 5a. As expected, the fatigue strength decreases with increasing grain size, /9/. The dependence of fatigue strength, σ_f, on R-ratio at different grain and defect sizes is shown in Fig. 5b. The fatigue strength decreases with increasing R-ratio like in Smith's and Goodman's diagrams.

Let us consider experimental data in Table 4. Long fatigue crack propagation should be described by using LEFM. In comparison with the properties of refined martensitic microstructure at R = –1, lower ΔK_{th} values are registered when coarse grain martensitic microstructures are tested. However, lower ΔK_{th} values are registered in refined martensitic microstructure at R = +0.1 than in case of R = –1.

In case of coarse grain microstructure with lower ΔK_{th} value, the fatigue crack that has initiated from small defect can hardly become non-propagating. Cracks start to propagate at lower stress levels. In case of fine grain microstructure with higher ΔK_{th} value the fatigue crack that has initiated from small defect can become easily non-propagating. Namely, cracks in finer grains start to propagate at significantly higher stress levels.

Higher *C* and *m* parameters in the Paris equation indicate faster long crack propagation, resulting in shorter fatigue life. Lower *C*-values for both coarse grain microstructures is linked with higher exponent *m* (Table 4). The same situation is when we compare the same parameters between coarse and fine grain martensitic microstructures.

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