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MICROMECHANICAL STUDY OF DUCTILE FRACTURE INITIATION AND PROPAGATION ON WELDED TENSILE SPECIMEN WITH A SURFACE PRE-CRACK IN WELD METAL

MIKROMEHANIČKA STUDIJA INICIJACIJE I ŠIRENJA DUKTILNOG LOMA U ZATEZNOJ ZAVARENOJ EPRUVETI SA POVRŠINSKOM PRSLINOM U METALU ŠAVA

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- ductile fracture
- welded tensile specimen
- complete Gurson model
- micromechanical approach

Ključne reči
- duktilan lom
- zavarena zatezna epruveta
- potpuni Gurson model
- mikromehanički pristup

Abstract

In this paper, crack growth initiation and propagation is predicted in the weldment using micromechanical approach and material properties determined by experimental and numerical procedure. The welded tensile specimen with a surface pre-crack in the weld metal is experimentally and numerically analysed. High-strength low-alloyed steel is used as base metal in quenched and tempered condition. Crack growth initiation values and J-R curves are experimentally and numerically obtained for the specimen. The complete Gurson model (CGM) is used in prediction of the J-R curve and crack growth initiation. Results show that the resistance to crack initiation and growth can be predicted using micromechanical analysis and material properties determined by experimental and numerical procedure.

Izvod

U ovom radu je obavljena procena inicijacije i širenja prsline u zavarenom spoju primenom mikromehaničkog pristupa, kao i osobine materijala eksperimentalnim i numeričkim postupcima. Zavarena zatezna epruveta sa unetom površinskom prslinom u metalu šava je analizirana numerički i eksperimentalno. Upotrebljen je niskolegirani čelik povišene čvrstoće u kaljenom i otpuštenom stanju, kao osnovni materijal. Veličine inicijacije rasta prsline i J-R krive su za datu epruvetu dobijene eksperimentalno i numerički. Primenjen je potpuni Gurson model (CGM) u proceni J-R krive i inicijacije rasta prsline. Rezultati pokazuju da se otpornost na inicijaciju i rast prsline može proceniti primenom mikromehaničke analize i osobina materijala dobijenih numeričkim i eksperimentalnim postupcima.

INTRODUCTION

Crack initiation and stable growth in ductile materials are conventionally characterized by J-R curves obtained from standard fracture mechanics tests. However, testing of the same types of welded specimens (notched in different positions) and loading conditions has revealed considerable differences in the J-R curves, due to the constraint caused by microstructural and mechanical heterogeneity, /1-3/. Therefore, transferring fracture parameters from specimens to components is questionable. Constraint effect is very important in homogeneous structures, where the fracture resistance depends on the geometry of the structure and on the crack, /4, 5/. Moreover, recently produced high strength steels typically exhibit large-scale deformation and plastic straining during tearing. This helps to prevent rapid unstable fracture. However, such fracture behaviour cannot be accurately predicted using existing correlations that are characterized by the J-integral. In the presence of large-scale yielding, the traditional J-integral approach to elastic-plastic fracture mechanics is known to become inaccurate or even inapplicable for engineering purposes as it cannot adequately characterize the crack tip stress field, /6/. Therefore, more accurate characterizations of defects in welded high strength steels are of particular interest to provide more accurate failure assessments.

Using local damage approach to model crack initiation and propagation in ductile materials seems to be the solution for the transferability problem in fracture mechanics. This approach can simulate the physical processes of void nucleation, growth and coalescence of investigated material using continuum mechanics. The advantage of the local damage model, compared with conventional fracture mechanics, is that the model parameters are only material, and not geometry-dependent. Therefore, the local damage approach is chosen to study the effect of heterogeneity on crack initiation and propagation, which are investigated in this study.

The ductile fracture process is often modelled by means of the Gurson model, /7/, which describes the progressive degradation of material load carrying capacity. This model has been later modified by Tvergaard and Needleman, /8, 9/; the resultant model is not intrinsically able to predict coalescence, and is only capable of simulating nucleation and growth of the micro-void. This deficiency is solved by introducing an empirical void coalescence criterion: coalescence occurs when a critical void volume fraction, f_c, is reached. The value of f_c is introduced in the model as a constant value which depends on material. Several studies have shown that different values of f_c can be obtained from the same tension tests, /10/. Therefore, the complete Gurson model (CGM) has been introduced by Zhang, /11/, who combined the Gurson-Tvergaard-Needlman (GTN) model and the coalescence criterion proposed by Thomason, /10/. The CGM has considered critical void volume fraction, f_c, is not a material constant but it depends on the stress state and the geometry.

The complete Gurson model /11/ has been shown to give accurate predictions for different levels of stress triaxiality, for both strain non-hardening and strain hardening materials, and is therefore selected to assess the fracture behaviour of welded joints in this work. Welded tensile specimens have been modelled; cracks in the middle of the weld metal (WM) are considered. The aim of this work is to predict ductile crack initiation and propagation of high strength steel weldments using the micromechanical model. Experimental work and a three-dimensional finite damage model for welded tensile specimens with a pre-crack in WM is performed. The crack initiation value and J-R curve for the tensile specimen with pre-crack have been obtained experimentally and numerically.

THE COMPLETE GURSON MODEL (CGM)

Micromechanical models have been recently developed for modelling the behaviour of ductile materials. Among these models, the micromechanical model proposed by Gurson is considered as most widely used for ductile porous materials. The yield function of Gurson, /7/, modified by Tvergaard, /12, 13/, and Tvergaard and Needlman, /8, 14/, is used to describe the evolution of void growth and subsequent macroscopic softening. The modified yield function is defined by the formula:

$$\varphi(q,\sigma_m,\bar{\sigma},f)= \\ =\left(\frac{q}{\bar{\sigma}}\right)^2+2q_1f^*\cosh\left(\frac{3q_2\sigma_m}{2\bar{\sigma}}\right)^2-\left(1+(q_1f^*)^2\right)=0 \quad (1)$$

where σ_m is the mean stress, $\bar{\sigma}$ is the flow stress of the matrix material, f^* is the modified void volume fraction, and q is the von Mises effective stress:

$$q=\sqrt{\frac{3}{2}S_{ij}S_{ij}} \quad (2)$$

where S_{ij} stands for deviatoric components of the Cauchy stress. The constants q_1 and q_2 are fitting parameters introduced by Tvergaard, /12/, to improve the ductile fracture prediction of the Gurson model. The modified void volume fraction, f^*, is the damage function /8/ given by:

$$f^*=\begin{cases} f & \text{for } f\le f_c \\ f_c+\dfrac{f_u^*-f_c}{f_F-f_c}(f-f_c) & \text{for } f>f_c \end{cases} \quad (3)$$

where f_c is the critical void volume fraction at the moment of occurring void coalescence; $f_u^* = 1/q_1$ is the ultimate void volume fraction; and f_F is the void volume fraction at final failure, and usually equals 0.15.

The increase in void volume fraction, f, during increment of deformation, is partly due to the growth of existing voids and partly due to the nucleation of new voids. Thus, the evolution law for the void volume fraction is given in the form:

$$\dot{f}=\dot{f}_{\text{nucleation}}+\dot{f}_{\text{growth}} \quad (4)$$

Nucleation is considered to depend exclusively on the effective strain in the material and can be estimated by following equation:

$$\dot{f}_{\text{nucleation}}=A\dot{\varepsilon}_{eq}^p \quad (5)$$

where $\dot{\varepsilon}_{eq}^p$ is the equivalent plastic strain rate; parameter A is a scalar constant concerning the damage acceleration. It is estimated by the following expression, /15/:

$$A=\frac{f_N}{S_N\sqrt{2\pi}}\exp\left[-\frac{1}{2}\left(\frac{\bar{\varepsilon}_{eq}-\varepsilon_N}{S_N}\right)^2\right] \quad (6)$$

f_N is the void volume fraction of nucleating particles, ε_N is the mean strain for void nucleation and S_N is the corresponding standard deviation. The nucleation parameters, $\varepsilon_N = 0.3$ and $S_N = 0.1$, determined by Chu and Needleman, /15/, are considered reasonable values for the current model.

The void volume fraction due to growth can be estimated by:

$$\dot{f}_{\text{growth}}=(1-f)\dot{\varepsilon}_{ii}^p \quad (7)$$

where $\dot{\varepsilon}_{ii}^p$ are diagonal components of the plastic part of the strain rate tensor.

The critical void volume fraction, f_c, is not considered as a material constant in the complete Gurson model. It is determined by Thomason's plastic limit-load criterion that predicts the onset of coalescence when the following condition is satisfied, /10/:

$$\frac{\sigma_1}{\bar{\sigma}}\ge\left(\alpha\left(\frac{1}{r}-1\right)^2+\frac{\beta}{\sqrt{r}}\right)\left(1-\pi r^2\right) \quad (8)$$

where $\alpha = 0.1$ and $\beta = 1.2$ are two constants fitted by Thomason. Zhang et al. /11/ proposed a linear dependence of α on hardening exponent n for elastic-plastic materials that exhibit strain hardening. σ_1 is the maximum principal stress and r is the void space ratio given by the formula:

$$r=\sqrt[3]{\frac{3f}{4\pi}e^{\varepsilon_1+\varepsilon_2+\varepsilon_3}}\left(\frac{\sqrt{e^{\varepsilon_2+\varepsilon_3}}}{2}\right)^{-1} \quad (9)$$

where ε_1, ε_2 and ε_3 are the principal strains.

MATERIALS AND EXPERIMENTAL PROCEDURE

The material studied in this investigation is high strength low alloyed steel, NIOMOL 490K, used as base metal. Shielded metal arc welding process (SMAW) is used with consumable VAC 60Ni to weld a plate (300×300×16 mm). A mixture of shielding gases: 3.8% CO_2 + 93.7% Ar + 2.5% O_2, is used in order to get acicular ferrite, which raises toughness of the welded joint. Specimens are cut transversely from the welded plate. Chemical compositions of the base metal and consumable are listed in Table 1.

Table 1. Chemical composition of base metal, NIOMOL 490K, and consumable in weight %.
Tabela 1. Hemijski sastav osnovnog materijala, NIOMOL 490K, i dodatnog materijala u tež. %

Material	C	Si	Mn	P	S	Mo	Cr
NIOMOL 490K	0.123	0.33	0.56	0.003	0.002	0.34	0.57
VAC 60 Ni	0.096	0.58	1.24	0.013	0.160	0.02	0.07

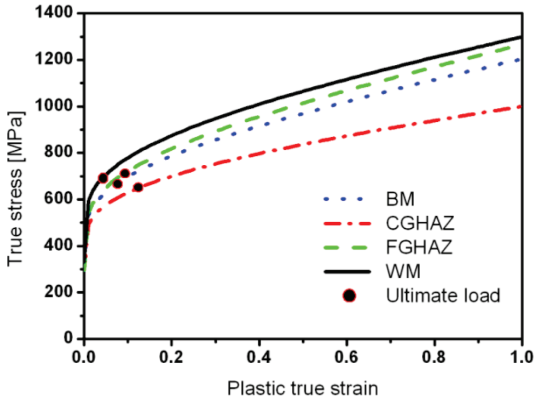


Figure 1. True stress–true plastic strain curves.
Slika 1. Stvarni napon–stvarna plastična deformacija

Table 2. Mechanical properties of used materials.
Tabela 2. Mehaničke osobine korišćenog materijala

Material	Young modulus, E (GPa)	Yield strength, σ_Y (MPa)
BM	202.9	520
CGHAZ	203	550
FGHAZ	195	500
WM	200	530

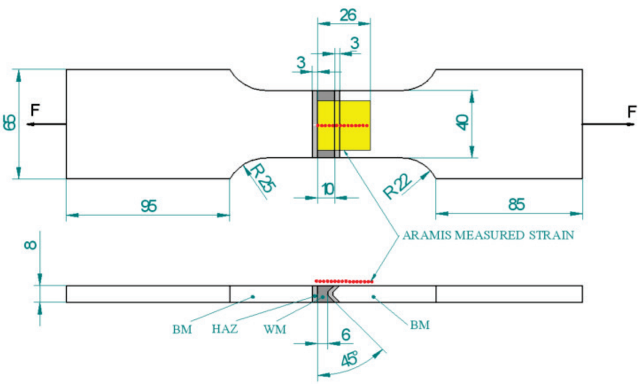


Figure 2. Geometry of smooth tensile specimen.
Slika 2. Geometrija glatke zatezne epruvete

Precise estimation of true stress–true strain curves for different welded joint regions is difficult due to the heterogeneity, metallurgical and mechanical properties of joint zones, especially for heat affected subzones. Therefore, they are obtained by testing a smooth tensile plate at room temperature (Fig. 1). Strains are monitored by ARAMIS measuring system for a specific area on the specimen, which includes welded joint regions (Fig. 2). The smooth tensile panel is numerically modelled to estimate true stress–true strain curves for all regions of the welded joint: base metal (BM); coarse grain heat-affected zone (CGHAZ); fine grain heat-affected zone (FGHAZ) and weld metal (WM). Numerical strains are compared with experimentally measured ones for different regions of the welded joints. The iteration procedure is performed by varying yield strength and hardening exponent up to obtaining a good combination, matching numerical strains with experimental ones (Fig. 3). Figure 3 shows the heterogeneity effect on strains in all regions of the welded joint. It is obvious that the welded joint exhibits slight strength overmatching. The average strain in HAZ changes rapidly in comparison with BM and WM, which can be explained as the effect of heterogeneity. The first iteration of true stress–true strain curves is chosen as the first estimated curves from experimental results. Large values of true stress–true strain curves beyond ultimate load are extrapolated by fitting the data up to the ultimate stress with calculated true fracture stress and strain. Estimated mechanical properties are given in Table 2 for BM, CGHAZ, FGHAZ and WM. The Poisson's ratio is assumed as $\nu = 0.3$.

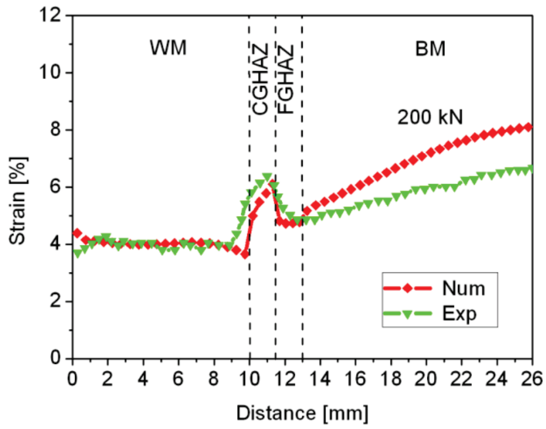


Figure. 3. Comparison between numerical and experimental strain distribution at ultimate load of 200 kN.
Slika 3. Poređenje numeričkih i eksperimentalnih rezultata raspodele deformacije pri maksimalnom opterećenju 200 kN

Quantitative microstructural analysis is performed to estimate the micromechanical parameters: volume fraction (f_v) and mean free paths (λ) between the non-metallic inclusions for the zones in the welded joint, according to /16/, (Table 3). In the initial stage of ductile fracture of steel, the voids nucleate mostly around non-metallic inclusions. Hence, the initial porosity f_0 is here assumed to be equal to the volume fraction of non-metallic inclusions (f_v).

Figure 4 shows the measurement field in the weld metal. A group of oxides and sulphides can be seen in the measurement field, which are marked by arrows.

Table 3. Microstructural parameters of materials.
Tabela 3. Parametri mikrostrukture materijala

Material	f_v	f_N	λ (μm)
BM (NIOMOL 490K)	0.0094	0.014748	578
HAZ	0.0086	0.014748	497
WM	0.0194	0.010685	202

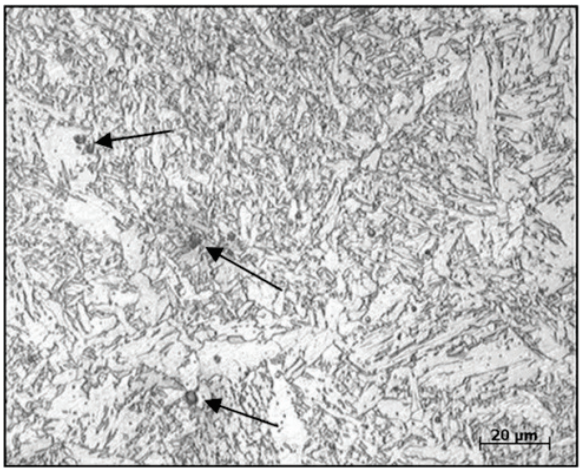


Figure 4. Optical micrograph of non-metallic inclusions in WM.
Slika 4. Optički mikrosnimak nemetalnih uključaka u WM

The volume fraction of sulphides and oxides in the tested steel is determined on the basis of equality with the surface fraction, /16/:

$$V_v = A_A = \frac{A_i}{A_T} \tag{10}$$

where: V_V and A_A are the volume and surface fraction of detected sulphides and oxides respectively, A_i is the area of the detected inclusions and A_T is the measurement field area.

f_v is determined as the mean value of surface fraction of non-metallic inclusions for all measurement fields:

$$f_v = \overline{V}_V = \frac{\sum V_{Vi}}{n} \tag{11}$$

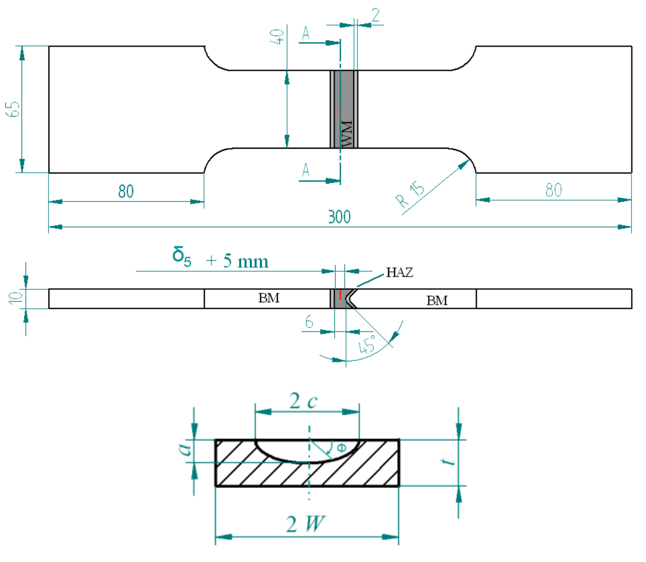
where n is a number of measurement fields.

The volume fraction of void nucleating particles (f_N) (Table 3), which represents the effect of secondary voids on ductile fracture, is calculated from the content of carbon in the tested materials using the lever rule, /17/:

$$\text{weight\% cem.} = \frac{\%\text{C} - 0.025}{6.67 - 0.025} \cdot 100\% \tag{12}$$

where %C is carbon content in the used steel, 6.67 is the weight % of carbon in Fe_3C compound and 0.025 is the weight % of carbon content in ferrite.

The tensile specimen is used to investigate the fracture behaviour of welded joints with pre-cracks in WM. Figure 5 shows the geometry of the tensile specimen. The initial crack length to width ratio is $a_0/W = 0.5$. The fatigue pre-crack is located in the centre of the weld metal for specimen with a pre-crack in the WM. Crack mouth opening displacement (CMOD) and the applied force (F) are experimentally monitored for the specimen (Fig. 6) and compared by numerical and von Mises results.



Section A-A

Figure 5. Geometry of tensile specimen with a pre-crack in WM.
Slika 5. Geometrija zatezne epruvete sa prslinom u WM

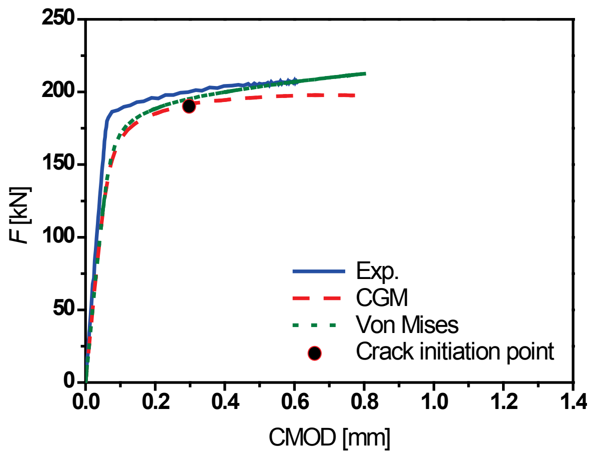


Figure 6. Crack mouth opening displacement (CMOD) vs. force (F) for tensile panel with surface crack in WM.
Slika 6. Pomeranje pri otvaranju usta prsline (CMOD) sa silom (F) za zateznu ploču sa površinskom prslinom u WM

FINITE ELEMENT MODELS

For the determination of the values of stress and strain components and the value of damage parameter (f) in the specimens exposed to external mechanical loading, the FEM programme Abaqus (www.simulia.com) is used, with CGM user subroutine, UMAT, developed by Zhang, based on /11/. To simplify the finite element analysis, materials of all regions of welded joint are assumed to be isotropic. The mesh size, l_c, is chosen to approximate the mean free path between non-metallic inclusions. A fixed mesh size l_c = 0.2 mm of elements is chosen on vertical planes on the crack front of the tensile specimen with semi-elliptical surface crack in WM, but along the crack front this value is about $5l_c$ because the variation of stress/strain in this direction is not significant (Fig. 7).

This size approximates the value of the mean free path (λ) between non-metallic inclusions in tested materials (see Table 3).

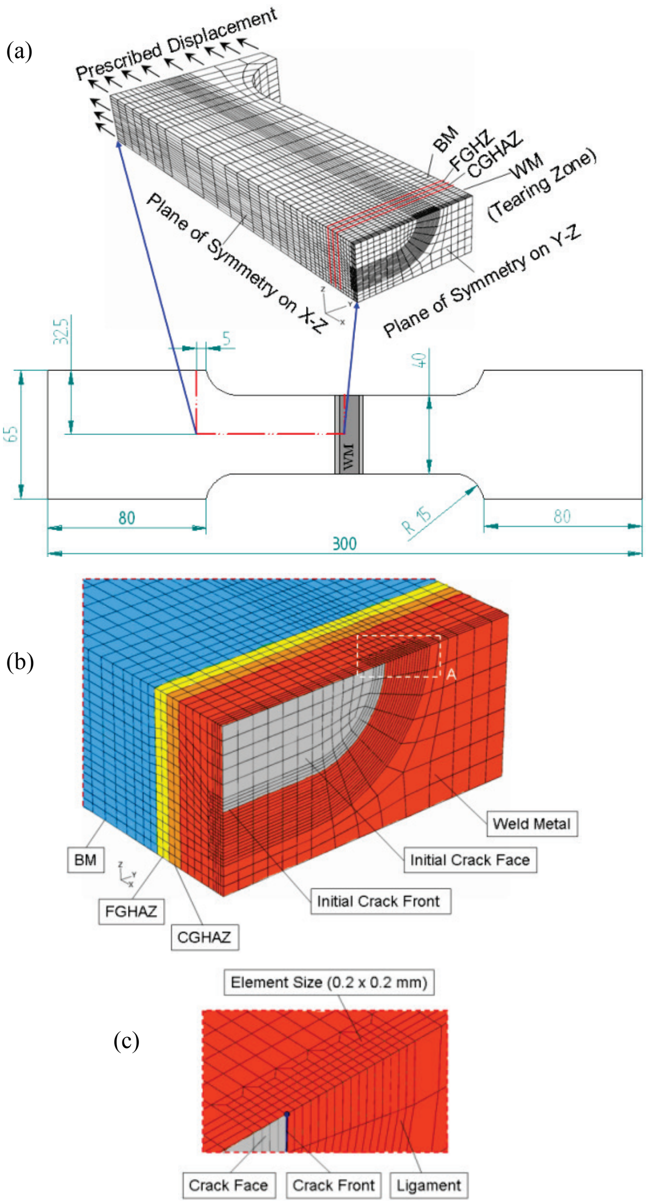


Figure 7. 3D FEM for tensile panel with surface crack in WM: (a) 3D element mesh for one quarter of specimen with boundary conditions; (b) detailed mesh for region near the crack front; and (c) detail A for the mesh near the crack front.

Slika 7. 3D MKE zatezne ploče sa površinskom prslinom u WM: (a) 3D mreža za četvrtinu epruvete sa graničnim uslovima; (b) detalj mreže za oblast blizu fronta prsline; i (c) detalj A mreže blizu fronta prsline

It is well known that ductile tearing of metals occurs by nucleation, growth and coalescence of microvoids with significant plastic deformation. Therefore, the zone of interest with the crack is modelled by CGM and the rest of the model obeys to elastic-plastic behaviour without damage, /18/. Based on this, ductile fracture is modelled in the current models by introducing a tearing zone surrounding the crack line, where material degradation and separation can occur. This tearing zone is embedded in a continuous elastic-plastic material where only plastic deformation occurs. Thus, the tearing zone is considered in the WM.

The incremental plasticity model provided by Abaqus is used for zones of the joint which do not contain the crack. The loading of the specimen is controlled by prescribed displacements.

In order to apply the CGM model to simulate the ductile tearing in the tensile specimen with a pre-crack in WM, various model parameters must be determined.

The first set of constitutive parameters are q_1 and q_2, which relate to the hardening of the matrix material. In this study, q_1 and q_2 are determined to be 1.6 and 1.0, in respect, for the tensile specimen with a pre-crack in WM. Values of q_1 and q_2 are considered according to the study in /19/. The second set of parameters are void initiation and coalescence parameters (f_0, f_c and f_F). As mentioned previously, the initial void volume fractions (f_0) are assumed to be equal to the volume fraction of non-metallic inclusions (f_v), given in Table 3 for BM, HAZ and WM material. The critical void volume fraction (f_c) is a crucial damage parameter in CGM, since it represents the end of stable void growth and the start of void coalescence. It is not a material constant according to CGM, but it is automatically determined during the numerical processing, based on stress and strain fields. Void volume fraction at final fracture (f_F) is determined according to the relation $f_F = 0.15 + f_0$, /11/, used in the complete Gurson model in the present study. The third set of parameters (ε_N, S_N, and f_N) is related to void nucleation. The volume fraction of void nucleating particles (f_N) has been evaluated from Fe_3C content in materials. Nucleation parameters, $\varepsilon_N = 0.3$ and $S_N = 0.1$, determined by Chu and Needleman /15, 20, 21/, are considered for the analysis model.

Numerical modelling of crack initiation

Crack initiation can be predicted by using the CGM model according to the failure criterion. Failure is defined by the instant when the first element in front of the crack tip becomes damaged. The condition for the onset of the crack growth (as determined by the J-integral, J_i, or crack tip opening displacement, $CTOD_i$) is most adequately defined by the micromechanical criterion, /22/:

$$f \geq f_c \tag{13}$$

when the condition given by Eq.(13) is satisfied, the onset of crack growth occurs. The critical void volume fraction (f_c) in the CGM model is determined by evaluating Eq.(8) at the end of every increment step. Once Eq.(8) is satisfied, void coalescence starts and the current void volume fraction is regarded as f_c for that integration Gauss point. To determine numerically the crack initiation, the increase of void volume fraction (f) should be monitored at the nearest Gauss point to the crack tip. When the current monitored f reaches f_c and Eq.(13) is satisfied, the fracture mechanics parameter at crack initiation (J_i or $CTOD_i$) is determined.

The ductile growth initiation and propagation for tensile panels have been modelled as follows.

Ductile crack growth initiation described here by the J-integral at initiation (J_i) is modelled for the tensile panel with a surface crack in WM based on critical void volume fraction criterion (f_c) which represents the end of stable void growth and the start of void coalescence in the material.

The value of J_i has been estimated numerically at the centre of specimen thickness in front of the crack line, where the highest value of void volume fraction occurs. The value of J_i for tensile panel with surface crack in WM (TP-WM) is given in Table 4 in comparison with values of J_i for SENB specimen with pre-crack in WM (SENB-WM). The J_i of TP-WM is experimentally determined using the stretch zone width (SZW).

Table 4 Numerical values of J_i for tensile and SENB specimens.
Tabela 4. Numeričke vrednosti J_i zatezne i SENB epruvete

Specimen designation	$J_{0.2/BL}$ (N/mm)	J_i (N/mm)	
		Using SZW	CGM
SENB-WM	64.7	-	57.6
TP-WM	-	120.3	114.7

Numerical modelling of ductile crack propagation

Crack growth in ductile materials is conventionally characterized by fracture resistance curves, obtained from standard fracture tests. However, these standard fracture tests introduce a high degree of conservatism in engineering critical assessment of real structures such as pressure vessels. Therefore, using specimens such as cracked tensile panels may present better integrity assessment.

The J-R curve for WM has been numerically obtained using the tensile panel with a surface crack in WM. It has been simulated by tracing the path of completely damaged elements, which appear completely in different colours in this work (Fig. 8). The element is assumed to be failed (completely lost its load carrying capacity) when the void volume fraction at final failure (f_F) is reached according to the relation $f_F = 0.15 + f_0$. Then, the corresponding value of J-integral is obtained. The crack growth resistance curve is presented in Fig. 9 for tensile panel with pre-crack in WM at the deepest point of the crack front for tensile panel (Φ = 90º, Figs. 5 and 8), where the largest crack growth occurs. The result has been compared with the J-R curve for SENB specimen with a pre-crack in WM, obtained previously in /23/. It is obvious that the J-R curve obtained using SENB specimen is more conservative than one obtained by using the tensile panel.

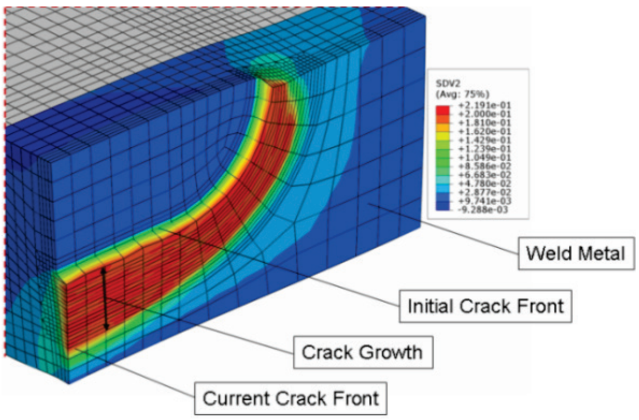


Figure 8. Distribution of void volume fraction (red damaged elements represent crack growth path) for tensile panel with surface crack in WM.
Slika 8. Raspodela zapreminskog udela šupljina (crveni oštećeni elementi su putanja površinske prsline) za zateznu ploču sa površinskom prslinom u WM

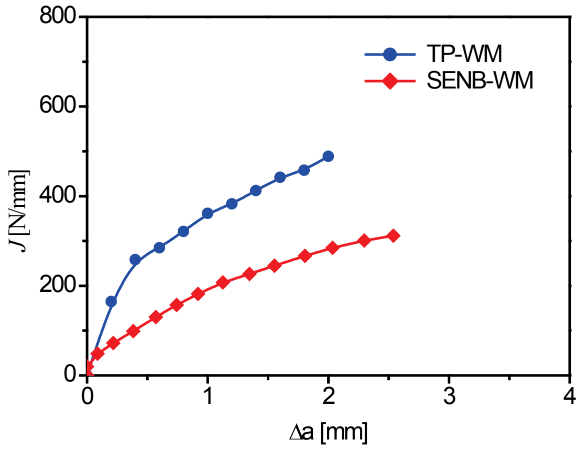


Figure 9. J-R curve for SENB and tensile specimen with a pre-crack in WM.
Slika 9. J-R kriva za zateznu i SENB epruvetu sa prslinom u WM

CONCLUSIONS

The micromechanical complete Gurson model (CGM) is applied to estimate damage level (void volume fraction, f) in the welded tensile specimen with a pre-crack in WM. True stress–true strain curves of welded joint zones are determined by a combined experimental-numerical procedure, utilizing stereometric strain measurements and finite element modelling.

The crack initiation value is successfully predicted using the CGM and true stress-true strain curves which are estimated by the experimental-numerical procedure. Results show that the constraint effect due to the geometry can be predicted as well.

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100th Obituary of August Wöhler

The ESIS Wöhler Medal is given at each ECF conference to distinguished experts standing in the tradition of August Wöhler.

In 2014 the technical community in Germany commemorates the 100th obituary of 'The Pioneer of Materials Testing' August Wöhler who lived nearly 95 years. Wöhler's famous experimental works date back to the middle of the 19th century. His fatigue tests running over a time of fourteen years had led to five publications between 1858 and 1870.

Wöhler is internationally regarded as the first who studied parameters of dynamic strength methodically. Therefore he is called the *'Father of Fatigue Strength'*.

Especially in England his tests caused sensation as the fatigue of components was already a big topic there due to many losses of railway wheels and axles in the century of industrialization.

At the time only fracture strength was known. All smaller loads did not have any significance. It turned out that repeated stresses below the tensile strength might cause fracture. There was no explanation for this phenomenon. And as fatigue cracks cause almost no deformation it was thought that there has to be a material crystallization process caused by repeated stress.

Five publications of Wöhler question component strength, engineering mechanics, with the design of testing machines and the execution of tests. Wöhler had to invent his testing machines by himself and thereby set a precedent. Today his testing machines are part of the permanent exhibition of the *"Deutsches Museum Munich"* among masterpieces of science and technology: http://www.deutsches-museum.de/en/exhibitions/materials-production/materials-testing/

Wöhler tested original railway axles and test bars made of iron, steel and copper. The loading was rotational bending, bending in one direction, torsion and axial loading. In axial loading the mean load was alterable.

In his publication of 1870 Wöhler very shortly sums up his experimental results self-confidently: *'The law I found, whose general validity for iron and steel shall be proved by these experimental results, is called: The fracture of materials can be caused by multiple repeated vibrations of which none itself causes absolute rupture stress. The stress differences, which are containing the vibration, are decisive for the destruction of the structure… The absolute magnitude of the dynamic stress limit is only insofar of influence as with growing stress the differences minimize which are causing the fracture.'*

Wöhler considered not only the endurance limit but also structures for a limited number of cycles during the lifetime. He has already stated that in case of bending of a shouldered shaft, the shape of the transition is relevant for the durability.

What did Wöhler say about fatigue of materials? Did he examine the fracture surface and did he compare the fractures of his single step tests with the fractures under service operation? Can a fracture surface be regarded without realizing that there is a crack which spreads and might also arrest? His experimental studies meant a lot to him. But also very important for him was the reliable strength dimensioning by an exact determination of the permissible stress, a classification of materials (iron and steel) regarding rather characteristic properties than application and the constitution of independent testing institutions, the institutes for materials research and testing.

Readers who are interested in these topics might be pleased as DVM German Association for Materials Research and Testing will publish a historical special issue on the occasion of August Wöhler´s 100th obituary in 2014. The special issue will illuminate the development of the research on fatigue strength in the 19th century, the age of industrialization. The volume is for now written in German but contains short abstracts in English for each chapter, an excellent History of Fatigue Strength in English, an extended bibliography up to publications in 2012 and many reprints of English journal articles of the 19th century such as from the journal *'Engineering'* 1867 and 1871. All enhanced by excellent reprints of the drafts of Wöhler´s unique testing machines.

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(from ESIS Newsletter #54-2014)