NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS

Book of Abstracts

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Cover image: Fracture of AISI 4340 high strength low alloy steel after Charpy test - Combination of transgranular ductile fracture and intergranular decohesion, by Doc. Ing. Jan Siegl, CSc., image magnification of 3000x.

Proceedings were designed by
Ing. Pavel Žídklik
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THE SEVENTH INTERNATIONAL WORKSHOP

NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS 2016

http://konference.fmmi.vsb.cz/work2016/

Organized by
VŠB-Technical University of Ostrava, Faculty of Metallurgy and Materials Engineering
Yokohama National University, Faculty of Engineering, Institute of Advanced Sciences

Under the auspices of
Ministry of Foreign Affairs of Japan
Embassy of the Czech Republic in Tokyo
Yokohama Convention & Visitors Bureau
Welcome Message

Dear Friends and Colleagues,

With great pleasure the Organizing Committee is sending the final circular of The Seventh International Workshop, New Methods of Damage and Failure Analysis of Structural Parts 2016. The workshop is held at the Yokohama National University on 1st–4th November, 2016.

The series of the workshop started in 2004 at VŠB-Technical University of Ostrava and ever since, the workshop has succeeded in discussing interdisciplinary approaches that integrate the knowledge of degradation processes in materials, reliability of engineering parts in service, and transfer of materials research results to industry. This workshop will continue in the tradition of earlier meetings. We believe Yokohama provides an exciting and relaxing place for discussing the latest development, sharing experiences, and planning future collaborations and projects.

The final circular provides updated information on the program. The technical program of the workshop will feature the latest research and state-of-the-art developments in all areas related to damage and failure analyses of structural parts. With best enthusiasms and hospitality, we look forward to seeing you in Yokohama.

Sincerely,

Conference chairman

Professor Bohumír Strnad
VŠB-Technical University of Ostrava

Conference co-chairman

Professor Osamu Umezawa
Yokohama National University
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Contact about registration fee
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C.A.T. (Creative & Academic Tomorrow)
**Workshop Information**

All relevant information about the workshop is introduced on the following website:

http://konference.fmmi.vsb.cz/work2016/

**Venue**

Yokohama National University, Education and Culture Hall (Building number: S1-2)

http://www.ynu.ac.jp/english/access/index.html
Program at a glance

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Round Table Discussion & Tour

The full day tour at Kamakura will include visits to Kamakura main sightseeing spots by walking. It also gives you a chance to experience Japan’s traditional culture through the visits exploring Zen temples established in the 12 century. Enrollment can be done at the registration desk. The tour will start from Yokohama Kokusai Hotel at 8:00 a.m. on November 4. We will move to Kitakamakura station by train and then divide into each courses.

- **Course 1: Kamakura tour (Visit ancient temples), Finish time: approx. 15:00.**
  You will visit traditional temples and shrines in Kamakura: Engakuji, Kenchoji, Tsurugaoka Hachimangu, Hasedera, Koutokuin. You can enjoy Japanese traditional temples, shrines and foods in this course.
- **Course 2: Kamakura and Enoshima tour (Visit ancient temples and island with great sea view), Finish time: approx. 18:00.**
  In addition to Course 1, you will visit Enoshima island. You can see good sea and sunset view in Enoshima island, and you will be impressed by these great views.
- **Course 3: Kamakura hiking tour (Walk Kiridoshi pass), Finish time: approx. 16:00.**
  You will walk the Kiridoshi pass which is man-made passageways for defensive purpose and beautiful green tunnel. After walking the Kiridoushi pass, you will visit temples and shrines: Kotokuin, Hasedera, Tsurugaoka Hachimangu, Kenchoji, Engakuji.
- **Course 4: Short Kamakura tour (Visit ancient temples), Finish time: approx. 12:30**
  We visit only highlight points of Kamakura: Kenchoji, Tsurugaoka Hachimangu, Kotokuin. The finish time is approx. 12:30, if you want to visit some other place on the day, please choose this course.

Publication

Contributions will be collected and published with a peer review process in the proceedings by **Key Engineering Materials**, Trans Tech Publications Inc.
APPLICATION OF SMALL PUNCH TEST FOR EVALUATION OF MECHANICAL PROPERTIES OF NEWLY DEVELOPED ODS STEELS

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KEY WORDS: small punch test, ODS, 9Cr1W

In testing areas such as inspection of long-term operation components of power plants or characterization of newly developed material, volume of testing material is usually limited. Therefore it is not possible to use conventional methods to determine mechanical properties. Small punch test (SPT) is one of techniques using miniaturised samples, specifically thin discs of thickness up to 0.5 mm and diameter up to 10 mm. This type of samples is also useful for characterization of the gradient of properties (e.g. heat affected zone of welds).

Principle of SPT is a penetration of hemispherical punch through the disc into a hole until the disc bursts (see Fig. 1). In dependence on load conditions we can obtain data that are analogous to three basic conventional tests: i) SPT-CDR – analogy of uniaxial tensile test (UTT) - punch penetrates the specimen with Constant Deflection Rate and the force-deflection dependency is recorded, ii) SPT-CF - analogy of conventional creep test - specimen is loaded with the Constant Force and deflection is recorded in time, iii) SPT-CD - analogy of relaxation test - the specimen is loaded to chosen Constant Deflection and the force is recorded in time. Whereas the method has been proposed as a comparative method, its certain weakness is complicated clarification of the relations between SPT and conventional tests. The aim of this study is to i) evaluate mechanical properties of steel with different variants of oxide dispersion-strengthening, ii) correlate SPT results with UTT and validate correlation formulas from the literature, iii) propose more advanced methodology of data evaluation.

Five variants of 9Cr1W ferritic steel (Eurofer type) were used as an experimental material. Samples were spark-cut from discs (ø30 mm), which were prepared by Spark Plasma Sintering (SPS) of mechanically alloyed powder (1150°C/50MPa/5’). Oxide dispersion was achieved using two approaches: i) admixing of commercial oxide powder (Al2O3 resp. Y2O3) into the alloy made of pure-metal powders (hereinafter referred to as Al2O3 resp. Y2O3), ii) using pure aluminium resp. yttrium powder and iron partly alloyed by oxygen as initial components (hereinafter referred to as Al + O2 resp. Y + O2).

Fig. 1. SPT setup.

Fig. 2. Comparison of steels with two types of SPT sample fracture.
Both SPT and UTT were carried out at room temperature with constant deformation rate. Force and deformation characteristics were determined from each $F$ vs. $u$ diagram (Fig. 2). Force $F_m$ is maximum force in $F$ vs. $u$ diagram; $u_m$ is value of deflection corresponding to maximum force. There are several approaches to determine force $F_y$, which is useful for conversion to $R_y$. In this work, so-called Offset method was used to determine $F_y/h_{0/10}$. There is an analogy with determining of proof strength, with the difference that offset is not defined by percentage of deformation (e.g. 0.2%) but using sample thickness (specifically $h_{0/10}$).

**DISCUSSION AND CONCLUSIONS**

Correlation of UTT strength $R_m$ and SPT force $F_m$ in form of normalization factor $F_m/(u_m h_{0})$ can be seen in Fig. 3. This factor reflects ductility of the material (by using $u_m$), which proved to be appropriate for this type of steels. Data were compared to formulas from the literature [1-3]. Good agreement of data points with equation by García and Rodríguez [1] is apparent. $Y_2O_3$ variant is the only one with significant deviation. However, there is relation between position of the point towards regressions and fracture behaviour of the material, which is reflected by the type of fracture. Points above regression have predominantly brittle “star” fracture type, whereas points bellow regression have predominantly ductile “cap” type of fracture (Fig. 2).

Dependency of $R_{p0.2}$ on normalization factor $F_y/h_{0^2}$ (Fig. 4) has also good agreement with equation by [1]. Whereas another factor $F_m/h_{0^2}$ (not in Fig.), which only counts with initial thickness of the specimen, matches equation by [1] only in case of ductile variants (oxide free, $Al_2O_3$).

We can conclude that regressions by [1] are suitable for estimating UTT properties of different variants of 9Cr1W ODS steel. There is no need to modify the regression equation for this type of material in case of $F_m/(u_m h_{0})$ and $F_y/h_{0^2}$ normalization factor. Further work will be focused on improvement of estimation methodology using microscopic analysis of fracture area.

**Acknowledgements:** This work was financially supported by Czech Science Foundation project No. 14-25246S and No. 15-21292Y.

**REFERENCES**


ELASTIC MODULI, ISOTROPY, AND DEFORMATION ACCOMMODATION MECHANISMS OF COLD SPRAYED DEPOSITS

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KEY WORDS: CGDS, kinetic spray, elastic properties, resonant ultrasound spectroscopy

Cold spraying [1, 2] is a versatile and efficient method for deposition of metals and alloys that readily undergo chemical or structural changes at elevated temperatures. Unlike the conventional thermal spraying methods, the CS process does not involve substantial heating or melting of the sprayed powders. Instead, formation of coating arises from a severe plastic deformation of the accelerated powder particles upon their impingement at the substrate. Thereby, the oxidation or phase changes in the feedstock material are effectively reduced [3]. Metallic coatings prepared by CS frequently exhibit superior mechanical and physical properties, such as Young's modulus and hardness, or electrical conductivity [4]. Further to that, the coatings deposited via high-temperature processes generally exhibit high levels of anisotropy [5], a property arising due to their heterogeneous lamellar microstructure. Resonant ultrasound spectroscopy (RUS) has been recently used to investigate elastic anisotropy of coatings prepared by thermal spray methods [6]. In all reported cases, the differences between the in-plane and out-of-plane properties of the coatings were significant (a factor of 0.36 to 0.87).

Four analyzed materials (copper, aluminum, nickel, titanium) were deposited via high-pressure cold spray system (Plasma Giken PCS-1000) onto polycrystalline aluminum substrate with thickness exceeding 10 mm. From each of the sprayed materials, a rectangular shape sample (3.5x2.5x1.5 mm³) was cut from the region close to the substrate-coating interface. RUS method was then applied to determine the elastic constants and the strength of anisotropy (free elastic vibrations spectrum in 0.1 – 2 MHz range).

In the coatings, a strongly heterogeneous (bimodal) distribution of the plastic strains within the individual sprayed particles was observed (Fig. 1): the dimensions of the individual differently plasticized subgrains along the particle boundaries are submicron, while these subgrains are of the order of micrometers in the middle of the particles. In the hcp Ti coating, twinning-like plastic deformations are seen all over the particles.

The Young's moduli of all materials in the out-of-plane (OOP) direction were higher than 70% of the moduli of the corresponding bulks. Furthermore, all coatings exhibited nearly perfect elastic isotropy. The ratios between the OOP and in-plane (IP) Young's moduli range between 0.98 and 1.03 and the differences from unity were fully comparable to the experimental errors. Although a similarly perfect elastic isotropy was reported for amorphous alumina coatings [7], for crystalline metallic coatings such results are very rare. Importantly, the isotropy was reached for all CS materials, although they belong
to different crystallographic classes and differ significantly in a number of mechanical properties. It can therefore be concluded that such unique properties are not particular to the individual sprayed metals, but follow from the CS technology itself.

In Fig. 2, the results obtained for the CS materials are compared with metals sprayed by other thermal spray methods (measured by ultrasonic methods, i.e., at small straining amplitudes and high straining rates). The CS materials obviously possess a unique combination of high $E_{OOP}/E_{bulk}$ ratio and very weak anisotropy.

The $E_{OOP}/E_{bulk}$ ratios correlated with observed differences in the internal friction. The increase of internal friction and the simultaneous decrease of the relative Young’s modulus are probably caused by the differences in the coating microstructures. The Al and Ni coatings contain highly misoriented, equiaxed nano-sized grains [8] probably formed by sub-grain rotation under impacts during the spraying, while no such grains can be found in the copper coatings. It could be safely assumed that these grains contribute to softening and internal friction of the analyzed coatings by grain boundary sliding [9].

Acknowledgement: The work has been financially supported by the Czech Science Foundation projects No. GA13-13616S and GA13-35890S.

REFERENCES

CRACK RESISTANCE CHARACTERISATION TIAL INTERMETALLICS SHOWING ENHANCED TOUGHNESS

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KEY WORDS: TiAl, intermetallics, fracture toughness, micromechanisms of fracture, shear ligament toughening, thermal treatment

INTRODUCTION

Intermetallic TiAl based alloys appears to be a structural material showing high potential for high temperature applications. They experienced rapid development during last decade, in particular in modifications of alloying base incorporating also microalloying issues, refinements of grain and lamellar microstructure, and fabrication technologies development including powder metallurgy procedures [1-4]. This intensive development has been primarily controlled by needs of design applications. Less effort has been paid to the titanium aluminides performance, including their fracture resistance, at room temperature and at increased temperatures. Just this low temperature behaviour appears to be a limiting obstacle for the more extensive applications [5].

There are two groups of mechanisms enabling crack resistance and toughness enhancement of TiAl intermetallics: intrinsic mechanisms and extrinsic ones (for a more detailed overview see [6]). In the first case, the above mentioned suitable alloying and heat treatment are the most effective ways. In the other case, crack tip shielding effects can produce toughness increase, e.g. change of crack trajectory due to crack bowing (leading to increase of fracture surface roughness) and/or crack bridging, in some cases connected with shear deformation. In the last it is possible to find certain potential for the toughening at room and increased temperatures; this is conditioned by corresponding knowledge on the micromechanisms of microcrack nucleation and brittle crack propagation. For almost all microstructures formed by lamellar colonies, it is usually possible to observe several damage locations at certain distances from the crack tip, the initiation occurs by joining the sharp crack tip with the nearest damage site / microcrack. Individual microcracks in front of the crack tip are usually formed (nucleated) at boundaries of two lamellas or on boundary of lamellar colony with equiaxed \( \gamma \)-TiAl grain (supposing they are present in the microstructure) on the lamellar colony boundary. Local influence of deformation twin and/or intensive slip band onto this boundary and microcrack nucleation in neighbouring grain in local tensile stress field [6,7] is usually observed.

Specification of conditions for microcrack nucleation and fracture initiation and explanation of the role of microstructure in fracture micromechanism at the same time is extremely important for further development of titanium aluminides. The aim of the contribution is seen in analysis of fracture micromechanisms in investigated TiAl model alloy in state prepared by mechanical - thermal treatment.

EXPERIMENTAL

The alloy used for investigation was a Ti-43\%Al model alloy. Conditions for phase transformations during heating and subsequent mechanical thermal treatment leading to lamellar colonies and the same with specific „zig zag” orientation have been investigated [8, 9]. The one step mechanical thermal treatment consisted of the sample forging in
compression at 1553 K (in α region) with total strain of $\varepsilon = -1.1$. Then the sample was cooled down to 1473 K (in α+γ region) and kept for 800 s followed by controlled cooling to room temperature in order to enable lamellar microstructure formation. The two step treatment consisted of the same forging regime at 1553 K, but after the lamellar formation at 1473 (in α+γ region) additional compression at $\varepsilon = -0.6$ was incorporated expecting „zig zag“ microstructure formation.

The samples of dimension 5×7×28 mm for fracture toughness determination have been tested in three point bending, the crack was introduced under four point bending keeping exactly the validity conditions of the standard. The fracture toughness was evaluated based on determination of crack resistance (J-Δa) curves by applying load/unload technique.

RESULTS AND CONCLUSIONS

In the study attention was paid to lamellar colony morphology and analyses of the effect of this morphology on crack development. High temperature mechanical treatment under conditions of compressive deformation contributed to formation of lamellar colonies. The “zig zag” type arrangement of lamellar colonies was obtained by the two step treatment, the average lamellar colony size was about 280 µm.

Thanks to compressive deformation and refinement of the lamellae thickness the microstructure was strengthened and this strengthening effect appeared to be stable also at high temperatures. Crack resistance curves showed toughness enhancement comparing to as received states. Not only the overall toughness increased but also the increasing crack resistance curves have been observed. Thus the thermal-mechanical treatment resulted into increase of strength and simultaneously to the fracture toughness increase showing clearly quite positive effect of the applied treatment. This performance of the material has been observed in both tested directions of crack propagation relating to compressive deformation, perpendicular and parallel to compression deformation direction.

Fractographic analysis enabled to evaluate fracture micromechanisms. The fracture surface was found to be very rough. Presence of islands not broken beneath the fatigue crack tip shows that the crack propagated by linking up the microcracks in front of the crack tip. Both these observations have been taken as evidence of shear ligament mechanism development and this was the key mechanism controlling the fracture behaviour of the microstructures prepared.

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KEY WORDS: fracture toughness, crack resistance curve, transferability, crack tip constraint, subsized specimen

Transferability of fracture mechanics characteristics. The fracture mechanics parameters characterise the stress singularity in front of the crack tip. In elastic material $K$ parameter and in elastic-plastic material $J$ parameter were derived to scale the stresses at the crack tip. The plastic zone and the stresses distribution at the crack tip are function of loading, boundary conditions and material’s properties. If the plastic zone at the crack tip is sufficiently small to be within the singularity dominated zone then the fracture mechanics parameter characterises the crack tip conditions. These conditions are referred as SSY (small scale yielding) conditions. Under SSY conditions a single parameter (e.g. $K$, $J$ or $CTOD$) can be applied as a geometry-independent fracture criterion. The transferability of fracture mechanics parameter from laboratory specimen to real structure/component is possible upon keeping the same stress singularity conditions at the crack tip in both geometries. When the structure and specimen are loaded to the same value of fracture mechanics parameter, the crack tip conditions are identical in both configurations and have the same critical value of fracture toughness parameter in a case of small plastic zone.

Quantification of the crack tip constraint and the constraint loss corrections. The presence of SSY conditions at the crack tip means high levels of the stresses. Any surpassing of the SSY conditions causes decrease of the level of the stresses in the vicinity of the crack tip, Fig. 1. This phenomenon is known as the loss of constraint. The single-parameter fracture mechanics then breaks down and fracture toughness depends on the size and geometry of the test specimens, data generated from a sample are in constraint dependent regime.

Two-parameter fracture mechanics has been developed in order to address/quantify the constraint loss phenomena by introduction of the second parameter describing the conditions at the crack tip. The principles of two-parameter fracture mechanics as well as the procedures characterising the different level of constraint, e.g. the boundary layer method (BLM), $T$-stress,
$Q$-parameter have been applied. The definition used for specification of the constraint parameter ($Q$-parameter) can be also used for correction of $K_c$ data in constraint dependent regime. The main principle of the toughness scaling diagrams [3-4] arises from diagram $J_0/b$ versus $J_{FE}/b\sigma_0$, where $b$ is unbroken specimen ligament and $\sigma_0$ is yield (flow) stress. Then the loading parameter is transferred from the tested (real/small) geometry to SSY state corresponding to full thickness specimen/component ($Q = 0$, standard specimen geometry).

**Small specimen test technique (SSTT).** The evaluation of mechanical properties from limited amount of test materials is one of important issues connected with development of new materials and its local characterization. Basic mechanical characteristics like strength and deformation characteristics, hardness can be nowadays relatively easily determined by various approaches, e.g. miniature tensile test, small punch test, micro-hardness test etc. However fracture toughness as an advanced material characteristic describing the resistance against crack initiation and propagation involves the introduction of a crack into the specimen. Moreover the valid determination of fracture toughness parameters ($K, J, C_TOD$) requires fulfilment of small scale yielding (SSY) conditions at the crack tip which depend on geometry and size of a specimen and on deformation behaviour of material. These two necessary requirements make the fracture toughness determination not an easy task in case of limited amount of test material. There are two basic ways of fracture toughness determination directly from limited amount of test material: reconstitution and application of miniature test specimen. The first is connected with production of specimen of sufficient size and needs suitable technological process of welding. The second scales the geometry of usually used fracture toughness specimens in order to be makeable from available amount of material, Fig. 2. An optimised methodology of the test results interpretation and correction of the fracture toughness values influenced by the loss of constraint are needed. This issue is solved within small specimen test technique (SSTT) which has been developed since 90’s together with testing and development of materials for nuclear applications (both fission and fusion).

Within research programme the SSTT has been solved for two miniature specimen types. The 3PB specimen was applied to measure fracture toughness of the JRQ steel (reference material for IAEA round robin testing) in transit region and of the Eurofer97 steel (proposed structural material for DEMO fusion reactor) in the ductile region. Compact tension specimen was used for fracture toughness evaluation of ODS steel MA956 in ductile region at elevated and high temperatures. The test results will be discussed from the point of view of validity of fracture toughness characteristics. Then the quantification of the constraint at the crack tip using FE via stress and strain distribution, T-stress and $Q$-parameter and will be performed. Correction of fracture toughness values corresponding to the SSY conditions will be given.

**Acknowledgement:** The works on paper have been partly financially supported by project of Czech Science Foundation Nr. 15-21292Y.

**REFERENCES**


FATIGUE LIMIT IMPROVEMENT BY PEENING FOR WELDED JOINT CONTAINING A CRACK-LIKE DEFECT – EVALUATION FOR THE DEFECT SIZE RENDERED HARMLESS –

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KEY WORDS: welded joint, peening, fatigue limit, compressive residual stress, fracture mechanics

INTRODUCTION

Welded joints are often used in large steel structures. Fatigue cracks often initiate at the welded joint, ultimately leading to structural fracture. Therefore, non-destructive inspections are typically conducted at regular intervals. However, small cracks below a detection limit cannot be detected. If these fatigue cracks could be rendered harmless by peening, the structural integrity could be significantly improved. From an engineering perspective, it is very important to estimate the maximum depth of the semi-circular slit that can be rendered harmless by peening in a weld-toe zone. In the present study, it was evaluated based on fracture mechanics.

TEST MATERIALS AND TEST METHOD

The material used is austenitic stainless steel (JIS-SUS316). The shape and dimension of the butt-welded specimen are shown in Fig. 1. A semi-circular slit similar to a surface crack was introduced within 0.2 mm of the weld toe. Portable pneumatic needle-peening (PPP) was performed at the weld toe. The details of the peening conditions are given in reference [1]. Plane bending fatigue tests were performed with constant load amplitudes and with a stress ratio of \( R = 0 \). All the tests were carried out with a frequency of \( f = 20 \) Hz at room temperature in air. The fatigue limit was defined as the maximum stress amplitude at which the specimen could endure \( 5 \times 10^6 \) cycles of stress.

FATIGUE TEST RESULTS

The fatigue limits of all the specimens were increased 60% - 133% by peening as shown in Fig. 2. If the fatigue test results of a peened specimen with a semi-circular slit meet either of the following two conditions, the slit is considered to have been rendered harmless.

Condition (a): The fatigue limit increased up to that of the peened specimen without a semi-circular slit.

* Fractured at the different location than the slit

Fig. 2. Relationship between stress amplitude and depth of the semi-circular slit [1].
Condition (b): The specimen fractured outside the slit.

From Fig. 2, a semi-circular slit under \( a = 1.0 \) mm could be rendered harmless by peening because of meeting condition (a).

**DISTRIBUTION OF RESIDUAL STRESS**

The longitudinal residual stress distributions at the weld toe along the thickness direction are shown in Fig. 3. The surface and the maximum compressive residual stress after peening were about 350 MPa and 500 MPa, respectively.

**STRESS CONCENTRATION OF THE WELD TOE**

Finite element method (FEM) analysis was conducted to calculate the stress concentration factor \( K_t \) of the weld toe. The \( K_t \) before and after peening was 2.5 and 1.7, respectively. Therefore, we concluded that stress concentration of the weld toe was relaxed by peening.

**EVALUATION OF THE SEMI-CIRCULAR SLIT SIZE RENDERED HARMLESS BY PEENING BASED ON FRACTURE MECHANICS**

In this study, we assumed that the positive value of the stress intensity factor \( \Delta K_T \) which can be calculated by equation (1) contributes to fatigue crack propagation.

\[
\Delta K_T = K_{\text{max}} + K_r.
\]

The values of \( K_{\text{max}} \) and \( K_r \) were evaluated by FEM analysis using a quarter of a specimen model that has a semi-circular crack. The relationship between \( \Delta K_{\text{th}} \) and the lengths of cracks was determined by the equation proposed by Tange et al. [2]. \( \Delta K_{\text{th}} \) calculated from experimental results are shown in Fig. 4, plotted as squares. The value of \( \Delta K_{\text{th}} \) obtained from the Tange Equation was consistent with that of \( \Delta K_{\text{th}} \) calculated from experimental results.

The intersection between \( \Delta K_T \) and \( \Delta K_{\text{th}} \) gives the maximum semi-circular slit size \( a_{\text{max}} \) that can be rendered harmless. From Fig. 4, the value of \( a_{\text{max}} \) was 1.09 mm. The evaluation result was consistent with the experimental results.

**Acknowledgement:** The authors express our appreciation to Toyo Seiko Co., Ltd. for peening treatment and measurement of residual stress.

**REFERENCES**


DEVELOPMENT OF EVALUATION METHOD THAT TAKES INTO ACCOUNT THE EFFECT OF THE FINE STRUCTURE OF ADHESIVE INTERFACE FOR DELAMINATION STRENGTH OF THE PACKAGING RESIN

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KEY WORDS: resin, adhesion reliability evaluation, fine structure

This paper describes the evaluation method of resin delamination strength at the adhesive interface. Power module consists of different materials. Difference of expansion between materials causes the delamination. The delamination behaviour at the interface between resin and Al or ceramics was studied in this study, and a new delamination test method was used to evaluate the delamination stresses.

Fig. 1 shows the new pudding-cup test method which can add a torque force to the resin interface by combining a pushing load. As a result, the delamination tests can be carried out by different balance of peeling stress and shear stress. Generally, the stresses are used to evaluate the delamination strength, however, it is difficult to include the effect of micro interface structure bonded compound. In this study, an interface structure was used to analyse the delamination deformation at the resin interface to develop an approach for the delamination behaviour. Deformation may be different from the other in spite of same stress in the interface according to condition of constraint. Therefore, stress in the interface is only as reference to evaluation of delamination strength of resin. Effect of the structure is very important to evaluation of delamination strength of resin. However, because of the complexity of the structure, it was difficult to recreate it in the analysis model. In this study, a method to simplify the interface structure was proposed. Deformation value of the interface is evaluated in this study. In order to reduce the load of analysis, a thin layer was introduced in place of the leg-shaped structure. The layer (interface layer) was operated material properties to agree with deformation of leg-shaped structure. New pudding-cup test was carried out with the condition that load span is 0, 40 and 100 mm. The layer model was analysed by using the load when the interface was delaminated in experiment. The deformation amount of the interface layer at certain load span compared at various load span at the delamination starting point.

Fig. 1. New pudding-cup test equipment.
Fig. 2. New pudding-cup test.
Fig. 3. Experimental results. Fig. 4. Analysis results graph of tensile stress-shear stress.

Fig. 3. shows experimental results. Fig. 4. shows analysis results by using experimental results. And Fig. 4. shows differences in balance of peeling stress and shear stress when load span is changed.

Fig. 5. shows analysis results by using a layer. The deformation of the interface layer was analysed by using the FEM model with considering the interface structure by the interface layer. The absolute value was calculated by the next equation.

$$\text{absolute value} = \sqrt{(in\text{-plane})^2 + (out\text{-plane})^2}.$$  \hspace{1cm} (1)

As a result, the analysis method to simplify the interface structure is a valid. Deformation absolute value in the delamination at the starting point was confirmed to agree at various load point in same material. This paper suggested effectiveness of hypotheses that delamination of the resin is deformation of interface structure. This analysis method clarify problematic delamination condition of resin. In future, this analysis method will be used easily when measurement method in leg size in leg-structure and physical property will be established. And this method contribute to improving the reliability of product.

REFERENCES


MECHANICS OF HERBERT PENDULUM HARDNESS TESTER AND ITS APPLICATION

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KEY WORDS: dynamics, herbert pendulum, hardness measurement, rolling resistance

The Herbert hardness tester was developed in 1923 [1]. It behaves as a pendulum, which swings from side to side, as the tip of indenter is a supporting point. New versions of original Herbert pendulum take advantage of modern measuring techniques and enable to investigate materials with various values of hardness ranging from lead to sapphire, see [2] and [3]. This conference paper is focused on explanation of basic physical principles, which describes the pendulum movement. Focusing on the case of elastic materials some interesting relationships are shown. As a consequence new possibilities of the modified Herbert hardness tester in testing of materials appear.

When rolling the centre of gravity of Herbert hardness pendulum is located below the contact surface, at the distance $e$ from the centre of the cylindrical indenter with radius $r$. The centre of gravity of the Herbert hardness pendulum follows so called cycloid, see the curve in Fig. 1. According to the picture the vertical coordinate can be easily expressed as

$$ y = r - e \cdot \cos(\phi). \tag{1} $$

Assuming very slow motion the influence of rotation speed is negligible and the force ($P$) applied in order to roll the pendulum over specimen is constant over time. The force $P$ do work on the length corresponding to the total length of arc, ie. on the angle $\phi_0+\phi_1$. Energy balance is considered in the situation at the beginning of the measurement $\phi_0$ (potential energy) and the situation at the end of the first half of oscillation when the angle $\phi = \phi_1$. Thus

$$ m \cdot g \cdot y_0 = m \cdot g \cdot y_1 + P \cdot r \cdot (\phi_0 + \phi_1), \tag{2} $$

where $m$ is weight and $g = 9.81 \text{ms}^{-2}$. Now, the rolling resistance coefficient [4] is defined similarly as the static friction coefficient

$$ c_{rr} = \frac{P}{G} = \frac{P}{m \cdot g}. \tag{3} $$

The property of the rolling resistance coefficient is that it is dimensionless. Generally, the rolling resistance coefficient depends on the load, surface roughness, velocity, radius $r$ etc. The angle $\phi_1$ cannot be expressed directly because (2) is a non-linear scalar equation, which can be solved iteratively:

$$ \phi_1 = \arccos \left( \frac{e \cdot \cos(\phi_0) + c_{rr} \cdot r \cdot (\phi_0 + \phi_1)}{e} \right). \tag{4} $$

![Diagram of pendulum](Fig. 1. Scheme of pendulum.)
In order to compare results of described energetic approach with the solution of linearized equation of motion we will consider in all further calculations these values: \( r = 1 \text{ mm} \), \( \varphi_0 = 30^\circ \), \( T_{10} = 200 \text{ s} \) (standardly used values in experiments [3]) and value \( e = 1.4 \text{ mm} \) (it has to be always precisely measured). We consider also very small value of rolling resistance coefficient \( c_{rr} = 0.02 \). Both approaches correlate very well, see Fig. 2a.

The regression line of energetic approach gives very good representation of the curve. On the other hand, the energetic approach shows a nonlinear change of swing angle increment, which is apparent from the Fig. 2b. A parabolic function describes the dependency between swing angle increment and number of swings very well. We can interpret the resulting graph at the Fig. 2 in such a way, that the dynamics approach becomes more accurate with decreasing value of the swing angle (linearization \( \sin \varphi \cong \varphi \)).

The solution of energetic approach is helpful for evaluation of the rolling resistance coefficient, because it can be expressed directly from the equation (4), thus

\[
  f_r = \frac{e \cdot (\cos(\varphi) - \cos(\varphi_0))}{r \cdot (\varphi_0 + \varphi)}.
\]

As it is clear now, the Herbert hardness tester can serve to determination of rolling resistance coefficient from so called scale hardness number \( \varphi_1 \) for various combinations of materials (indenter vs specimen), surface roughness, load, indenter radius etc. Results of our experimental study will be presented in the full paper of this contribution.

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**REFERENCES**


MECHANICAL PROPERTIES OF FUNCTIONALLY GRADED POROUS ALUMINUM OF DISSIMILAR ALUMINUM ALLOY

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KEY WORDS: foam, functionally graded

Porous aluminum is a lightweight material with excellent energy absorption properties. Because of these superior properties of porous aluminum, it is expected to be used in structural applications such as automotive components to improve fuel consumption and the collision safety of drivers, passengers and pedestrians[1]. Functionally graded (FG) metallic foams, in which the properties vary with the position, are expected to improve the performance of metallic foams. It is expected that FG metallic foams have controlled compression deformation behavior with the desired plateau stresses corresponding to the compression properties of metallic foams by controlling the pore structures or type of aluminum alloy at each position[2, 3]. However, accurate control of the pore structures and type of aluminum alloy at the desired position is relatively difficult; therefore, there have been few investigations related to the properties of the FG metallic foams as structural materials. To realize the accurate control of pore structures, Zhao and Sun developed a novel sintering and dissolution process (SDP) for fabricating open-cell porous aluminum on the basis of powder metallurgy[4]. In this SDP, Al powder and sodium chloride (NaCl) powder as space holders are mixed and sintered. Then, the sintered mixture is placed in water to remove the NaCl and obtain an open-cell porous aluminum.

In this study, FG porous aluminum consisting of two layers with low strength commercially purity A1050 aluminum and high strength A6061 aluminum alloy was fabricated. The pore structures of the fabricated porous aluminum were observed non-destructively by X-ray computed tomography (X-ray CT) to confirm that the pore shape was similar to that of the NaCl particles. In addition, compression tests were conducted to reveal the compression behavior of the fabricated FG porous aluminum by comparing it with those of uniform porous aluminum with A1050 and A6061.

The mixture with A6061 aluminum alloy powder and NaCl powder was first placed in a die. Then the mixture with A1050 pure aluminum powder and NaCl powder was placed in the die on top of the mixture with A6061 aluminum alloy powder and NaCl powder. Next, aluminum powders were subjected to spark plasma sintering. The sintering temperature, sintering pressure and sintering time were fixed at 570°C, 50 MPa and 3 min.

X-ray CT observations of the pore structures of the fabricated FG porous aluminum were conducted using an SMX-225CT microfocus X-ray CT system (Shimadzu Corporation). A cone-type CT system was employed, with which only one rotation of the specimen was sufficient to obtain a set of two-dimensional cross-sectional X-ray CT images of the entire specimen with a slice pitch equal to the length of one pixel in the X-ray CT image. FG porous aluminum was subjected to compression tests after the X-ray CT observations. Compression tests were carried out at room temperature in ambient air using an Autograph universal testing machine (Shimadzu Corporation)
by considering Japanese Industrial Standards JIS H 7902[5].

Fig. 1 shows photos and Fig. 2 shows cross-sectional reconstructed X-ray CT images of the fabricated FG porous aluminum. The white regions indicate the cell walls of porous aluminum and the black regions indicate pores. It can be seen that there are little differences between upper layer (A1050 layer) and lower layer (A6061 layer).

Fig. 3 shows a typical stress-strain curve for the FG porous aluminum obtained during the compression test, along with those of the uniform porous aluminum with A1050 and A6061. To enable the direct comparison of the stress-strain curves between the FG porous aluminum and the uniform porous aluminum, the nominal strain of the uniform porous aluminum was modified similar to that in a previous study[3]. It was found that the first and second plateau regions with low and high compression stresses, respectively, appeared independently in the FG porous aluminum and corresponded to the plateau regions appearing in the uniform porous aluminum.

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**REFERENCES**


EFFECT OF STRESS RATIO ON FATIGUE CRACK GROWTH THRESHOLD FOR STAINLESS STEELS IN AIR ENVIRONMENT

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KEY WORDS: fatigue crack growth threshold, stress ratio, austenitic stainless steel

Fatigue crack growth reference curve for ferritic steel in air and water environment is provided by Appendix A in ASME (American Society of Mechanical Engineers) Section XI [1], where fatigue crack growth threshold \( \Delta K_{th} \) (unit in MPa√m) in air and water environments is expressed by:

\[
\Delta K_{th} = \begin{cases} 
5.5(1 - 0.8R) & \text{for } 0 \leq R \leq 1 \\
5.5 & \text{for } R < 0 
\end{cases}
\]

(1)

where \( R \) is the stress ratio given by \( R = \frac{K_{min}}{K_{max}} \), \( K_{min} \) is the minimum stress intensity factor and \( K_{max} \) is the maximum stress intensity factor.

However, fatigue crack growth threshold \( \Delta K_{th} \) for austenitic stainless steel in air environment is not provided, although fatigue crack growth reference curve for stainless steel is provided by Appendix C in the ASME Code Section XI. Task Group in the ASME Section XI Code Committee has launched to discuss fatigue crack growth threshold \( \Delta K_{th} \) for austenitic stainless steel in air environment.

British Standards BS7910 [2] provides fatigue crack growth threshold of \( \Delta K_{th} = 2.0 \) MPa√m for austenitic stainless steels in air environment up to 100°C. API 579/ASME FFS Code [3] also provides \( \Delta K_{th} \) for stainless steels. When in air and at temperature less than 600°C and the yield stress is less than or equal to 600 MPa, \( \Delta K_{th} = 2.0 \) MPa√m. In addition, when upper bound crack growth data of \( da/dN \) curve is used, the threshold is given as \( \Delta K_{th} = 7(1-0.85R) \) MPa√m in air at room temperature or other non-aggressive environment.

Experiments on fatigue crack growth rates were performed in the world. Experimental data on \( \Delta K_{th} \) for austenitic stainless steels were collected for the wide range of \( R \) ratio. Figure 1 shows \( \Delta K_{th} \) data for Type 304 and 316 austenitic stainless steels in air and dry air environments, as a function of \( R \) ratio [4-7]. There are lots of \( \Delta K_{th} \) data with \( 0 \leq R \leq 1 \). On the other hand, \( \Delta K_{th} \) data with \( R < -1 \) is few. As can be seen in Fig. 1, the threshold \( \Delta K_{th} \) tends to increase with decreasing \( R \) ratio.

Figure 2 shows threshold \( \Delta K_{th} \) for Type 304 austenitic stainless steel in air at high temperature [5, 6]. The threshold \( \Delta K_{th} \) increases with increasing temperature and decreasing \( R \) ratio. Particularly, \( \Delta K_{th} \) is shown to be high value at \( R = -1 \) and 500°C.
When making codification, it is required for to express simple equations, convenience and conservativeness. Using Fig. 1, \( \Delta K_{th} \) for stainless steel at room temperature is derived from lower bound solid line, where \( \Delta K_{th} \) is constant value between \(-1 \leq R < 1\), and \( \Delta K_{th} \) increases linearly with decreasing \( R \) ratio. In case of high temperature, \( \Delta K_{th} \) for code is shown to be dotted line in Fig. 2, based on the room temperature line in Fig. 1, although there is no data at \( R < -1.0 \).

Conclusively, fatigue crack growth threshold \( \Delta K_{th} \) for austenitic stainless steels can be proposed as follows,

\[
\begin{align*}
\Delta K_{th} &= 2.0 & \text{for } -1 \leq R \leq 1 \\
\Delta K_{th} &= -2.0(2R+1) & \text{for } -5 \leq R < -1
\end{align*}
\]

for Room Temperature, \( \text{(2)} \)

\[
\begin{align*}
\Delta K_{th} &= 4.0 & \text{for } -1 \leq R \leq 1 \\
\Delta K_{th} &= -4R & \text{for } -5 \leq R < -1
\end{align*}
\]

for Temperature \( \geq 200^\circ\text{C} \). \( \text{(3)} \)

In case of temperature between room temperature and \( 200^\circ\text{C} \), linear interpolation is permissible.

REFERENCES


NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS
1 – 4, NOVEMBER, 2016, YOKOHAMA, JAPAN

POTENTIODYNAMIC POLARIZATION AND ELECTROCHEMICAL
IMPEDEANCE SPECTROSCOPY USED FOR PREDICTION OF NITINOL
STENT’S LIFETIME

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KEY WORDS: potentiodynamic polarization, EIS method, nitinol, stent, biomaterial

This paper is focused on prediction of nitinol stents lifetime. Electrochemical impedance
spectroscopy and potentiodynamic polarization in Tyrode’s solution was used for these
particular experiments. Stents are relatively small mesh tubes made by welt knitting of wires.
These tubes are placed into blood vessels, oesophagus or urinary system as a temporary or long-
term reinforcement. During application the shape memory effect is used and for proper function
the superelastic behaviour is necessary. Wires of 0.22 mm in diameter were used for testing.
The wires were manufactured from Nitinol #8 by Fort Wayne (USA). According to the
ASTM 2063 standard, nitinol alloy for surgical or implantology use can’t contain more than
500 ppm of carbon, 500 ppm of oxygen and 50 ppm of hydrogen. The material used completely
fulfils the requirements and possesses the corresponding thermo-mechanical properties as
shown in Table 1.

Table 1 Mechanical properties of used material.

<table>
<thead>
<tr>
<th>Material</th>
<th>Austenite start temperature (°C)</th>
<th>Cold worked</th>
<th>Heat treated for shape memory effect</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nitinol #8</td>
<td>(+10 to +35°C)</td>
<td>&gt; 1500</td>
<td>&gt;3</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&gt;1100</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&gt;10</td>
</tr>
</tbody>
</table>

After knitting the stents were stabilized by annealing in vacuum chamber at temperature
575°C for 5 minutes and cooled slowly also under vacuum. The pressure in chamber during these
procedures was 2 Pa. Thin titanium based oxide layer was formed on wire surface during the
annealing and caused characteristic blue-like colouring. The layer thickness is illustrated at Fig.
1 where cross-section of tested wire is captured. There is an assumption that this layer can acts as
an additional barrier against corrosion effect of body environment [1]. Four samples marked
SA1-SA4 were prepared for electrochemical testing. SA1 was reference as received state
sample without annealing. SA2 was only annealed, SA3 was annealed and 10% deformation
was applied. SA4 was deformed until rupture-18% deformation was applied. Relation between
electrochemical properties in conditions simulating body environment and deformation of the
wires was observed by electrochemical measurements. Potentiostat Voltalab PGZ 100 was used
for both methods. Basic corrosion properties were studied by potentiodynamic polarization
according ASTM F 746. Values of corrosion potentials, corrosion rates and polarization
resistance are illustrated in Table 2.
Table 2 Corrosion properties of tested samples.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Corrosion potential (mV vs. SCE)</th>
<th>Corrosion rate (µm/year)</th>
<th>Polarization resistance $R_c$ (kΩ/cm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Taffel method</td>
<td>Stern method</td>
<td>Taffel method</td>
</tr>
<tr>
<td>SA1-ref</td>
<td>238</td>
<td>235</td>
<td>0.34</td>
</tr>
<tr>
<td>SA2-0%</td>
<td>696</td>
<td>692</td>
<td>0.06</td>
</tr>
<tr>
<td>SA3-10%</td>
<td>430</td>
<td>428</td>
<td>0.58</td>
</tr>
<tr>
<td>SA4-18%</td>
<td>162</td>
<td>161</td>
<td>1.12</td>
</tr>
</tbody>
</table>

Electrochemical impedance spectroscopy was used for characterization of oxide layer. Frequencies between 100 kHz and 100 MHz and amplitude 50 mV were used during the experiment. According to results of the test equivalent electric circuit model was set up and parameters of each component were found. Electric circuit is illustrated at Fig. 2, where $R_2$ is resistance caused by solution and contact (usually very low), $R_1$ and CPE are parameters of oxide layer [2]. According to equation of ZARC element capacitance of oxide layer was calculated. Values of capacitance found by EIS method are illustrated in Table 3.

Table 3 Capacitance of oxide layer.

<table>
<thead>
<tr>
<th>Samples</th>
<th>SA1-ref</th>
<th>SA2-0% def.</th>
<th>SA3-10% def.</th>
<th>SA4-18% def.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Capacitance (µF/cm²)</td>
<td>41</td>
<td>7</td>
<td>10</td>
<td>15</td>
</tr>
</tbody>
</table>

By potentiodynamic polarization was confirmed that titanium based oxide layer on surface positively affect corrosion behaviour-corrosion potential of annealed samples was more noble than unannealed reference sample. Also corrosion rate was lower. It was found that deformation of wire can damage the oxide layer, decrease corrosion potential and increase corrosion rate. Sample covered by undamaged layer has also significantly lower capacitance than reference samples. Higher values of samples capacitance after deformation indicate that oxide layer was cracked and disrupted.

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REFERENCES


EFFECT OF MICROSTRUCTURE AND TEXTURE ON THE BEHAVIOR OF CYCLIC BENDING DEFORMATION AND FRACTURE OF ALUMINIUM

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KEY WORDS: cyclic bending, orientation change, crack initiation site, deformation continuity

The authors have studied the characteristics and mechanisms of deformation and fracture of cyclic bending on aluminium and Al-Mg alloy wires and reported that grain size had a strong effect on the fatigue life and the cracks were preferentially formed on the grain boundaries at the outer surface of the specimens [1]. Also it was shown that the slip deformation in the crystal grains might be related to the crack initiation. However, the crystallographic analysis as well as the examination of the effect of microstructure on the crack formation was not enough to clarify the role of the slip deformation inside the crystal grains. Thus, the characteristics of the site of crack initiation is investigated from the viewpoints of microstructure, orientation distribution along grain boundaries, activated slip systems and deformation continuity at the grain boundaries.

Aluminium wires with the rectangular cross section of 0.8 x 0.4 mm are prepared by wire drawing at room temperature. The purity of the aluminium is 99.99% ~ 99%. The grain size of the wires is adjusted to be about 100 μm in the recrystallized state by the heat treatment after the wire drawing. The wire is cut to 200 mm in length for the cyclic bending tests by the system shown in Fig. 1. The angle and the rate of bending tests are 90° and 50 rpm, respectively. The maximum bending strain is 0.02 at the outer surface. Before the tests, microstructure is observed by the optical microscopy. Then, the observed area is successively traced up to 200 cycles by the scanning electron microscope equipped with EBSD (Electron Backscatter Diffraction) system.

Figure 2 is an example of the orientation map derived from EBSD measurements. The measurement is conducted with 1 μm interval in this case. Figure 2 (a), (b), and (c) correspond to the microstructures (a) before cyclic bending, (b) after 30 cycle bending and (c) after 100 cycle bending, respectively. Before the bending (a), no obvious orientation differences are seen within crystal grains, suggesting that the material is in the recrystallized state. After 30 cycle bending (b), orientation change is seen in the grain 1 along the grain boundary with the grain 2. After the bending up to 100 cycles (c), slip lines obviously develop in the grains and the orientation change along the grain boundary becomes extensive. At this stage, formation of a crack on the grain boundary is confirmed by SEM observation. The crack appears in Fig. 2(c) as the black region on the grain boundary between grain 1 and 2. As shown in (b) and (c), orientation change occurs in the grain 1 along the grain boundary, and it is seen that the amount of orientation...
change is not constant but varies along the grain boundary. It is considered that this results in the changes in slip activity and the slip systems site to site in the region neighbouring the grain boundary. Thus it is expected that strains generated by slip deformation vary depending on the site in the grain boundary region.

Detailed observation of the SEM micrograph shows that many small cracks exist separately on the grain boundary. Namely, it is found that cracks are formed on the site where specific conditions are fulfilled. Strain components close to the grain boundary is calculated on ten sites along the grain boundary with many cracks. It is found that that cracks are formed on the site where strain component normal to the grain boundary is relatively large. It is suggested that the crack initiation during the cyclic bending deformation is attributable to the orientation change along the grain boundary caused by the plastic deformation at the outer surface during the cyclic bending.

REFERENCES

RELATIONSHIP BETWEEN ELECTROCHEMICAL PARAMETERS AND DEGRADATION PROCESS OF GLASS FLAKE COATINGS

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KEY WORDS: glass flake coating, electrochemical parameter, degradation, blister, impedance

Organic coating containing glass flakes on bottom plate of oil storage tank is used in corrosion protection of a few decades. Degradation of the coating is one of important factor which dominates the inspection period of oil storage tank like back-side local corrosion. The impedance method is a non-destructive inspection technique. However, in the present situation, the measured data and the analysis result aren't sufficiently utilized in fields. In this study, the degradation state of a real tank is evaluated by new analytical method using two electrochemical parameters.

In recent years, and high-performance coating is used for corrosion protection of large steel structures, such as infrastructure. In particular, an organic resin coating containing glass flakes is applied to the inner surface coating of oil tanks. Glass flakes in the coating has the effect to delay penetration of water. As a result, anti-corrosion effect is higher than the general coating. Performance of the coating in the oil tank is up to about 30 years. However, the state of degradation is not clear. In addition, the degradation process is also not been elucidated. In this study, the degradation state of a real tank is evaluated by electrochemical techniques. Impedance measurement method of coating on bottom plate of oil storage tank is shown in Fig.1. First, film thickness of the coating was measured. The carboxyl methyl cellulose sodium was sandwiched between the working electrode (coating on plate) and the counter electrode (aluminium foil of 100 cm²). Measurements were performed using the LCR meter (Agilent Technologies 4284A), the frequency range was measured from 20 Hz to 1 MHz. Data were analyzed using Zview software, developed by Scribner Associates, Inc.

Figure 2 shows degradation state of surface of coating and under coating. In the case of visual inspection, blistering was observed on the coating surface in all cases. However, after the peeling of the coating film, the three degradation processes was confirmed. Their morphology is no rust, black rust only, black rust and red rust.
Figure 3 shows all impedance spectra (17 field point data) obtained for a vinyl ester resin organic coating (containing glass flakes, average coating thickness of 550 µm) of internal bottom plate of an oil storage tank. These impedance behavior show various degradation levels of the coating. However, these data depend on the frequency in wide range, it is difficult to evaluate at a certain frequency.

Degradation behaviour of protective coating used electrochemical method has been explained by Hirschorn etc. [1-3]. At the initial stage, the non-degraded coating can usually be represented by a capacitance. However, this element does not satisfactorily explained water and some ions penetration process in the high performance coating. Therefore, the authors used an equivalent circuit model which contains the constant phase element (CPE). The CPE was replaced with capacitance and resistance of bulk of coating. Figure 4 shows electrical equivalent circuit model for blistering coating systems. The relationship between the sound areas (CPEs) and damaged areas (CPEd) were examined in this model. Two circuit elements are significant of initial blistering process. When water and some ions penetrate in the coating, \( T \) and \( n \) of CPEs, CPEd parameters will be changed.

Figure 5 shows plots of \( n_d \) versus common logarithm of \( T_d \). In this relationship, a positive correlation exists between \( n_d \) and log \( T_d \). In addition, if blistering occurs, \( n_d \) becomes a large value. In the case of under-film corrosion, \( n_d \) value is same or slightly smaller. Therefore, this result suggests a correlation between the degradation state of coating and electrochemical parameters. As a conclusion, we have completed a new quantitative degradation evaluation method of coating using two electrochemical analyzed parameters.

REFERENCES


EVALUATION OF ECONOMICAL RISK USING BAYESIAN THEOREM FOR STRUCTURAL HEALTH MONITORING

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KEY WORDS: statistical analysis, probability of detection, risk analysis, probability of failure, bayesian theorem

This study is about the method for numerical evaluation of probability of failure using Bayesian theorem, from diagnostic result of real-time condition monitoring. When performing maintenance based upon the results of real-time monitoring, it is ideal to obtain diagnostic results without inspection error. However, failure does not occur if the monitoring method overestimates sufficiently small damage; indeed, failure does not occur even if there is a slight underestimation of a large damage. To reduce the PoF, it is important to accurately estimate the specific damage. This study proposes a method for diagnosing the specific damage level with improved accuracy; this improved accuracy is achieved by using a weight function to control the sampling ratio of the training data for learning. The consequences of overestimation and underestimation of damage differ. The risk caused by the underestimation is called failure risk, and that caused by the overestimation is called economic risk. This paper discusses the shape of the weight function used to reduce the economic risk. And for the validation of the method, proposed method is applied to the delamination identification problem of CFRP beam using the electric potential method.

Procedure of estimation of the probability of failure over the estimated result: The probability of failure (PoF) is estimated by the following formula by the limit state function method.

\[ P_{of} = P[g(R - S) < 0], \]

where, \( R \) is strength, \( S \) is applied force and \( g \) is the limit state function. The procedure of probability of failure estimation over the estimated result is shown in Figure 1. First, the occurrence probability distribution of the true damage size over the estimated size is deduced by the Bayesian theorem[1]. Residual strength is the function of the damage properties. In this paper, the buckling failure because of the delamination crack is assumed and the distribution of residual buckling strength is calculated by the proposed method.

Evaluation of the accidental risks and the economic risks: Figure 2 shows the evaluated accident risk and economic risk from this method. The accident risk is risk of the failure caused by underestimation of the damage size. On the other hand, the economic risk is risk of unnecessary maintenance costs caused by overestimation of the damage size. The vertical axis shows the adjusted PoF (PDF of POF vs PDF of occurrence of estimated damage size: PoO) and the horizontal axis shows the estimated size. The accident risk is evaluated by the area surrounded by the adjusted PoF and the threshold for maintenance operation. The threshold is
set as the accident risk takes constant value (0.03). The economic risk is evaluated by the area surrounded by the adjusted PoF, the PoO and the threshold. In the other words, reducing the economic risk is the purpose of this study.

Estimation of the probability of failure from proposed method: The proposed method is validated by applying it to the delamination identification problem of a CFRP beam using the electric potential method[2]. The result is shown in Figure 3. The abscissa is the estimated size, and the vertical axis is the probability of failure. The external force is assumed in order that failure might arise in an average of 15 mm delamination in this case. As shown in the figure, the probability of failure started the lifting by the damage smaller than 15 mm, and it is saturated with about 17 mm. From the results, it can be said that it is possible to evaluate the probability of failure over the estimated result by the proposed method.

Conclusions: This research proposes a method for the numerical evaluation of PoF using the Bayesian theorem and employing diagnostic results of real-time condition monitoring. The proposed method uses the Bayesian theorem to determine the occurrence probability of the cause event. As the result, PoF starts the lifting by the damage smaller than the critical level, and is saturated with over the critical level. In conclusion, this study confirms that the proposed method evaluates the PoF at arbitrary estimated result.

REFERENCES
DEGRADATION MECHANISM OF DISSIMILAR METAL WELD JOINTS ON STEAM GENERATOR COLLECTORS VVER 440MW TYPE

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KEY WORDS: degradation, dissimilar weld, steam generator, cracks

The Unit WWER 440MW has six loops with six horizontal steam generators (SG). Steam generators WWER 440 are horizontal pressure vessels with vertical cylindrical hot and cold collectors. Collectors are made from stainless steel SA321 type, as well as main primary cooling piping system. Steam generator body is made from carbon steel. Connection of stainless steel (SS) with carbon material (CS) thus creates dissimilar metal welds (DMW), appears to be critical to the long term operation of the SGs. The Units WWER 440MW has twelve critical DMWs on the SG collectors.

The weld connection is carried out through two weld layers, wherein the first deposit is with high nickel content (Fig. 1). Chemical composition of the base material and the first weld deposit material are presented in Table 1. NDT controls have started to detect indications in this DMW after 20 years of operation. Detected indications have started slowly grew-up every year on two SGs, so it had to be approached to repair of these welds. The presence of defects in DMWs during repair has been confirmed. Repair technology was designed so that during the repair could be created the DMW test sample with defects to perform experimental analysis to determine the principal causes of DMW damage. The comprehensive experimental program to determine the root cause of the damage was proposed.

Table 1 Material properties of base material and first high nickel deposit.

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>S</th>
<th>P</th>
<th>Mo</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>Carbon steel</td>
<td>0.23</td>
<td>0.4</td>
<td>1.00</td>
<td>0.3</td>
<td>0.3</td>
<td>0.03</td>
<td>0.03</td>
<td>0.4</td>
<td>-</td>
</tr>
<tr>
<td>First deposit</td>
<td>0.11</td>
<td>0.47</td>
<td>1.74</td>
<td>15.6</td>
<td>25.3</td>
<td>0.017</td>
<td>0.015</td>
<td>5.85</td>
<td>0.184</td>
</tr>
</tbody>
</table>

Experimental work demonstrated high sensitivity of the material of the first deposit on the stress corrosion cracking (SCC). It was the first project mistake - the material of the first deposit was not tested on SCC. Root of the first deposit is in contact with corrosive medium, which is the second project mistake. Post weld heat treatment (PWHT) was done after first deposits during manufacturing process. Material of the first deposit is not stabilized steel, so that due to a gradient the carbon diffused into the material of the first deposit and carbide M₂₃C₆ was created to a depth of 50 microns. The crack grows at this depth, as shown in Fig. 2. PWHT application during manufacturing process is third project mistake because different thermal expansion of carbon and stainless steel increase level of residual stress during cooling. Metallographic analysis showed a large corrosion of the base material along the length of the crack. It seems two degradation mechanisms causes damage to the DMW.
The DMWs are located at the lowest point of the SG in the area of closed pockets on secondary side (Fig. 1). Corrosion products of secondary side are deposited on the lowest point of the SG and it’s creating a corrosive environment with higher concentrations of salts. Conditions for the development of electrochemical corrosion due to a difference in the electrochemical potential of the carbon and stainless steel are made. Electrochemical corrosion produces pitting, which are significant stress concentrators. Stress from differing thermal expansion of dissimilar materials is even increased by stress concentration factor (SCF). This process created conditions for second degradation mechanism - corrosion cracking. The crack is initiated from a corrosion pits during the heating of Units, the stress relaxation occurs at the crack tips and stopping its growth. Electrochemical corrosion in the closed area of the pocket starts degradation of DMWs again. Described degradation mechanism during Unit operation is continuously repeated. It can be seen on Figure 2.

We can do following conclusions:

1) Three mistakes in project can be defined – material of first deposit was not tested on SCC, deposit root with high residual stress is in the contact with corrosion medium and application of PWHT after deposit increase tensile residual stress.

2) The initiation of the DMW degradation is caused by the electrochemical corrosion, which creates stress concentration on the border with basic carbon steel.

3) Electrochemical corrosion velocity is determined by the amount of corrosion products in the collector pocket. Removing corrosion products from the SG pocket can manage the initiatory degradation mechanism.

4) The main degradation mechanism of DMW damage is SCC. The DMW is damaged on the boundary of the diffusion layer, wherein the base matrix has lower chromium content and undesirable carbides on grain boundary are examined.

5) The degradation process of DMW damage is repeated due to cyclic startup of Units.

6) Operation condition, mainly temperature, leads to the formation of secondary carbides due to diffusion process because first deposit is not stabilized steel. Diffusion process extends the area with lower chromium content and thereby increases the area for SCC.

7) Degradation mechanism can be managed by regularly removing corrosion products from the pocket of SGs and maintaining a low salt content in the pocket PG. Corrosion rate will be reduced.

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NON-DESTRUCTIVE STRESS EVALUATION OF A TOOL STEEL USING A SCANNING HALL PROBE MICROSCOPE: EFFECT OF STRESS DIRECTION ON THREE DIMENSIONAL MAGNETIC FIELDS

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KEY WORDS: non-destructive evaluation, stress, magnetic field, tool steel

Fatigue under cyclic loadings causes failures of machine components and civil structures. In order to avoid the failures, non-destructive crack detection methods have been developed. Recently, these methods are applied to ‘structural health monitoring’ systems. In our previous works, scanning Hall probe microscopes (SHPM) were developed to observe magnetic fields around fatigue cracks of steels. The SHPM was used in air at room temperature and covered whole crack tip area [1-4]. Although these previous researches helped to obtain the basic features of magnetic fields under fatigue (stress conditions, plastic deformations and crack growth), the relevance between these factors was not discussed. We believe the next step of the SHPM as an application of damage evaluation is cyclic stresses whose levels are below the yielding stress. Basing on our previous measurement into change in magnetic field of a steel plate including no hole under a single tensile loading [5], in the present work, three dimensional observations of magnetic fields of steel samples (JIS SKS93, tool steel) under tensile loads were carried out to further investigate the effect of stress on magnetic fields.

A box-shaped specimen of SKS93 tool steel (JIS B 4404: 2006, equivalent to AISI W4 tool steel) in the as-received condition (hardness, HV 191) were prepared in the present tests. Fig. 1 is the schematic illustration showing dimensions, applied stress direction and coordinates. The center of the coordinate system ‘o’ is on the one side of the specimen. After normalizing the initial magnetic fields using a magnet coil, the center of the specimen was magnetized with a permanent magnetic block, whose size and surface inductive flux \( B \) were \( 1(x)-10(y)-10(z) \text{ mm}^3 \) and 99 mT. The block was slid on the \( Y \)-axis to magnetize the specimen. The magnetic fields were measured in the area including the magnetization area. When applying a tensile stress through a stainless bolt along \( X \)-axis, both sides of the box-shaped specimen are elongated.

![Diagram](image)

(a) Dimensions of the specimen. (b) Stress direction and coordinates.

Fig. 1. Coordinate axis of specimen and applied stress direction.

While five levels of stresses ranging from 100 MPa to 300 MPa were applied to the specimen, three-dimensional magnetic fields \( (B_x, B_y \text{ and } B_z) \) were observed. For each stress level, a square observation area \((-6 \text{ mm} \sim +6 \text{ mm} (X), +4 \text{ mm} \sim +16 \text{ mm} (Y)) \) was divided into twelve hundred ‘10 µm-width segments’ whose longitudinal axes were parallel to the \( X \)-axis. The ranges of maximum and minimum values, \( B_{range} \) of the three-dimensional magnetic fields in all segments were calculated. After the \( B_{range} \) calculations, the maximum range value \( B_{r,max} \)
among all segments was obtained for each stress level. After the $B_{r,max}$ was compared to the initial magnetic component in the segment $B_{r,i}$, ratio ($R_B$) of the change in the range, ‘$B_{r,max} - B_{r,i}$’ to the initial range, $B_{r,i}$ before testing was calculated using the following equation.

$$R_B = \frac{B_{r,max} - B_{r,i}}{B_{r,i}}.$$  \hspace{1cm} (1)

Fig. 2 shows the ratios $R_{Bx}$ of $Bx$. We can see the data converge to a straight line. When comparing the features of all three-dimensional components of magnetic fields, $Bx$, $By$ and $Bz$ and stress values, it is found that the $Bx$ is more sensitive to applied stress than other two components. The maximum stress value of the present tests is about half the yielding stress of the material. From these results, we can conclude that the magnetic field component that is parallel to a tensile loading direction is strongly correlated to stress values. This factor is applicable to stress measurement.

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EFFECT VERIFICATION OF WELD-PERIPHERY HEATING ON WELDING SOLIDIFICATION CRACK PREVENTION FOR LASER WELDING OF THIN STEEL PLATE

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KEY WORDS: welding solidification crack, weld-periphery heating, laser welding, FEM, temperature-dependent interface element

Laser welding contributes to weight reduction of automotive components such as hat-shaped parts because it can reduce the flange widths compared to resistance spot welding. On the other hand, welding solidification crack tends to occur in the case of laser welding near the flange edge [1]. A possible cause for the cracking is as follows: shorter flange width has lower resistance to rotational deformation behind a weld zone caused by the difference in longitudinal thermal expansion between the welding point and its peripheral area. This leads to an increase in the incremental plastic strain in the brittleness temperature range (BTR). The cracking occurs when the strain reaches a critical strain.

One possible preventive measure for welding solidification cracking is weld-periphery heating. Its concept is shown in Fig. 1(b): the heat source for weld-periphery heating runs parallel that for welding, heating near the weld zone to cancel or reduce the difference in longitudinal thermal expansion between the weld zone and its peripheral area. This heating is expected to reduce the rotational deformation. However, little is known about the effect of weld-periphery heating. Therefore, finite element (FE) analysis and experiment were conducted to clarify the effect of the heating.

FE analysis was conducted as preliminary verification before the verification test. In order to model the crack formation and propagation, the finite element method using a temperature-dependent interface element [3] was adopted. The FE analysis results implied that cracking occurred when welding near the flange edge and the weld-periphery heating prevented the cracking.

Then, the verification test was conducted under the conditions of Table 1. Figure 2 shows the schematic drawing of the test. The specimen, an austenitic stainless steel SUS310S, is fixed at one side end by a clamp. The specimen is 100 mm wide, 100 mm long, and 1 mm thick. The weld zone and weld-periphery heating zone are heated by a twin laser beam. No crack was observed when the weld-periphery heating was adopted under the condition that the cracking occurred in the case of welding without weld-periphery heating as shown in Fig. 3. As predicted by the FE analysis, it is probable that weld-periphery heating reduces the incremental plastic strain in BTR, which results in no cracking.
From these results of the simulation and the experiment, we conclude that weld-periphery heating prevents or reduces the occurrence of welding solidification cracking.

Table 1 Laser welding conditions for the weld solidification cracking test.

<table>
<thead>
<tr>
<th>Shielding gas</th>
<th>Focal length of focusing lens (μm)</th>
<th>Spot diameter (mm)</th>
<th>Welding speed, v (mm/s)</th>
<th>Output power of the laser beam (kW)</th>
<th>Heat input (J/mm)</th>
<th>Distance from the side edge to the weld line, d₁ (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ar 100%</td>
<td>600</td>
<td>0.6</td>
<td>30</td>
<td>1.5</td>
<td>50</td>
<td>3-10</td>
</tr>
</tbody>
</table>

![Fig. 2. Schematic drawing of the solidification cracking test in laser welding.](image)

![Fig. 3. Observed cracking (v = 30 mm/s, d₁ = 3 mm, d₂ = 0.25 mm).](image)

**REFERENCES**


ESTIMATION OF FRACTURE TOUGHNESS PROPERTY USING FLAT PUNCH INDENTATION TEST

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KEY WORDS: fracture toughness, instrumented indentation test, flat punch, cracked round bar

Fracture toughness is important material property to ensure structural integrity. However, it is difficult to measure fracture toughness in-service because the test process is destructive and complex. Since Instrumented Indentation Test (IIT) is non-destructive method, it is expected to be an alternative method to measure various mechanical properties in field[1].


In this research, flat punch shaped indenter was suggested to estimate fracture toughness instead of spherical shaped indenter. Since its geometrical similarity between flat punch indentation and cracked round bar (CRB) fracture test, stress distribution beneath the indenter is similar with that ahead of crack in CRB specimen.

Even though, unlike the fracture toughness test, there is no crack initiation in indentation test. To determine the crack initiation point in indentation test, limit load concept and fracture toughness standard test were adapted. Also, two fracture estimation model were suggested according to fracture behavior of the material; ductile and brittle.

To verify the model, the J-integral fracture toughness test and flat punch instrumented indentation test were performed. All specimens have an orientation corresponding to loading in the longitudinal direction and crack propagation in the transverse direction from rolled plate. 5 brittle fracture behavior material and 9 ductile fracture behavior materials were prepared and the results of the test is shown in Fig. 2.
Fig. 2. Results of fracture toughness estimation model. (Left) Ductile model, (Right) Brittle model.

REFERENCES


LOW-STRESS CREEP IN NEW SANICRO 25 STEEL AND ITS RELATION TO LONG-TERM CREEP LIFE

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KEY WORDS: creep, superaustenitic steel

Ever growing demands for the higher efficiency of power plants are leading to higher temperatures of the working media, but this trend is limited by the creep life of the structural materials. Thus new materials are developed to meet the requirements. Process of creep deformation is too slow under conditions corresponding to industrial use of the steels. Laboratory experiments have to be accelerated by application of higher temperature and/or stress. Subsequent extrapolation process may be a source of potentially dangerous errors. Extrapolation can provide false results mainly if the creep deformation mechanism is changing. Evidences were presented that such change at very low creep rates occurs [1]. In the most cases, both stress and temperature dependencies of the creep rate become weaker at low stresses. As was demonstrated by Kimura et al. [2], the creep life standards based on extrapolation dangerously overestimate the real time to fracture for extremely long creep tests of about 100,000 hrs.

It is well known that cavities which are able to growth to dangerous sizes are nucleated during primary creep [3]. The primary stage can be assumed as an important indicator of overall creep properties. Thus, any models of creep life should be able to describe the primary creep stage correctly. Unfortunately, most current approaches tend to ignore the primary stage completely, which makes their reliability questionable.

Primary creep can be measured experimentally even at conditions close to application ones, if high strain sensitivity creep technique is employed. Helicoid spring specimens technique [4] have been used successfully for many materials.

Example of the primary creep curve obtained for the Sanicro 25 creep resistant steel is in Figure 1. The curve can be fitted by the Li equation [5] successfully

\[
\varepsilon = \dot{\varepsilon}_s t + \dot{\varepsilon}_s \tau \log \left( 1 + \left( \frac{\varepsilon}{\dot{\varepsilon}_s \tau} \right) - 1 \right) \left( 1 - \exp \left( -\frac{t}{\tau} \right) \right)
\]  

(1)

where \( \varepsilon \) is creep strain, \( \dot{\varepsilon}_s \) is secondary stage strain rate, \( \tau \) is primary stage relaxation time, \( \varepsilon_t \) is primary transient strain and \( t \) is time.

With the equation above, secondary stage creep rate can be estimated and then used in Monkman-Grant relation to derive time to rupture. For the particular case in figure 1, the creep rate \( \dot{\varepsilon}_s = 6.9 \times 10^{-12} \text{s}^{-1} \) was obtained, from which the time to rupture \( t_f \approx 7 \times 10^6 \text{ hrs} \) can be derived. In contrast, the extrapolation in [6] gives \( t_f \approx 12 \times 10^6 \text{ hrs} \) for the given conditions. The former value is definitely more conservative and probably more realistic.
Fig. 1. Creep curve of Sanicro 25 superaustenitic steel at 700°C and 39MPa.

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REFERENCES


NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS
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STRAIN DISTRIBUTION ANALYSIS ON CYCLICALLY DEFORMED HIGH STRENGTH STEEL USING DIGITAL IMAGE CORRELATION

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KEY WORDS: strain distribution, digital image correlation, replica, low cycle fatigue

INTRODUCTION

Digital image correlation (DIC) [1-2] has been developed as a convenient strain analysis method calculating strain from the difference of images between before and after deformation. It shows an advantage to apply to any deformation mode or materials as long as significant contrast without large strain. On the other hands, replica method is commonly adopted to detect a crack growth on the specimen surface during fatigue test [3]. Hamada et al. [4] demonstrated the DIC strain mapping with high spatial resolution by replica film, in which the strain distribution on replica was almost the same with that on its specimen surface. Then, it is possible to obtain strain distribution data intermittently from small strain to just before fracture of the specimen using the replica. In this study, we have demonstrated strain distribution around the crack from its initiation stage to propagation one under low cyclic loading test by means of DIC for replica.

EXPERIMENTAL PROCEDURE

A high strength steel sheet (Fe-0.16C-0.4Si-2.0Mn, in mass%) hot-rolled in the laboratory was used. The steel was solution treated at 1193 K for 0.09 ks and subsequently followed by water-quenched, and then was tempered at 833 K for 0.09 ks and air-cooled. Fatigue test was carried out using round-bar specimens with 10 mm in gauge length and 4.5 mm in diameter under total strain (Δε_t) control from 0 to 0.01 with minimum strain (Δε_min) 0. An acetyl cellulose replicating film (Bioden R.F.A.) was immersed in methyl acetate solution and then was placed on the specimen surface etched with 5% Nital solution. The replica films were taken at every 100 cycles until a large crack was detected. They were observed by scanning electron microscopy (SEM), since platinum thin layer was vapor-deposited on the replica film. SEM observation was conducted at 2.7 keV accelerated voltage as low as possible for the damage of replica film by the electron beam. The DIC on the films was done using VIC-2D software (Correlated Solutions Inc.), where the images at 200 and 1700 cycles were chosen as the reference for visualizing strain distribution, respectively.

RESULTS

The replica on the specimen surface after 2000 cycles with total strain range Δε = 0.01 successfully transcribed an intrusion and extrusion due to the localized slip deformation. Although no such surface protrusion image related to fatigue crack generation was observed until 1900 cycles, the localized plastic strain distribution in the area was clearly detected after 200 cycles in which tensile strain along loading direction concentrated at the site. The intrusion and extrusion is generated by the persistent slip bands on the specimen surface and results in the formation of crack. Thus, the DIC plastic strain mapping using replica film showed an advantage to detect highly localized and accumulated strain under cyclic deformation. Its spatial
resolution was enough to analyze the strain distribution even though small cyclic strain regime. The crack propagated in the high tensile strain region. It suggests that the plastic strain distribution is one of the important factors controlling fatigue crack initiation and propagation.

REFERENCES


SOLID-STATE DIFFUSION BONDING OF TITANIUM BY USING METAL SALT COATED ALUMINUM SHEET

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KEY WORDS: surface modification, fracture, bonding strength, titanium, aluminium, formic acid

In the past, methods such as brazing, friction stir welding and laser welding have been used for bonding titanium alloy. However, these techniques have some shortcomings: (1) micro cracks are developed owing to the softening of the weld zone; (2) the gap in the weld zone results in corrosion; and (3) a high heat input is required to compensate for high heat radiation from titanium. In addition, tools used for friction stir welding have short lifespan and this translates to higher running costs. Moreover, aluminium is an excellent heat radiating and electricity conducting element; therefore, it is difficult to bond titanium and aluminium using other welding methods. Because of these limitations, solid-state diffusion bonding is considered to be the most suitable method for bonding materials at low temperatures. In recent years, the applicability of low temperature and low deformation bonding in industrial processes has been explored in an effort to contrive miniaturization and weight reduction of medical equipment and transport equipment. In an earlier study, we showed that modification of an oxide film with formic acid greatly improves the strength of bonding between tin and tin and copper [1, 2]. Therefore, in this study, in order to achieve bonding at low temperature and low deformation of titanium, an insert material with metal salt coating has been developed.

In this paper, the effect of metal salt coating processing of aluminium surface on the bond strength of the solid-state diffusion bonded interface of titanium and aluminium has been investigated by SEM observation of the interfacial microstructures and fractured surfaces. A cylindrical Ti specimen (Table 1) with dimensions of φ10 mm × 20 mm, Ti plate specimen (Table 2) with dimensions of 18 mm × 15 mm × 5 mm³ and pure Al sheet specimen (0.5 mm³, 99.999% purity) were used in this study. The metal salt coating processing was carried out by boiling the Al sheet surface in formic acid (98%) for predetermined time after modifying it with 5% NaOH(aq). Solid-state diffusion bonding was performed in N₂ gas at bonding temperature of 713-773 K under a load of 12 MPa (bonding time of 900 s).

Fig. 1 shows the relationship between the bonding temperature and tensile strength of the joint. In order to illustrate the effect of metal salt coating processing, the corresponding

---

**Table 1 Chemical composition of titanium rod.**

<table>
<thead>
<tr>
<th>Elements</th>
<th>H</th>
<th>O</th>
<th>N</th>
<th>Fe</th>
<th>C</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt%</td>
<td>0.0012</td>
<td>0.109</td>
<td>0.004</td>
<td>0.034</td>
<td>0.004</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

**Table 2 Chemical composition of titanium plate.**

<table>
<thead>
<tr>
<th>Elements</th>
<th>H</th>
<th>O</th>
<th>N</th>
<th>Fe</th>
<th>C</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt%</td>
<td>0.0013</td>
<td>0.200</td>
<td>0.003</td>
<td>0.025</td>
<td>0.008</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

---

**Fig. 1.** Effect of metal salt coating processing on the relation between bonding temperature and tensile strength.
relationship for a non-modified joint are also shown. As shown in Fig. 1, the tensile strength increased with bonding temperature irrespective of metal salt coating processing. However, the tensile strength of the joint reached 100 MPa by performing the metal salt coating processing on the Al sheet. On the other hand, at the bonding temperature of 773 K, the tensile strength of the joint slightly decreased when the metal salt coating processing were applied.

To examine the factors determining fracture at the bond interface, the area of the fractured surface was observed. As shown in Fig. 2, when the metal salt coating processing was not applied, substances are not observed to adhere to either surface. When the metal salt coating processing was applied, the fractured surface started to show ductile fracture characteristics, although it was not observed when the metal salt coating processing was not applied.

Fig. 3 shows the FT-IR spectra of the Al surface that is treated by formic acid for metal salt coating processing after modifying the surface with NaOH(aq). Also, the result from the analysis points toward the formation of Al formate. From these results, it is understood that dehydration reaction and endothermic reaction occur when Al formate undergoes thermal decomposition at around 323-423 K and that fine Al oxide particle is formed at temperature around 573 K. Thus, it was hypothesized that high-tensile strength joints were obtained at a lower bonding temperature with metal salt coating processing because the contact is between atomic plane of Al and Ti bonding surface was increased in the bonding process.

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**REFERENCES**


LBB APPROACH USABILITY STUDY FOR CORROSION DEFECT LEAKAGE DETERMINATION

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KEY WORDS: Leak Before Break (LBB), corrosion defect, vent pipe, Steam Generator (SG), NPP

During the start-up of the second unit of Temelin Nuclear Power Plant (NPP), after outage for refueling, there was identified internal primary circuit leak inside the steam generator (SG) number 4. The cause of this leak was a corrosion defect on SG collector’s vent pipe. Vent pipe is pipe of $\varnothing 16 \times 2.5$ mm dimension, which goes from the SG collector through the secondary side out of SG and is connected to the venting system. When the event has occurred due to corrosion erosion damage in weld joint’s heat affected zone of the venting tube, which goes through the SG secondary side, it caused leakage of medium from the primary circuit (I.C.) into the secondary circuit (II.C.) and other systems.

Real size and shape of the corrosion defect was determined, and subsequently the leakage value was calculated using the standard LBB methods for leakage determination. Because standard LBB methods for leakage determination are assuming defect type of a flaw, it wasn’t sure if these methods could be used also for this type of corrosion defect.

Therefore a special CFD analysis, with the aim to verify usability of the standard LBB approach for corrosion type of defects, was carried out. Under normal operating conditions is inside the vent pipe circulating the primary coolant medium, and outside of the pipe is steam.

LBB leak results were in good comparison with CFD results.

Finally the calculated leakage value was compared with the existing limit value for this particular vent pipe.

Fig. 1. Steam generator collector’s vent pipe (with leakage area detail).
REFERENCES


COMBINATION CRITERION FOR MULTIPLE LAMINAR FLAWS IN STEEL COMPONENTS

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**KEY WORDS:** combination criterion, multiple laminar flaws, allowable area of laminar flaw

A laminar flaw is a planar subsurface flaw parallel to the rolling direction of the plate, where the applied stress is typically parallel to the rolling direction. The laminar flaw oriented within 10 degree of a plane parallel to the component surface is defined as a laminar flaw, in accordance with the ASME (American Society of Mechanical Engineers) Code Section XI \cite{1}.

Multiple discrete laminar flaws have been detected in steel structures \cite{2}. The ASME Code provides combination criterion for multiple laminar flaws. If there are two or more laminations, these laminations are projected to a single plane and the separation distance of the projected laminations is less than or equal to 25.4 mm, the laminations shall be combined into a single large laminar flaw in the assessment, as shown in Fig. 1. The combination criterion was established based on non-destructive examination capabilities long years ago \cite{3}. This methodology did not consider offset distance of the laminations and existence of a large number of laminar flaws.

When multiple discrete laminar flaws are close to each other, interaction between the flaws may occur and these flaws shall be combined to a single laminar flaw for assessment. Stress intensity factor interactions for inclined laminar flaws were analysed. Based on the interaction, new combination criteria were developed as follows.

\begin{align*}
S_1 &\leq 0.37 \min \left( \max(w_1, \ell_1), \max(w_2, \ell_2) \right), \\
S_2 &\leq 0.37 \min \left( \max(w_1, \ell_1), \max(w_2, \ell_2) \right), \\
H &\leq 0.17 \min \left( \max(w_1, \ell_1), \max(w_2, \ell_2) \right),
\end{align*}

where \( S_1, S_2, H \) are the distances between laminar flaws and \( w_i \) and \( \ell_i \) (\( i = 1, 2, 3 \)) are the dimensions of laminar flaws as shown in Fig. 2. If all three equations above are met, the multiple laminar flaws shall be combined into a single flaw, where the single laminar flaw sizes become \( w \) and \( \ell \).

The ASME Code provides tables of allowable areas of laminar flaws for vessels and pipes. Figure 3 shows the allowable diameter for vessels converted from the areas in the table. The allowable laminar diameter increases with increase of the wall thickness. Czech Code also provides allowable laminar flaws for nuclear steam generators \cite{4}.

\begin{figure}
\centering
\includegraphics[width=\textwidth]{fig1.png}
\caption{Current combination rule of ASME Code.}
\end{figure}
If a detected laminar flaw is less than the allowable area of the laminar flaw, repair or replacement is not necessary for vessels and piping items. Therefore, it is necessary to give an area for the combined laminar flaw. The bounded square or rectangle of the combined laminar flaw is defined as elliptical in shape. Then, the area of the combined laminar flaw is deemed to be 0.75 times the area of $w \times \ell$.

![Fig. 2. Proposed combination rule of ASME Code.](image1)

![Fig. 3. Allowable laminar flaw diameter.](image2)

It is concluded that the multiple laminar flaws shall be combined if the all Eqs. (1) to (3) are satisfied, and the area of the combined laminar flaw is 0.75 times the bounded square or rectangle.

REFERENCES


USE OF THE FINITE STRAIN THEORY TO DETERMINE FATIGUE PROPERTIES

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KEY WORDS: strain, stress, fatigue life

This article presents different types of determining strains. Absolute strain and relative strain also known as engineering strain are described. Moreover, the article gives more consideration to logarithmic strain or relative logarithmic strain expressed in the general form as

$$\delta = \int_{l_0}^{l_1} \frac{1}{l} dl.$$  \hspace{1cm} (1)

The majority of analyses in static relate to so-called small strains. The large strain theory, i.e. the finite strain theory, can be also applied (1). This theory is useful in case of significant plastic strains. In addition, it seems that this theory can be applied to the fatigue of materials with a very small number of cycles and significant plastic strains. So far, the small strain theory has been generally used in the fatigue analysis. The elaboration [1] is the only exception in which the possibility of using this theory in relation to fatigue was mentioned; however, there was no need to apply it in this article due to the fact that plastic strains were not high enough to make it necessary to use it.

It was shown that for certain materials, in case of a very small number of cycles in the fatigue analysis for uniaxial tension-compression, the possibility of using relative logarithmic strains should be considered (1) instead of a traditionally understood normal strain determined on the basis of the theory of small strains, also known as "engineering strains". Here, one may use the theory well known in static for finite strains, i.e. large strains internationally known under the term of „true strain” and express it as the strain amplitude in the form of

$$\varepsilon_{at} = \ln(1 + \varepsilon_a).$$  \hspace{1cm} (2)

In this case, true stress should be also defined in a different way. In case of assuming constant volume of the material, i.e. for large strains, the following is obtained after transformations

$$\sigma_{at} = \sigma_a(1 + \varepsilon_a).$$  \hspace{1cm} (3)

In order to determine fatigue properties, 9 materials found in the literature were used [2]. The author chose these materials for which tests were carried out at high amplitude of strains, i.e. significant plastic strains. On the basis of data from periodic tests taken from the literature, basic fatigue properties were determined. These properties were determined on the basis of strains $\varepsilon_{at}$ and stresses $\sigma_{at}$ evaluated using the theory of finite strains determined based on the theory of small strains (2) and (3). The basic fatigue properties in the scope of a small number of cycles are presented by the model of Manson-Coffine-Basquin linking the total strain amplitude with the number of cycles. This is the most popular and the most commonly used approach. The original properties were evaluated by recording the strain amplitude $\varepsilon_a$, the stress amplitude $\sigma_a$ and the number of destruction cycles $N_f$. Thus, on the basis of the elastic strain amplitude $\varepsilon_a$ and the number of destruction cycles $N_f$, these properties are determined i.e.

$$\varepsilon_a = \frac{\sigma_f}{E} \left(2N_f\right)^{1/2} + \varepsilon'_f \left(2N_f\right)^{1/2},$$  \hspace{1cm} (4)
where: $\varepsilon_f$ - fatigue plastic strain coefficient, $c$ - fatigue plastic strain exponent, $\sigma_f$ - fatigue strength coefficient for tension-compression, $b$ - fatigue strength exponent.

Similarly, in case of large strains

$$\varepsilon_{af} = \frac{\sigma_{af}}{E} \left(2N_f\right)^{\varepsilon_f} + \varepsilon_f \left(2N_f\right)^{c},$$

where: $\sigma_f$ - fatigue strength coefficient for tension-compression for finite strains, $b''$ - fatigue strength exponent for finite strains, $\varepsilon_f$ - fatigue plastic strain coefficient for finite strains, $c''$ - fatigue plastic strain exponent for finite strains.

The relationship between the strain and stress amplitudes can be described using the Ramberg-Osgood equation in the form of

$$\varepsilon_a = \frac{\sigma_a}{E} + \left(\frac{\sigma_a}{K'}\right)^{1/n'},$$

where: $K'$ - cyclic strain hardening coefficient, $n'$ - cyclic hardening exponent.

The above-mentioned model of Ramberg-Osgood for determining fatigue properties for (1) for large strains can be presented as

$$\varepsilon_{af} = \frac{\sigma_{af}}{E} + \left(\frac{\sigma_{af}}{K''}\right)^{1/n''},$$

where: $K''$ - cyclic strain hardening coefficient for finite strains, $n''$ - cyclic hardening exponent for finite strains.

Figure 1 presents the comparison of fatigue properties of aluminium alloy 1100Al determined using the models of Ramberg-Osgood and Manson-Coffine-Basquin.

![Figure 1: Cyclic properties according to the small and finite strain theories for aluminium 1011.](image)

On the basis of selected data from fatigue tests taken from the literature [2], the differences occurring during the determination of fatigue properties using the small strain theory and the finite (large) strain theory, taking the constants from the models formulated by Manson-Coffine-Basquin and Ramberg-Osgood into consideration, were presented. This theory may also be applied in relation to materials which are characterised by a relatively high plasticity.

REFERENCES


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KEY WORDS: fatigue life, critical plane

The paper presents a new model for estimating fatigue life and the analysis of the influence of the critical plane orientation on such an estimation. The algorithm for estimating fatigue life uses a criterion of maximum shear or normal stress in the fatigue damage plane.

General form of the equivalent stress [1] according to the proposed criteria can be written as

\[ \sigma_{eq}(t) = B \tau_p(t) + K \sigma_n(t), \]

where: \( K \) and \( B \) constants used for selection of specific criterion form.

For the analysis authors used criterion (1) where

\[ B = \frac{B_2 \sin(90^\circ + 2\beta)}{\sin 2\beta \sin(90^\circ + 2\beta)} \]

\[ K = 2 + B \sin 2\beta = 2 - \frac{\sigma_{eq}}{\tau_{eq}}, \]

where: \( B_2 \) is ratio of normal and shear stresses

The critical angle of the plane is increased by the angle proposed by Carpinteri [2]

\[ \beta = \frac{3}{2} \left[ 1 - \left( \frac{1}{B_2} \right)^2 \right] 45^\circ, \]

relative to the angle defined by the maximum normal stress.

A modified angle with the \( \beta \) angle value was adopted by the authors as an inclination angle of critical plane orientation and it depends on the type of material, relative to the angle defined by the maximum normal stresses. The paper involves calculations carried out for several construction materials, based on the results of scatters defined as root-mean-square value.

\[ E = \sqrt{\frac{\sum \log^2 N_{exp}}{n} \frac{N_{cal}}{N_{cal}}}, T = 10^E. \]

For each of the analyzed materials, values of \( \beta \) angle were selected where the values of scatter were the smallest (Fig. 1). On the basis of these calculations, a model was defined which allows for a new dependence between normal stresses from bending and shear stresses from torsion. The authors propose a new expression for \( \beta \) (Fig. 2):

\[ \beta = \arctan \left( \sqrt{\frac{1 + \sqrt{3}}{2} - \frac{\sigma_n(N_{\beta})}{\tau_p(N_{\beta})}} \right), \]

\[ \arctan \left( \frac{1}{4} \right) = 22.5^\circ. \]
Fig. 1. Relationship between scatter values $T(6)$ and the angle $\beta$ for CuZn40Pb2.

Fig. 2. The dependence of critical plane angle from normal to shear stresses ratio for the selected materials.

REFERENCES


DUCTILE FAILURE SIMULATION OF SMALL PUNCH TEST USING STRESS-MODIFIED FRACTURE STRAIN ENERGY MODEL

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KEY WORDS: ductile failure simulation, small punch test, finite element damage analysis

Often fracture toughness testing specimens cannot satisfy size requirement due to unavailability of materials. In this respect, semi – destructive testing methods such as small punch test would be practice. The small punch test has been applied to estimate fracture toughness using energy method [1-2]. In this paper, ductile failure of fracture toughness test and small punch test are simulated using FE analysis. In FE analysis, stress – modified fracture strain energy model was used as FE damage model. To validate the simulation results, analysis results are compared with mechanical test data for three different materials.

Three materials were considered in experiments, commonly used as structural materials in pressurized water reactor nuclear power plants. To define material properties, standard smooth bar tensile test were performed. To investigate the effect of tri-axial stress states on fracture characteristics, notched bar tensile tests were also carried out. C(T) specimens were used as fracture toughness tests. In addition to tensile and fracture toughness tests, small punch test were performed. As specimens, disc-type smooth and notched specimens were used. More detailed information for experiments is explained in Ref [3].

Fracture strain energy means area under the stress–strain curve to fracture initiation point. The multi-axial fracture strain energy, \(W_f\), is assumed to be given in terms of stress triaxiality.

\[
W_f = A \exp\left(-C \frac{\sigma_m}{\sigma_c}\right) + B, \tag{1}
\]

where \(A\), \(B\) and \(C\) are material constants which can be founded by notched bar test FE analysis. Once material constants in Eq. (1) are determined for given materials, incremental damage of material due to plastic deformation, \(\Delta \omega\), is calculated using Eq. (2).

\[
\Delta \omega = \frac{\Delta W_p}{W_f}, \tag{2}
\]

where is equivalent plastic strain energy increment, calculated from FE analysis. When the accumulated damage became unity, \(\omega = \Sigma \Delta \omega = \omega_c = 1\), crack growth is simulated by reducing all stress component at the gauss point sharply to a small plateau value. To determine material constants, FE analysis for smooth and notched tensile tests were performed. The material constants \(A\), \(B\) and \(C\) are found to be

\[
A=2474; \quad B=1.48; \quad C=80 \text{ for SA508 Gr.3}, \tag{3}
\]

this equation is shown in Fig. 1.

3D FE damage analysis were performed to estimate C(T) tests. To determine proper element size for FE damage analysis, several FE meshes were prepared. Sensitivity analysis results suggest that proper element size is 0.05 mm for SA508 Gr.3. Simulated results using proper element size are compared with experiment data in Fig. 2.
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Fig. 1. Stress –modified fracture strain energy model.

Fig. 2. Comparison of experimental C(T) results with simulated ones using FE damage analysis.

Using the damage model, SP and NSP tests are also simulated. FE analysis results show that SP test simulation results are not dependent on the element size. In contrast, NSP test simulation results depend on the element size. By comparing with experimental data, proper element size of NSP tests was found to be 0.025 mm for SA508 Gr.3. The simulation results compared with experimental data are shown in Fig. 3.

Fig. 3. Comparison of experimental SP and NSP test results with simulated ones.

REFERENCES
ESTIMATION OF FRACTURE TOUGHNESS FROM SMALL PUNCH TEST USING STRESS-MODIFIED FRACTURE STRAIN ENERGY MODEL

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KEY WORDS: small punch test, finite element damage analysis, fracture toughness

Often fracture toughness testing specimens cannot satisfy size requirement due to unavailability of materials. In this respect, semi – destructive testing methods such as small punch test would be practical. The small punch test has been applied to estimate fracture toughness using energy method [1-3]. In this paper, fracture toughness has been estimated from SP test data using FE damage analysis. In FE analysis, stress–modified fracture strain energy model was used.

Three materials, low alloy steel SA508 Gr.3, austenitic stainless steel TP316L and cast austenitic stainless steel CF8M, were considered in experiments, commonly used as structural materials in pressurized water reactor nuclear power plants. To tensile and fracture toughness test, small punch (SP) test were performed. As specimens, conventional disc – type smooth specimens are used. All tests are performed at room temperature. More detail information for experiments is explained in Ref [4].

Tensile properties are obtained from SP test using iterative FE analysis. The material property is assumed to following elastic, power-law plastic behaviour:

\[
\begin{align*}
\sigma &= E \varepsilon & \text{for } \varepsilon \leq \varepsilon_y \\
\sigma &= \sigma_y + K (\varepsilon_p) \quad \text{for } \varepsilon \leq \varepsilon_y,
\end{align*}
\]

where \(E\) is elastic modulus; \(\varepsilon, \varepsilon_p\) and \(\varepsilon_y\) mean total strain, plastic strain and yield strain; and \(K\) and \(n\) are hardening parameter of materials. Eq. (1) has 3 material parameters to be defined; \(\varepsilon_y, n\) and \(K\). Determined material parameters from iterative FE analysis are \(\varepsilon_y=450\text{MPa}, n=0.26\) and \(K=430\), for SA508 Gr.3. The SP test simulation result is shown in Fig. 1. Fracture toughness of materials are also estimated from SP test data using modified energy method [2]. Estimated fracture toughness is \(J_{IC} = 283.91\text{ kN/m}\) for SA508 Gr.3.

Fracture strain energy means area under the stress – strain curve to fracture initiation point. The multi – axial fracture strain energy, \(W_f\), is assumed to be given in terms of stress triaxiality.

\[
W_f = A \exp \left( -1.5(n+1) \frac{\sigma_m}{\sigma_c} \right),
\]

Fig. 1. Comparison of experimental data with FE result for load – deflection curve.
where $A$ is material constant which can be founded by notched bar test FE analysis. Once material constant in Eq. (2) are determined given materials, incremental damage of material due to plastic deformation, $\Delta \omega$, is calculated using Eq. (2)

$$\Delta \omega = \frac{\Delta w_p}{w_f},$$

where is equivalent plastic strain energy increment, calculated from FE analysis. When the accumulated damage became unity, $\omega = \sum \Delta \omega = \omega_c = 1$ crack growth is simulated by reducing all stress component at the gauss point sharply to a small plateau value.

To determine material constants, FE analysis for smooth and notched tensile tests were performed. The material constant $A$ is found to be

$$A = 3340.75 \text{ for SA508 Gr.3},$$

this equation is shown in Fig. 2.

To estimate $C(T)$ tests, 2D plane strain damage analysis were performed. To determine proper element size for FE damage analysis, several FE meshes were prepared. Sensitivity analysis results suggest that proper element size is 0.13 mm for SA508 Gr.3. Simulated result using proper element size is compared with experiment data in Fig. 3.

Fig. 2. Stress-modified fracture strain energy for SA508 Gr.3.

Fig. 3. Comparison of experimental $C(T)$ results with simulated ones using FE analysis.

REFERENCES


MICRO-CRACK GENERATION IN CYCLICALLY DEFORMED Ti-Fe-O ALLOY AT LOW TEMPERATURE

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KEY WORDS: fatigue, subsurface crack, low temperature, electron backscatter diffraction

In general, fatigue crack initiation is understood to occur on the specimen surface due to irreversible process of extrusion and intrusion through slip deformation. In titanium alloys, however, subsurface crack initiation not associated with pre-existing defects has been reported in high-cycle regime and at lower temperature. The subsurface crack initiation sites commonly appear crystallographic transgranular facet or facets in both near α and α+β titanium alloy [1]. The facet plane was mostly formed on or near basal plane and some models of its formation mechanism have been proposed. Although various dominant factors such as slip off, high normal stress on facet plane or the combination of shear stress and tensile stress normal to the basal (facet) plane have been pointed out in the models, dislocations pile-ups induce a local stress concentration near the grain boundary and are responsible to the facet formation in the neighbour grain [2],[3]. Dislocation movement is fairly planar and dislocation arrays on {01-10}<11-20> are piled-up in the vicinity of grain boundaries. As increase of oxygen content, deformation twin and <c+a> slip are strongly suppressed, and basal slip becomes active. However, no evidences directly show microcrack initiated at localized deformation regime. In the present study, the subsurface crack generation and its growth has been clarified by scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD) analysis.

Two kinds of Ti-Fe-O materials in the annealed condition were used. One was cross-rolled (CR) material, and the other was grooved-rolled and cold-swaged (CS) material. They also exhibited the subsurface fatigue crack initiation and the facet was detected as {0001} crack. In addition, {01-10} facet was also detected in CS material.

In the CR material, subcracks were detected in α grains, β grains and α-β interfaces. Since β phase is softer than α phase, the interface especially at the triple point of α-β-β grains becomes a site for strain concentration. It will assist a microcrack generate at the interface or β grain boundary. The subcracks in α grains were divided into two types.

One is that their origins associated with microcrack or void in the neighboring β grain and showed a near basal orientation crack plane. The microcrack generated at β grain boundary and the local stress concentration induced a very localized basal slip in α grain. The slipped plane becomes softer than non-slipped area due to an increase in mobile dislocations. The combination of the localized basal slip and higher tensile stress may be responsible for a microcrack generation on the basal plane.

The other is that subcracks formed at (0001) twist boundary of about 15 to 30 degree. There would be a higher shear stress inconformity along with the twist boundary. In addition, strain concentration generated at the twist boundary due to the planar slip deformation on a prismatic plane in one of the two grains may induce a normal stress on the twist boundary. Under the combination of shear stress and the normal stress on the twist boundary, a microcrack generated and grew along the boundary. In the initial stage of crack generation, the microcracks were generated almost parallel to each other and coalesce to a crack.
In the CS material, many subcracks were detected near the fracture surface. The most of them were in β grains and a few of them were in α grains and α-β interfaces. The microcracks in β grains were initiated at the β boundary of about from 40 to 60 degree. Since the α grains showed a (01-10) fiber texture, basal slip system was less active rather than prismatic one. The microcracks in β grains may give a site for the very localized prismatic slip in the neighbor α grain. In the present material, deformation twins were detected even though higher oxygen content.

REFERENCES


CRACK CLOSURE AT FATIGUE CRACK GROWTH UNDER NEGATIVE R RATIO

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KEY WORDS: fatigue crack growth, negative R ratio, crack closure, stress intensity factor range

Fatigue crack growth rate \( \frac{da}{dN} \) for ferritic steel in air was provided by the ASME (American Society of Mechanical Engineers) Code Section XI [1] as follows;

\[
\frac{da}{dN} = 1.317 \times 10^{-8} \Delta K_{eff}^{3.07},
\]

where \( \Delta K_{eff} \) is the effective stress intensity factor range which is given by \( \Delta K_{eff} = U \Delta K \). Crack closure \( U \) is obtained by \( U = (P_{max} - P_{op})/(P_{max} - P_{min}) \), where \( P_{max} \) is the maximum load, \( P_{op} \) crack opening load and \( P_{min} \) minimum load. From the survey of the crack closure \( U \) as shown in Fig. 1, the ASME had decided \( U = 1.0 \) for \( R \geq 0 \) and \( U = 0.667/(1-R) \) for \( -2 \leq R < 0 \) [2], where \( R \) is the stress ratio. Crack closure \( U \) decreases with decreasing \( R \) ratio. The fatigue crack growth rate expressed by Eq. (1) is re-written as;

\[
\frac{da}{dN} = 3.78 \times 10^{-9} S(\Delta K_1)^{3.07},
\]

where \( S \) is the scaling parameter and \( S = 1.0 \) for \( R < 0 \). The stress intensity factor range \( \Delta K_1 \) for \( R < 0 \) is categorized as applied load given by;

\[
\Delta K_1 = K_{max} \quad \text{for} \quad K_{max} - K_{min} \leq 1.12\sigma_f \sqrt{a},
\]

\[
\Delta K_1 = K_{max} - K_{min} \quad \text{for} \quad K_{max} - K_{min} > 1.12\sigma_f \sqrt{a},
\]

where \( \sigma_f \) is the flow stress and \( a \) the crack length, \( K_{max} \) maximum stress intensity factor and \( K_{min} \) minimum stress intensity factor. This means that the crack growth rate calculated by \( K_{max} - K_{min} \) is higher than that by \( K_{max} \). This is because crack closure \( U \) increase with increasing applied stress.

Figure 1 shows the relationship of \( U \) and \( R \) ratio. When \( R \) is positive, \( U \) is around 0.7 to 1.0 and crack almost opens at \( K_{min} \). When \( R \) is negative \( U \) decreases with \( R \), and crack closes at \( K_{min} \) side. This implies that compression stress does not contribute crack growth rate. However, when the applied stress range increases, \( U \) increases under the same \( R \) condition. Relationship of \( U \) and applied stress level for \( R = -1.0 \) is plotted as closed circles, as shown in Fig. 1. When the stress levels of \( \sigma_{max}/\sigma_f \) were 0.1, 0.3, 0.5 and 0.7 under \( R = -1.0 \), \( U \) were obtained as 0.33, 0.36, 0.42 and 0.48, respectively [3].

Fig. 1. Comparison of proposed crack closure \( U \).
The negative $R$ ratio and $U$ are discussed at the ASME Code Section XI Working Group, commented by US NRC (Nuclear Regulatory Commission). NRC suggested $0.8 \times 1.12 (=0.9)$, instead of $1.12$ in Eqs. (3) and (4). If using $K_{\text{max}}-K_{\text{min}} > 0.8 \times 1.12 \sigma_f \sqrt{(\pi a)}$, fatigue crack growth rate $da/dN$ increases at the low level of $\Delta K_i$.

Fatigue crack growth experiment was performed for JIS STPT 410 carbon steel plate specimen at ambient temperature in air environment. Yield stress and tensile strength of STPT 410 employed are 280 and 495 MPa, respectively. Then, the flow stress is $\sigma_f = 387.5$ MPa. Width and thickness of the centre notched specimen are 30 and 8 mm, respectively. Initial through wall notch length is 10 mm. Applied loads started at 10 kN with $R = -1.0$, increasing 2.5 kN stepwise load for each 500 cycles. The opening load $P_{\text{op}}$ was calculated from load-displacement curve, where displacement was measured by clip gauge.

Figure 2 shows the relationship between crack closure $U$ and stress intensity factor range $\Delta K_i$. Crack closure $U$ increases from 0.25 to 0.75 with increasing $\Delta K_i$. $\Delta K_i$ at 1.12$\sigma_f \sqrt{(\pi a)}$, is about 66 MPa$\sqrt{m}$. In accordance with the ASME Code, $\Delta K_i$ is used Eq. (4) for $\Delta K_i > 66$ MPa$\sqrt{m}$ and $\Delta K_i$ is used Eq. (3) for $\Delta K_i < 66$ MPa$\sqrt{m}$. However, the $U$ is not significant changes around 66 MPa$\sqrt{m}$, even when $\Delta K_i$ at $0.8 \times 1.12 \sigma_f \sqrt{(\pi a)} = 48$ MPa$\sqrt{m}$, as shown in Fig. 2. The transient of the $U$ is shown to be $\Delta K_i = 36$ MPa$\sqrt{m}$ which corresponds to $\Delta K_i = 0.54 \times 1.12 \sigma_f \sqrt{(\pi a)}$.

![Fig. 2. Crack closure $U$ and stress intensity factor $\Delta K_i$ for carbon steel plate under $R = -1.0$.](image)

It is suggested that current code of $1.12 \sigma_f \sqrt{(\pi a)}$ might be less conservative and 1.12 shall be reduced to be 0.60 ($= 0.54 \times 1.12$).

REFERENCES
NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS
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FATIGUE CRACK GROWTH BEHAVIOUR FOR ADJACENT TWO SURFACE FLAWS IN ACCORDANCE WITH COMBINATION RULES

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KEY WORDS: fatigue crack growth, combination rule, two surface flaws, interaction of two flaws

If multiple flaws are found in a structural component, fitness-for-service (FFS) codes such as ASME Code Section XI [1], BS 7910 [2], FKM Guideline [3] provide flaw combination rules. However, the specific criteria are different among FFS codes. Therefore, it is easily expected that the profiles of fatigue flaws might be different during fatigue flaw growths.

The ASME Code provides combination rule for adjacent two surface flaws as follows:

\[
S = \begin{cases} 
0.5 \times \max(a_1, a_2) & \text{for fracture} \\
0 & \text{for fatigue and SCC}
\end{cases}
\] (1)

The BS 7910 has a similar combination rule as given by:

\[
S = \begin{cases} 
\min(\ell_1, \ell_2) & \text{for } \frac{a_1}{\ell_1} \text{ or } \frac{a_2}{\ell_2} > 0.5 \\
0.5 \times \max(a_1, a_2) & \text{for } \frac{a_1}{\ell_1} \text{ and } \frac{a_2}{\ell_2} \leq 0.5
\end{cases}
\] (2)

The combination rule provided by the FKM is expressed as:

\[
S = \min(\ell_1, \ell_2),
\] (3)

where \( S \) is the distance between two flaws, \( a_1 \) and \( a_2 \) are flaw depths, \( \ell_1 \) and \( \ell_2 \) are flaw lengths, as illustrated in Fig. 1. After combination of the two flaws, flaw depth and length are \( a = \max(a_1, a_2) \) and \( \ell = \ell_1 + S + \ell_2 \), respectively, for all FFS codes.

In order to obtain fatigue crack growth behaviour for two similar surface flaws in a stainless steel pipe (165.2 mm diameter Schedule 80, wall thickness \( t = 11.0 \) mm), the initial flaw depth is \( a = 1.74 \) mm, length is \( \ell = 34.8 \) mm, which corresponds to the allowable flaw in Acceptance Standards defined by the ASME Code. Initial distance between two flaws is \( S_0 = 1 \) mm. The pipe is subjected to a uniform cyclic tensile stress with the maximum stress \( \sigma_{\text{max}} = 123 \) MPa which corresponds to the allowable design stress for the stainless steel, and the minimum stress \( \sigma_{\text{min}} = 0 \) MPa. Stress intensity factor and equation of fatigue crack growth rate provided by the ASME Code are used for all flaw growth calculations. This is because it makes clear the differences of flaw growth behaviours among FFS codes based on only combination rules. The fatigue flaw growth calculations were terminated when the single combined flaw depth reaches 75% of the nominal wall thickness \( t \), i.e., \( a/t = 0.75 \).
In calculating fatigue crack growths for two flaws, code users perform calculations for a single flaw independently, and always check the geometries of the flaws of depths, lengths and the distance of two flaws. Both flaws are not considered to have interaction by the neighboring flaw. On the other hand, fatigue flaw growth calculations by the X-FEM (extended finite element method) were conducted by the same conditions. The calculations by X-FEM were conducted considering the two surface flaws concurrently, taking into account the actual interaction of two neighboring flaws.

Figure 2 shows the results of circumferential flaw angle, where \( 2\theta \) is an angle for each flaw before combined, and it is a total angle after combination. The curve of the FKM shows that two flaws are already combined before calculations. The difference of the combination for the BS 7910 and the ASME is based on \( S = 0.5a_1 \) (or \( 0.5a_2 \)) and \( S = 0 \). Flaw growth angle by the ASME is close to that by X-FEM. After combination, the flaw angles for all cases converge to the same angle, that is \( 20/2\pi = 0.16 \).

However, growing fatigue flaw depths by these three FFS codes are different. Figure 3 shows the relationship between flaw depth and number of cycles. FKM gives short number of cycles, and flaw depth by the ASME is almost the same with that by X-FEM with interaction effect. It can be said that the flaw profile of fatigue growth calculated by the ASME flaw combination rule is appropriate.

**REFERENCES**


CRACK GROWTH RATE OF R7T STEEL UNDER UNIAXIAL LOADING

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KEY WORDS: fatigue, ferritic-pearlitic steel, R7T steel, crack growth rate, fatigue crack initiation

Fatigue is a phenomenon affecting most structural components during their operational life. The fatigue process has a direct effect on the lifespan of components, so the study of this degradation process is extremely important. In cases where a component is loaded under its design yield strength, it is very difficult to detect a crack in the early stages and thereby to avoid catastrophic failure. This paper investigates fatigue crack initiation and the fatigue crack growth rate in ferritic-pearlitic steel. It is primarily the microstructure of the material that plays an important role in the initial stages of fatigue cracks. Fatigue cracks in areas close to the threshold stress intensity factor range $\Delta K_{\text{th}}$ are strongly influenced by the microstructural characteristics of the material, such as grain size, interlamellar distance, etc., together by the mean value of the applied stress. [1].

The material employed in this paper is hypoeutectoid ferritic-pearlitic steel R7T (Fig. 1), which is commonly used for the manufacture of railway wheels. This is because the structure of railway wheels is heavily exposed to the effects of fatigue and environmental conditions. The material is a highly appropriate subject for study.

In the case of this steel, the microstructure consists of lamellar pearlite and ferrite netting around grain boundaries, where a key role is played by the size of grains and pearlitic colonies as well as interlamellar spacing. In the area subject to Paris’ law, the resistance to fatigue crack propagation can be influenced just by controlling the microstructure. Paris’ law, gives the relations between the amplitude of the stress intensity factor and subcritical crack growth rate. It can be expressed using the following equation [2]:

$$\frac{da}{dN} = C \cdot \Delta K^m,$$

where $N$ is the number of cycles, $C$, $m$ are materials constants, and $\Delta K$ is the amplitude of the stress intensity factor at the crack tip.

The experimental part of the paper is mainly focused on the influence of stress amplitude on the crack growth rate from the area of initiation to a stable stage of fatigue crack growth. The measurement was based on cyclic loading, followed by subsequent metallographic observation and evaluation of the crack length. Figure 2 shows a plot of the relation between the crack length and the crack growth rate combined with a depiction of the final shape of the main crack in the final stages before final fracture.
The detailed effects of microstructure on crack growth in the area of initiation and the area subject to Paris’ law are not fully described in the current literature. The literature describes the fatigue characteristics and failure mechanisms for many different materials, which are in many fundamental ways similar to the steel investigated here, but the detailed behaviour of fatigue cracking differs from material to material. [3, 4, 5]. Despite the growing body of theoretical and experimental results, there still remain many unsolved problems for new and existing materials which require innovation in order to improve their properties. [6]. Therefore, it is important to continue to perform experimental measurements and to carry out studies of fatigue failure.

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REFERENCES


RELATION BETWEEN CHARPY IMPACT VALUE AND VICKERS HARDNESS OF REPEATEDLY QUENCHED HIGH CARBON HIGH CHROMIUM STEEL (SUJ2)

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KEY WORDS: repeated quenching, charpy impact test, SUJ2, high carbon high chromium steel

It is well known that the relation between toughness and hardness is important for the materials which are used for mechanical components. In particular, we should be careful when quenching steels because uncontrollable quenching sometimes makes the material brittle.

Our research group has investigated the repeated quenching of high strength steels [1-2]. This method was developed for producing ultrafine grained steel by Grange, and he investigated the relation between repeated heating and material strength. We applied this method to high carbon high chromium steel and performed the following tests: rolling contact fatigue tests, fatigue tests under reciprocating motion, and rotating bending tests. We have obtained two important results from these investigations over the past five years: (1) repeated quenching refines prior austenite grains (PAG); (2) retained austenite increased with quenching times. Generally, the grain size and retained austenite were strongly related to the toughness. Therefore we are continuing to research the effect of repeated quenching on material toughness.

In this study, we performed Charpy impact tests and Vickers hardness tests on the repeatedly quenched steel specimens (SUJ2). We focused on the relation between toughness and hardness and also the effect of PAG size on the fracture surfaces.

We prepared JIS-SUJ2 material and its chemical composition in weight percent was: 1.00% C, 0.24% Si, 0.37% Mn, 0.014% P, 0.007% S, 0.12% Cu, 0.07% Ni, 1.343% Cr, 0.04% Mo. Before the heat treatment, the microstructure was ferrite and spheroidized cementite. The dimensions of the specimen were 55 × 10 × 10 mm and a notch was machined in the center of the specimen (The depth and radius of the notch was 2 mm). We applied three types of quenching patterns to the specimens. The specimens were quenched once, twice and three times, and these are referred to as Q1T1, Q2T1 and Q3T1.

Figure 1 shows the Charpy test machine. The hammer weight was 25.53 kg and the radius of the pendulum was 0.659 m. We set the starting angle of the pendulum at 30 degrees. The angular resolution was 0.5 degrees which is 8 kJ/m² on the Charpy impact value scale.

Figure 2 shows the relation between the Vickers hardness and Charpy impact value. In these tests, the Charpy impact values ranged from 50 to 82 kJ/m² and Vickers hardness values ranged from 750 to 800 HV. A detailed discussion will be given in the conference presentation.
Acknowledgement: This study was partially supported by the Iron and Steel Institute of Japan (This is the 24th time financial support has been given for iron research).

REFERENCES


Fiber-reinforced self-healing ceramics [1] exhibit unique damage tolerance, thereby, anticipated to be next generation structural material, such as turbine blades in jet engine. Its damage tolerance originates from the competition between crack propagation and crack re-bonding due to self-healing. Self-healing function is generated from the high temperature oxidation of healing agent, which is located at the interface between ceramic fiber bundle and matrix, thus, influenced strongly by service temperature. In order to actualize the components made of the fiber-reinforced ceramics, the fracture criteria with the competition is necessary.

In the present study, the creep fracture behaviour of the fiber-reinforced self-healing ceramics was investigated at high temperatures. From the time change in displacement under tensile stress, the competition behaviour between crack propagation and crack re-bonding was discussed.

The used sample is typical fiber-reinforced ceramics, which consists of alumina fiber bundle, alumina matrix and SiC interlayer as healing agent. Figure 1 shows the creep curves of the sample at 1000°C. As shown in figure, when tensile stress is 150 MPa, the creep rate cannot reach to 0, thus, the specimen fractured after 57 h. On the other hand, under tensile stress of 137 MPa the creep rate reached to 0 after 80 h and the specimen could survive for more than 300 h. The specimen has cracking strength at 1000°C of 47 MPa and final fracture strength of 199 MPa, thus, the creep behaviour was found to include the competition between crack propagation and crack re-bonding due to self-healing. Moreover, creep strength of 137 MPa is quite high, compared to the final fracture strength, therefore, it was found that self-healing function affects the creep strength of self-healing materials strongly.

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REFERENCES

RELIABILITY EVALUATION OF CAR POWER MODULE USING ELECTRICAL- THERMAL-STRUCTURAL COUPLED ANALYSIS BASED ON FIELD DRIVING DATA

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KEY WORDS: power module, junction temperature, miner’s law, fatigue life

Power module is the electronic components to perform the power conversion. Power module is a significant impact on the operation of the equipment and its reliability evaluation is very important. Especially, intermittent electric current flowing in the power module causes repeating the heat generation and cooling. It produces a large thermal stress and cause thermal fatigue failure in the solder part.

Car makers usually use the power cycle test to measure the thermal fatigue life. Power cycle test is a test to flow a constant current at a fixed period and then cool down the module to low temperature repeatedly. But, this test method is difficult to simulate the field conditions. Because, in the actual use environment, the irregular current flows in an irregular cycle. In this research, analysis in the conditions near real usage environment is conducted. It aims at establishment of the reliability assessment method of the power module near a real operating condition.

Analysis model is shown in Fig. 1. This model is referring to IGBT modules which are often used in automobiles. Table 1 is electrical load conditions. An electrical-thermal-structural coupled analysis were carried out to simulate the fatigue life of the power module. The conditions shown in Table 1 are used to investigate the effect of current loading on the reliability. Conditions 1-4, which the electrical load conditions are different, are set to get the same maximum temperature of the device at 123°C.

Figures 2 show the strain range in solder joint under Si-chip, where the plastic and creep behavior of the solder material were considered in the thermal-structural coupled analysis. The result show that the maximum difference between each condition is lower than 4.9%. That means that if the \( T_{\text{max}} \) are the same, these fatigue life can be considered to be identical.

Based on the analysis results and the result of driving test, driving-model is created and calculated solder initial fatigue life. Then it was compared when considering high load only and when considering low load and high load. It was shown that the impact of current loads lower than 50[A] on the fatigue reliability is small enough to be neglected.

<table>
<thead>
<tr>
<th>Table 1 Current profile.</th>
<th>Load Current[A](On Time[s])</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>200(2.0)</td>
</tr>
<tr>
<td>2</td>
<td>120(1.0)+208(1.0)</td>
</tr>
<tr>
<td>3</td>
<td>210(1.0)+160(1.0)</td>
</tr>
<tr>
<td>4</td>
<td>210(1.0)+120(2.0)</td>
</tr>
</tbody>
</table>
Figure 3 shows a sample of field driving condition [1]. Based upon the motor information, the current profile of the general driving condition as shown Fig. 4 and the profile of a hard driving condition as shown in Fig. 5.

The relation between the fatigue life and inelastic strain range can be given by the Manson-Coffin’s Law, as shown follows [2].

\[
N = 1328 \times \left( \frac{\Delta \varepsilon}{0.01} \right)^{-1.43},
\]

where \(\Delta \varepsilon\) means the inelastic equivalent strain.

Miner’s Law, as shown follows was applied to estimate the conditions with irregular loads as shown in Fig. 4 and Fig. 5.

\[
D = \sum \left( \frac{n_i}{N_i} \right),
\]

where \(n_i\) means actual number of cycles, \(N_i\) means fatigue lives of cycles.

Consequently, when considering high load only, initial crack occurs in the solder, compared to 8.63 years.

REFERENCES


NUMERICAL SIMULATION OF FRACTURE TOUGHNESS TEST UNDER MONOTONIC AND CYCLIC LOADING WITH LARGE PLASTIC DEFORMATION

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KEY WORDS: ductile fracture simulation, seismic loading condition, finite element analysis

In order to design and maintain piping systems, fracture mechanics analysis under seismic loading is important. Seismic loading condition can be characterized by high strain rates and cyclic loading with large plastic deformation. Unlike high cyclic fatigue crack growth test, it has been shown that low cycle fatigue tests result in decreasing J-R curves and thus various low cycle fatigue tests with different test condition are required [1, 2]. In this respect it would be useful to develop a numerical methods to predict ductile tearing and J-R curves under very low cycle fatigue loading conditions. The author have recently proposed a numerical method to simulate ductile tearing under quasi-static, dynamic loading conditions based on the ductility exhaustion concept using the multi-axial fracture strain energy model [3]. In this paper, a numerical method to simulate ductile tearing is extended to cyclic loading conditions.

To determine the damage model, tensile and fracture toughness test were performed [2]. For experiments, SA508 Gr. 1a at room temperature was considered. To define monotonic and cyclic tensile properties, tensile tests were performed under monotonic and cyclic loading conditions. Saturated hysteresis loops are shown in Fig. 1. For fracture toughness (J-resistance) tests, standard 1T C(T) specimens were used. In order to verify the R-ratio effect of cyclic loading condition on J-R curves, tests with two value of the R-ratio (R=-0.5, -1) were performed.

To define cyclic material properties, 3th order non-linear kinematic hardening model in ABAQUS [4] was adopted. Kinematic hardening coefficients were fitted from experiment results and relevant values are tabulated in Table 1. FE simulation results using these parameters were compared with experiment results as shown in Fig. 1.

The damage model is based on the multi-axial strain energy. The multi axial fracture strain energy, $W_f$, is assumed to be given in terms of stress triaxiality by the following from,

$$W_f = A \exp \left( -C \frac{\sigma_{\text{eq}}}{\sigma_e} \right) + B,$$

Fig. 1. Cyclic stress-strain curves at 0.4, 0.8, 1.2%.

<table>
<thead>
<tr>
<th>$C_1$</th>
<th>$\gamma_1$</th>
<th>$C_2$</th>
<th>$\gamma_2$</th>
<th>$C_3$</th>
<th>$\gamma_3$</th>
</tr>
</thead>
<tbody>
<tr>
<td>200000</td>
<td>10000</td>
<td>35000</td>
<td>420</td>
<td>446</td>
<td>10</td>
</tr>
</tbody>
</table>
where $A$, $B$ and $C$ are material constants which can be determined by smooth and notch bar tensile test results under monotonic loading condition. Using monotonic tensile test results, the material constants were determined as below,

$$A = 2980; \quad B = 1.82; \quad C = 70,$$

(2)

Base on the this locus, incremental damage due to plastic deformation, $\Delta \omega$, can be calculated using the following equation,

$$\Delta \omega = \frac{\Delta W_p}{W_f}.$$

(3)

When the accumulated damage becomes critical ductile failure is assumed locally and incremental crack growth is simulated simply by sharply reducing all stress components at the gauss point.

FE simulation is carried out using the multi-axial fracture strain energy model. The proper value of critical accumulated value is chosen to fit the crack initiation toughness. Based on the determined damage parameters, C(T) test results under large-amplitude cyclic loading conditions are the simulated Fig. 2 shows FE results compared with experiment results. Simulated results shows good agreement with experiment results.

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**REFERENCES**


INFLUENCE OF SINTERING CONDITIONS ON MECHANICAL PROPERTIES OF AG-NANO SINTERED MATERIAL

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KEY WORDS: Ag-nano, SiC, sinter bonding, power device

Recently, the electrical vehicles have become the top targets of car makers. High electric power is necessary for those cars to run in motor drive, and power devices such as inverter are used for control to improve the efficiency. Since operative temperature of Silicon (Si) power device is about 150°C level, a big cooling system is needed for the module. These weight and size are big burden for vehicles, so lightening and downsizing are the key points. On the other hand, Silicon Carbide (SiC) power devices can increase the operative temperature up to 300°C and this advantage can downsize the cooling system. However, that temperature exceed melting point of solder, which is a common bonding material.

In this research, Ag-nano sinter bonding was chosen as an alternative material. When Ag-nano particles are sintered, it is possible to bond at a much lower temperature than the melting point of Ag bulk, and after sintering, the bonding layer shows the same melting point as the Ag bulk [1]. However, sintering mechanism is not completely understood because of its complexity, so sintered layer would contain a lot of voids. In general, it is said that voids affect mechanical property [2], so when sintered material is used as reliable bonding material, the presence of voids would be a major challenge. Therefore, in order to reduce these voids, they have to select proper sintering conditions. So in this study, the authors proposed a new approach getting mechanical property of sintered material in order to examine the effects of the pressure.

The authors reproduced a two-dimensional shape of the sintered material by using cross-section image and carried out FEM simulation. After the simulations, the results were compared with each pressure conditions and relationship between pressure and mechanical property of the sintered material was investigated.

Figure 1 shows an image of dummy-chip and acquisition point of cross section images. The dummy chip’s pressure conditions at bonding are 2.5 MPa, 5 MPa, and 10 MPa. Fig. 2 shows the cross-section images in the center of the chip by using a Scanning Electron Microscope (SEM), and FEM models from the acquired image. The size of these models is 4 µm square and the models were meshed using 0.01 µm square mesh.

And boundary conditions of analysis are shown in Fig. 3. By using ANSYS 13.0, tensile analysis was carried for each model. Forced displacement was applied to 1.0 % strain by 5 seconds and it was carried for X and Y
directions, and the pressure were applied in the y-direction. Mechanical properties of Ag are given to the blue area shown in Fig. 3.

The results of analytical are shown in Fig. 4 and Fig. 5. Fig. 4 shows Stress-Strain (S-S) curve. Fig. 5 shows Young’s modules calculated from S-S curve, and void ratio.

Fig. 4 and Fig. 5 show that the mechanical properties of the sintered material are greatly improved under the high pressure conditions, from the viewpoint of the Young’s modulus and plasticity characteristics.

And from these results, anisotropy can be confirmed, but the trend is different by the pressure. At the high pressures (5 MPa, 10MPa), the pressure direction is stronger structure than the other directions. However, at the low pressure (2.5 MPa), the trend is reversed. It means that, by the application of high pressure, a strong structure is formed by the pressure direction, in this case, Y direction.

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PLASTIC DEFORMATION AND FRACTURE MECHANISMS OF AZ31 MAGNESIUM ALLOY DURING TENSILE DEFORMATION: FROM THE VIEWPOINT OF TEXTURE

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KEY WORDS: magnesium, texture, plastic deformation

INTRODUCTION

It has been recognized that magnesium is one of the lightest structural materials. However, due to the limited ductility, most of structural magnesium parts are produced by casting and machining. Although modification of the ductility has been demanded for long time, the mechanisms of plastic deformation and following fracture are not fully understood.

It is widely said that the deformation of magnesium is dominated by the activation of the basal slip, (0001) <1120>. This is based on the studies using single crystals. Yoshinaga and Horiuchi [1] showed that the basal slip was the slip system with the lowest critical resolved shear stress (CRSS) and non-basal <a> slips (e.g. {1100} <1120>) had moderately high CRSSs. So-called <c + a> slip, {121} <1213> was the most difficult to be activated and they could not observe any slip band corresponding to this system. Therefore, the non-basal slip systems are often regarded as the immobile systems at room temperature.

However, there are only two independent slip systems belong to the basal slip. Therefore, if one considers von Mises criterion, it is impossible to deform polycrystalline magnesium only with the basal slip. However, in fact, a conventional magnesium alloy, AZ31, showed limited but certain ductility. In recent literature, numerical simulation for the texture formation during plastic deformation was often applied to explain the experimental result by room temperature deformation. In those cases, CRSS (or threshold stress for VPSC) of the <c + a> slip was set as 1.5 ~ 3 times as high as that for the basal slip [2]. These results suggest that the plastic deformation of magnesium is achieved not only by the basal but also non-basal slip systems.

In this study, uniaxial tensile deformation was conducted in order to observe change of texture during the deformation. The change of texture inferred the activation of non-basal slip systems as well as deformation twinning systems. It is concluded that the fracture of the sample is not because of the luck of slip systems but of the shear band formation resulted from strain localization.

EXPERIMENTAL PROCEDURE

A tensile specimen having cylindrical gauge with length of 20 mm and diameter of 5.0 mm was machined from an extruded bar of AZ31 (Mg-3Al-1Zn). The tensile axis was parallel to the extrusion axis. The sample was deformed with the crosshead speed of 3.33 × 10^-2 mm/s. After achieving a desired strain, the deformed gauge length was cut out by wheel saw and served for the texture measurement at the TOF neutron diffractometer, iMATERIA [3]. EBSD measurement was conducted on the cross section parallel to the tensile axis. In this paper, we reports the results of EBSD measurement.
RESULTS AND DISCUSSION

Fig. 1 is the engineering strain - engineering stress curve for the sample deformed up to fracture. The maximum elongation is 20.5 %, which agrees with the values reported by others. Fig. 2 is the result of EBSD measurements at different positions in the fractured sample. Since this sample showed obvious necking, the region closer to the fracture surface is more strained. Therefore, it can be said that the tensile axis moves toward <10¯10> and the fraction of twins increases with increasing strain. In the crystal direction maps, the gray regions have the tensile axis is close to <hki0> (threshold is 15°). It is seen that inside of twins (coloured boundaries) is mostly away from <hki0>. Therefore, the movement of tensile axis toward <10¯10> occurs in the matrix, suggesting the slip deformation. Since the basal slip merely be activated in the <hki0> textured specimen, the change of the texture should be attributable to the non-basal slip systems. The spatial distribution of twins is not homogeneous but forming shear bands in the region close to the fractured surface. Since the fractured surface is tilted 50° away from the tensile axis, it is likely that the shear bands formed by concentrated twins are the cause of fracture.

Fig. 1. Engineering stress – engineering strain curve for tensile deformation of AZ31 magnesium alloy at room temperature.

The rule of colouring for the crystal direction maps are indicated inside the figure.

REFERENCES

FRACTURE TOUGHNESS OF MASSIVELY TRANSFORMED AND SUBSEQUENTLY HEAT TREATED TIAL INTERMETALLIC COMPOUND

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KEY WORDS: TiAl intermetallic compound, massive transformation, fracture toughness

TiAl intermetallic compounds are considered to use for the high temperature structural applications due to their attractive properties. Especially, (α2+γ) two phase intermetallic compounds having fully lamellar structure has drawn considerable attention because of their superior high temperature strength and creep properties. However, low fracture toughness at room temperature restricts the use of these intermetallic compounds in practical applications. One of the methods to improve their fracture toughness is to control the lamellar colony size simultaneously with α2+γ lamellae size refining [1,2]. Generally, high temperature processing of TiAl intermetallic compounds is used in order to control the lamellar colony size by dynamic recrystallization and by crashing the boride precipitates if it is included [3],[4]. On the other hand, the method which uses massive transformation and subsequent heating can control the lamellar colony size without using thermo-mechanical processes. According to previous study, reheating of massively transformed microstructure into the (α+γ) two phase region may form fine convoluted microstructure which is the mixture of α single phased grains and γ single phased grains [5]. However, the process enabling to control the lamellar colony size in fully lamellar microstructure is still not known. Further, the relationship between microstructure formed by massive transformation and fracture toughness is also an open issue.

The objective of the present work is to understand the method to control the lamellar colony size in fully lamellar structure using massive transformation and subsequent heat treatment. Further, the relationship between fracture toughness and massively controlled microstructure is investigated.

All the specimens are held at α single phase region (1643 K) for 24 hours to homogenize the microstructure. The homogenized specimen shows fully lamellar structure where the average colony size is more than 1000 μm. Then, the specimens are heat treated at 1643 K for 30 min and cooled at 100 K/s by blowing He gas. Massively transformed γ (massive γ) structure with an average grain size of less than 10 μm is formed. Massive γ grains initiate from α grain boundary by satisfying the Blackburn relationship. Further, next neighbouring massive γ grains are often in a relationship of twinning.

Massively transformed specimens are reheated in α single phase region (1643 K), (α+γ) two phase region (1573 K) and both temperature range in various holding time. In case of the specimen where the massive γ is reheated at α single phase region for 10, 20 min, fully lamellar structure with the colony size of 690, 740 μm is obtained. In case of the specimens reheated at (α+γ) two phase regions (1543 K and 1573 K) for 60 min, fine convoluted microstructure with an average grain size of less than 50 μm is obtained. There exist equiaxed and acicular α grains. These two α grains initiate from massive γ grains during the reheating in two phase region. Acicular α grains satisfy Blackburn relationship with parent γ phase. In addition, when it is further heated at α single phase region for 1 min after heating at two phase region at 1543 K and 1573 K for 60 min, the average colony size of fully lamellar microstructure is 330 μm and
380 μm, respectively. The colony size formed by the processes is smaller than that in initial microstructure. However, many cracks formed in the specimen where that massive transformation has occurred. It is due to the initiation of volume expansion during massive transformation.

Fracture toughness was evaluated by three point bending test method where the chevron notch was introduced at the centre of the specimen having dimensions 3 x 4 x 22 mm. Room temperature fracture toughness of the massively transformed specimen showed lower value compared to that of the specimen where the massive transformation has not occurred. This is in contradiction with another our research obtained with experimental alloy where quite remarkable effect of similar treatment has been observed; fracture toughness increased by more than 50% for the room temperature properties and by about 25 % for high temperature properties simultaneously the strength increase by about 30 % [6]. The positive effect of the massive transformation products preceding the final heat treatment was there assured mainly to grain boundary behaviour of the lamellar colonies under absence of the γ equiaxed grains at the boundaries of lamellar colonies. The same probably is responsible for premature initiation of cracks at boundaries of lamellar colonies followed by rapid crack propagation along interlamellar boundaries.

REFERENCES
EFFECTS OF SLIP RATIO ON DAMAGE AND MICROCRACKS IN CARBURIZED SCM420 STEEL UNDER ROLLING CONTACT FATIGUE

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KEY WORDS: rolling contact fatigue, pitting, carburized steel, crack, internal stress

Carburized steel is widely used to the automotive gear parts, and the abrasion by rolling contact fatigue as called pitting or flaking determines these lives. The components running gear are demanded to reduce energy loss and to increase fuel efficiency in service, so that the inhibition of the abrasion is essentially needed to introduce low viscosity oil and lightweight and high strength of the parts. Not only microstructure but also the test conditions such as slip ratio affect rolling contact fatigue properties. Slip ratio influences internal stress distribution, and the test conditions have closely related to the fracture manner.[1] However, the mechanism of crack generation and the effects of test conditions on rolling fatigue characteristics have not been clarified yet. In this study, we discuss the effects of slip ratio on the damage formed by roller pitting test with high surface pressure for a typical carburizing steel SCM420.

The test material, SCM420 (C: 0.22, Si: 0.26, Mn: 0.86, Cr: 1.21, Mo 0.20 in mass %), was cut into a round bar specimen (small roller with outer diameter of φ26). The specimens were carburized in the vacuum condition with carbon potential, CP = 0.8 and carburized depth, ECD = 1.0 mm at 860°C for 40 min, and subsequently quenched into 80°C oil. SUJ2 quenched and tempered material was used for large roller (crowning 150R). Roller pitting test conditions were as follows: rotation speed 1500 rpm, slip ratio 0 % or -40 %, Hertz stress 3.5-3.9 GPa. Lubricant which was ATF oil (Idemitsu, ZEPRO ATF ECO) was supplied with amount of 2 l/min (from discharge side) at 90°C. Microstructural analyses were done by field emission scanning electron microscopy (FE-SEM) and electron backscatter diffraction (EBSD).

Figure 1 represents secondary electron images in the cross section of the samples at beneath the contact trail. Microcracks are seen at the rolling contact surface in all specimens. The angle between crack propagation direction and rolling direction increases as the increase of slip ratio. The reason why is attributed to change of tangential force by slip ratio.

Misorientation (KAM: Kernel Average Misorientation) in the cross section beneath the non-contact trail and contact trail was evaluated. KAM value beneath the contact surface was higher than that beneath the non-contact surface. Their KAM maps revealed an increase of average KAM value with the increase of slip ratio. Average KAM value was significantly high in a deeper region than 160 µm from the contact trail in the samples of slip ratio -20 %, pressure 3.5 GPa and slip ratio -40%, pressure 3.7 GPa. It suggests that the higher strain was induced in the region matching to the maximum shear stress applied and related to the internal stress distribution beneath the contact trail.
Fig. 1. SEM images in the cross section beneath the contact trail for the samples: (a) pressure 3.9 GPa, slip ratio 0%, (b) pressure 3.5 GPa, slip ratio -20% and (c) pressure 3.7 GPa, slip ratio -40%. Arrows show the rolling direction of the specimen.

REFERENCES

EXTRAPOLATION OF IMAGINAL MINIMUM CREEP RATE
IN COMPRESSION BY A CONCEPT OF SATO-INDEX

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KEY WORDS: creep curve, extrapolation, primary creep, high temperature deformation

Creep characteristics of alloys and compounds have been evaluated mainly by the minimum creep rate or the steady state creep rate, and by its stress and temperature dependences. In some cases, however, direct comparison of the minimum creep rate or the steady state creep rate are not practically easy due to the very long period of primary stage of creep deformation. The minimum creep rates are not always representative value directly evaluated from experiments. It should be valuable if one could estimate the minimum creep rate from creep curve in primary stage. I have proposed a method of quantitative evaluation of creep curve based on the evaluation of strain rate change and its strain dependence during creep [1-2]. The value that reflects a shape of creep curve is named “Strain Acceleration and Transition Objective-Index (SATO-Index)” [3]. SATO-Index and a differential equation show a strain dependence of strain rate and lead creep curve by numerical integration. The method provides quantitative information of shape of a creep curve. The SATO-Index, \( \alpha \), is defined as follows.

\[
\alpha = \left[ \frac{d^2}{d\varepsilon^2} \log_{10}(\dot{\varepsilon} / s^{-1}) \right]_{\varepsilon}.
\]

The \( \alpha \) corresponds to the curvature of the common logarithm of strain rate as a function of strain. The value is defined at a strain, \( \varepsilon \), and at a time, \( t \). Based on the definition of SATO-Index, creep curve, i.e., strain as a function of time, \( \varepsilon(t) \), can extrapolated with suitable initial conditions. Common logarithms of strain rate, \( \log \dot{\varepsilon} \), as a function of strain, \( \varepsilon \), can be described as the following equation (2).

\[
\log \dot{\varepsilon} (\varepsilon) = \frac{\alpha}{2} (\varepsilon - \varepsilon_{\text{min}})^2 + \log \dot{\varepsilon}_{\text{min}},
\]

Equation (2) can be solved numerically and gives creep curve, \( \varepsilon(t) \). Here, \( \varepsilon_{\text{min}} \) and \( \dot{\varepsilon}_{\text{min}} \) are an imaginal strain at a minimum strain rate and an imaginal minimum strain rate extrapolated from observed creep curve. These values can be reasonably extrapolated from experiments. In solid solution alloys, and entire creep curve is reasonably extrapolated from a part of creep curve based on evaluation of SATO-Index [2-3]. In this presentation, examples of evaluation and extrapolation of creep curve in compression are presented. It is concluded that the extrapolation with the concept of SATO-Index reasonably provides imaginal minimum creep rate in compression. Possibility of comparison and evaluation of creep behaviour from a part of creep curve, especially from creep curve in primary stage will be discussed.

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REFERENCES
NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS
1 – 4, NOVEMBER, 2016, YOKOHAMA, JAPAN

DAMAGES OF MACHINES AND STRUCTURES IN GREAT EAST JAPAN EARTHQUAKE DISASTER AND LESSONS LEARNED FROM THE DISASTER

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KEY WORDS: great east japan earthquake disaster, tsunami, damages of machines and structures, robots, traffics, energy infrastructures, severe accident in NPP, codes and standards, crisis management

The Tohoku Region Pacific Coast Earthquake and following tsunami, which occurred on March 11th 2011, caused unprecedented devastation in Japan, especially to the Tohoku and North Kanto regions. This event has become known as the Great East Japan Earthquake disaster.

Furthermore, the earthquake and tsunami seriously damaged the Fukushima Daiichi Nuclear Power Plants (NPP), resulting in the meltdown of the fuel in the reactor core, the destruction of the nuclear reactor buildings due to hydrogen explosions and large-scale release of radioactive materials into the environment, which has destroyed the lives of people living in that area. A catastrophe of this extremity has never before been experienced by Japan.

This disaster was unique in the following ways:

- The magnitude of the earthquake was enormous at M9.0.
- The scale of the tsunami caused by the earthquake was huge.
- An extensive area was affected and a great number of people suffered as a result.
- Previously unencountered challenges were met in tackling the NPP incident and controlling the release of the radioactive materials.

The Japan Society of Mechanical Engineers formed the “JSME Research Committee on the Great East Japan Earthquake Disaster” soon after the earthquake under the direct leadership of the executive committee. There were many areas to be assessed and many subjects to be addressed. In order to do this job effectively, the JSME established eight working groups (WGs) under the committee:

WG0: Characteristics of the Earthquake and Tsunami
WG1: Damage to Machines and Equipment and Good Practices for Seismic Countermeasures
WG2: Understanding the Mechanism of Tsunami-induced Damage to Machines and Structures Based on Mechanical Analysis
WG3: Application of Robot Technologies to the Disaster Sites
WG4: Analysis of Traffic and Physical Distribution Systems within the Disaster Areas
WG5: Damages to Energy Infrastructures
WG6: Codes and Standards Issues and Future Perspective
WG7: Crisis Management for Earthquakes, Nuclear Power Plant Accidents and Other Events

The committee began its activities at the end of March, 2011. Each WG worked with great motivation and gathered many data about the damages sustained. They also determined the lessons that have to be learned from the disaster and how to incorporate them into our practices in the future.

The extended summary of the report was published in English in August, 2014, entitled “Lessons Learned from the Great East Japan Earthquake Disaster -Report of the JSME Research Committee on the Great East Japan Earthquake Disaster-“. The readers can download the pdf from the website: http://www.jsme.or.jp/English/

In the first chapter, four lessons learned through research are presented in the form of proposals. These were arrived at through extensive discussions between the committee members about what we, as mechanical engineers, can learn from the disaster and what we can contribute to society as a result. The four proposals are summarized as follows.

I To develop the approach to system integration of large scale systems.
II To review how the design basis is determined and how we can prepare for events beyond the scope of the design basis.
III To better inform the public about risks associated with new products.
IV To incorporate the lessons learned into our codes and standards, and foster engineers with the skills to tackle disaster related tasks, with the aim of passing these lessons on to future generations.

Then, the reports and proposals obtained in each working group are described in detail in the following chapters.

This presentation will give you a brief summary of the report.
FAILURE ANALYSIS OF BIG TURBINE BLADES

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KEY WORDS: low-pressure turbine part, fatigue, fracture morphology, beach marks, striation, microstructure

The blade of the third turbine wheel (3TW) placed on the third low-pressure part (3LP) of the TG 1000 MW turbine broke during the start of turbine. The low-pressure part blades of the third wheels are made from modified 12 % Cr martensitic steel AK1 TD.9, and the blades of the fourth wheels are made from X2CrNiMo13-4 steel. The study of causes and fracture mechanisms and also an analysis of the microstructure were performed. The main goal of these analyses was the description of the failure (crack initiation and propagation) mechanisms. Service loading history combined with fractographic findings offered information for a reconstruction of the blade failure history. Obtained results are summarised in presented article.

Fig. 1. Damage of the third low-pressure part LP3 after one blade fracture.

Fig. 2. The fracture of the blade (No. 152) in the third wheel from LP3.

Fig. 3. Fractographic reconstruction of fatigue failure history of blade.

Acknowledgement: The authors gratefully acknowledge the support by Faculty of Nuclear Sciences and Physical Engineering, Czech Technical University in Prague.

REFERENCES

The effect of chloride ion concentration on SCC susceptibility of 15Cr-6Ni-2Mo [1] martensitic stainless steel was investigated by slow strain rate test (SSRT). In addition, the potential of the hydrogen embrittlement at high temperature was examined by SSRT with electrochemical polarization.

Transgranular SCC occurred in the water containing CO$_2$ and chloride ion at 180°C with 4.17 x10$^{-7}$ sec$^{-1}$ strain rate. The chloride ion concentration affected SCC susceptibility. Quasi-cleavage fracture was observed on the fracture surface. In addition, cathodically charged SSRT specimen showed the embrittlement surface similar to SCC fracture surface at high temperature. On the other hand, the SCC was accelerated by anodic polarization but not by cathodic polarization. It seems that hydrogen contribute the mechanism for SCC of martensitic stainless steel at high temperature. The crack propagation of high temperature SCC is discussed in relation with hydrogen embrittlement.

Table 1 Chemical Composition (mass%) and Mechanical properties (MPa).

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>YS</th>
<th>TS</th>
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<tr>
<td>15Cr SS</td>
<td>0.03</td>
<td>14.7</td>
<td>1.92</td>
<td>6.22</td>
<td>987</td>
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</tr>
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<td>13Cr SS</td>
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<td>12.5</td>
<td>0.01</td>
<td>0.09</td>
<td>585</td>
<td>760</td>
</tr>
</tbody>
</table>

Fig. 1. Effect on SCC susceptibility by SSRT, 4.17x10$^{-7}$ sec$^{-1}$, 180°C, CO$_2$ 3MPa.

Fig. 2. Effect of polarization on SCC susceptibility in SSRT test.

REFERENCES

EFFECTS OF WEIGHT OF HERBERT PENDULUM ON HARDNESS EVALUATION

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KEY WORDS: hardness, pendulum, free damped vibration, measurement, mechanical testing

E. G. Herbert developed a Herbert pendulum hardness tester in 1923 [1]. A Herbert pendulum swings with the indenter as a fulcrum on a specimen. Hardness of the specimen is obtained using the attenuation curve of the pendulum. The attenuation behavior varies with the hardness of the specimen because the rolling resistance of the indenter on the specimen is dependent on the hardness. A Herbert hardness tester is not used for industrial purposes due to poor measurement accuracy. Recently, Matsubara et al. improved a Herbert hardness tester and highly accurate swing angle detection of the Herbert pendulum became possible [2-3]. The improved Herbert hardness tester is industrial useful tester and can measure hardness of various specimens such as metals, ceramics and resins. In this study, the effects of the test load (pendulum weight) on the damping behavior and damping hardness are investigated using the two types of the pendulums with different weights. Damping hardness, \( \alpha \), is the damping coefficient of an attenuation curve of the pendulum (Fig. 1).

Lightweight Herbert pendulum (about 1.5 kg) made of an aluminum alloy and heavy Herbert pendulum (about 7.4 kg) made of a stainless steel are used for the hardness test (Fig. 2). The Herbert hardness testing system is shown in Fig. 3. The mean swing cycle of the pendulum on the sapphire is adjusted to 20 s as the calibration before the Herbert hardness test. Brinell standard hardness blocks (HBW 150, 200, 250, 300, 350, 400, 450, 500, 550 and 600) are used as the specimen. A surface of the specimen is polished with emery paper up to a 1200 grade and treated with acetone. The pendulum is fixed with an initial angle \( \theta = 30^\circ \) using a solenoid on the specimen and is released from the solenoid. The swing angle is measured with two laser displacement sensors until the pendulum swings 5 cycles.

Fig. 1. Schematic drawing of an attenuation curve of a Herbert pendulum.

Fig. 2. Herbert pendulums made of (a) an aluminum alloy and (b) a stainless steel.
The attenuation behaviors of both pendulums for HBW 150 are shown in Fig. 4. The attenuation behavior of the stainless steel type pendulum is the same as that of the aluminum alloy type pendulum, although high pendulum weight causes high rolling resistance of the indenter. This is because that the stainless steel type pendulum has not only high rolling resistance but also high inertia moment.

Damping hardness obtained by the aluminum alloy and the stainless steel pendulums is plotted as a function of Brinell hardness shown in Fig. 5. The relationship between damping hardness and Brinell hardness obtained by the stainless steel type pendulum is similar to that obtained by the aluminum alloy type pendulum. This result indicates that the difference of the pendulum weight has little effect on the damping hardness for metals.

![Fig. 3. Herbert hardness testing system.](image)

![Fig. 4. Attenuation behaviors for HBW150 obtained by the aluminum type and the stainless type Herbert pendulum.](image)

![Fig. 5. Damping hardness plotted against Brinell hardness.](image)

**REFERENCES**


FATIGUE CRACK GROWTH SIMULATION USING S-VERSION FEM: APPLICATION TO INTERACTING SUBSURFACE CRACKS

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KEY WORDS: S-version FEM, fatigue crack growth, subsurface cracks, proximity rule

Subsurface cracks are known to be generated in materials because of the growth of initial defects inserted at the manufacturing process. During the use of the materials, the materials are subjected to a cyclic loading, and as the consequence, fatigue crack growth of the subsurface cracks are occurred. In order to evaluate simply the fatigue crack growth behaviour and the residual lifetime of the materials, ASME provides us with proximity rules for cracks [1]. In the proximity rules, cracks are converted to a single elliptical cracks, if the distance between the cracks lowers a prescribed critical distance. On the other hand, if a crack comes to the free surface within a critical distance, the crack is approximated to a single semi-elliptical surface crack. Using the database of the stress intensity factor for elliptical and semi-elliptical surface cracks, the fatigue crack growth behaviour and residual lifetime can be simply evaluated. However, in some cases, the proximity rules change the shape and size of cracks drastically so that the accuracy and reliability of the proximity rules must be carefully evaluated.

In this study, fatigue crack growth of interacting subsurface cracks using s-version finite element method (SFEM) [2] is presented. Using the remarkable property of SFEM particularly in mesh generation processes, the fatigue crack growth simulation can be easily performed. In order to evaluate the accuracy and reliability of the proximity rules, the fatigue crack growth simulation with and without the application of the proximity rules are carried out. Finally, in order to improve the accuracy and reliability of the proximity rules, the proximity rules are slightly modified, and are verified using the fatigue crack growth simulation using SFEM.

The fatigue crack growth simulation is enabled by using the SFEM. In the SFEM, the specimen and crack geometry can be separately modelled with a global and local meshes. The global and local meshes do not need to have a smooth connection in-between so that the SFEM drastically reduces the requirement and complexity in the meshing process during the fatigue crack growth simulation. At each fatigue crack growth step, we update only the local mesh in accordance with the updated crack shape, while the global mesh is only made once at the beginning of the simulation, and is used repeatedly at each fatigue crack growth step. The stress intensity factor is calculated using the virtual crack closure method (VCCM), and crack growth amount is calculated using the Paris law.

Fatigue crack growth simulation of two interacting subsurface cracks aligned in the depth direction of specimen is performed. The specimen and the crack geometries are shown in Fig. 1. The specimen is subjected to a cyclic tension loading with a stress ratio of \( R = 0.1 \). The crack growth behavior is shown in Fig. 2. The color shows the normal stress acting on the cross-section, where the cracks are located. Initially, two cracks are separated with a distance each other. As the cracks grow, the cracks directly meet each other. Then,
the overlapped parts of the cracks are manually eliminated, and the cracks are connected for making a single crack. The single crack is modelled with a single local mesh. After the coalescence of the cracks, the crack comes to the free surface of specimen. When the crack penetrates the free surface, a part of crack located at the outside of specimen is removed, and the crack is converted to a surface crack. Finally, the surface crack grows to be a semi-elliptical surface crack. In order to evaluate the accuracy and reliability of the proximity rules, the fatigue crack growth simulation is again performed with the application of the proximity rules. In the simulation, the coalescence of two cracks and crack penetration are dealt with in accordance with the proximity rules.

The crack growth behavior in the depth direction of the specimen is summarized in Fig. 3. Comparing the fatigue crack growth behavior simulated with and without the application of proximity rules, the results clearly illustrate that the crack growth behavior with the application of the proximity rules substantially faster than that without the application, meaning that the proximity rules provide us with a substantially conservative evaluation of crack growth behavior and rate. To improve the accuracy of the proximity rules, the proximity rules are slightly modified. In the modified proximity rules, the shape of approximated cracks can be an elliptical shape with a major axis along the depth direction. The new proximity rules give better approximation of cracks in terms of the crack shape and size. The fatigue crack growth simulation results using the modified proximity rules is also plotted in Fig. 3. Comparing the three types of fatigue crack growth simulation results, it can be found that the modified proximity rules can provide better evaluation of the fatigue crack growth behavior. Therefore, the fatigue crack growth simulation results suggest that the modified proximity rules have a great potential to have better fatigue crack growth evaluation.

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REFERENCES
DYNAMIC BEHAVIOR OF HIGH STRENGTH ARMOR STEEL PLATES

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KEY WORDS: armor steel, tensile test, high strain rate, strain rate sensitivity, fractography, dimple fracture.

High strength armor steels are utilized especially for the ballistic protection of automobiles and combat vehicles therefore they are required knowledge of their behavior under dynamic loading. Armor steel response on load at the short period of time and dependence of deformation speed must be studied in order to understand ballistic impact phenomenon, create reliable behavior predictions and construct elements with optimized design. It is well known, that mechanism of high-speed deformation is different from quasi-static tensile test due to adiabatic heating and transformation of plastic work into heat, which are not dissipated before the end of deformation and specimen rupture.

This paper describes mechanical properties of armor steels Armox 500T and Secure 500 under dynamic tensile loading. Both steels have mid-carbon content and are heat-treated by quenching and subsequent tempering. Chemical compositions of the steels are tabulated in Table 1.

Table 1 Chemical composition of steels used in this study.

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>Al</th>
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<tr>
<td>Armox 500T</td>
<td>0.32</td>
<td>0.4</td>
<td>1.2</td>
<td>0.015</td>
<td>0.010</td>
<td>1.0</td>
<td>1.8</td>
<td>0.7</td>
<td>0.7</td>
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<tr>
<td>Secure 500</td>
<td>0.32</td>
<td>0.4</td>
<td>1.0</td>
<td>0.015</td>
<td>0.005</td>
<td>1.5</td>
<td>0.6</td>
<td>3.7</td>
<td>0.05</td>
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</table>

Steel plate specimens with initial gauge length of 50 mm were tested under quasi-static tensile test with strain rate 0.001 s⁻¹ and high speed tensile test machine INSTRON VHS 80-20 within range of intermediate strain rates of 100 - 400 s⁻¹ at the room temperature. Specimens were also tested for hardness. Average value of hardness HV10 was 495 in case of Armox 500T and 508 for Secure 500, respectively. Quasi-static tensile tests were conducted on Zwick Z150 machine and their results confirmed material properties specified by the manufacturer.

Under quasi-static tensile conditions, Secure 500 steel exhibits average values of yield strength R_p0.2 1427 MPa and ultimate tensile strength 1667 MPa. Armox 500T yield strength is 1342 MPa and UTS 1614 MPa on average.

The results of high speed deformation did not exhibit strain rate sensitivity of yield strength. Sensitivity of elongation on strain rate was observed and Armox 500T steel reached higher elongation values, up to 15.6 %. Stress-strain curves at various strain rates are shown in Figs. 1 and 2.

Total absorbed energy was calculated from the stress-strain curves by means of total area under the curve, which rises with increasing strain rates. The morphology of fracture surfaces observed by scanning electron microscope (SEM) indicates ductile fracture pattern with dimples. It is found that with increasing speed of deformation, size of the dimples shrinks for both steels. This points out to localized increase of the deformation and shorter time period between coalescence of cavities and specimens rupture.
NEW METHODS OF DAMAGE AND FAILURE ANALYSIS OF STRUCTURAL PARTS
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Fig. 1. The stress-strain curves of Armox 500T steel. Fig. 2. The stress-strain curves of Secure 500 steel.

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OSTRAVA!!!

REFERENCES


MICROSTRUCTURE CHARACTERIZATION OF A WELDED ROTOR

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KEY WORDS: heterogeneous welds, PWHT, heat affected zone, microstructure stability, hardness profile, minor phases

The paper deals with microstructure characterization of a X14CrMoVNbN 10 1/27NiCrMoV 15 6 heterogeneous weld in a welded rotor being developed for modern turbines of coal fired power plants. The aim of this development is to reduce production costs of heavy rotors for applications in thermal power plants [1]. Detailed knowledge about the effect of welding on both short-term and long-term mechanical properties and on microstructure stability of welded rotors is needed.

Welding of the rotor was performed on a vertical Polysoude welding equipment in a protective gas atmosphere using the TIG technology with a supply of preheated filler wire (HOT WIRE TIG technology). The weld investigated was manufactured in two steps. In the first step, a multi-bead overlay of the P24–IG filler material was deposited on the X14CrMoVNbN 10 1base material. Subsequently, the PWHT at 690°C was applied. In the second step, welding of the X14CrMoVNbN 10 1 and 27NiCrMoV 15 6 rotor parts using the NiCrMo2.5-IG filler material was carried out. After that the final PWHT of the welded rotor at 590°C was performed. Figure 1 shows a macroetch through the weld investigated. No welding defects were detected. Hardness profile across the weld is shown in Fig. 2. As evident, hardness in heat affected zones (HAZ) of both base materials increased towards the fusion zones. In the overheated part of the HAZ of the 27NiCrMoV 15 6 steel hardness values exceeded the critical value of 350 HV1.

Microstructural characterization of the weldment was focused on identification of minor phases in basic parts of the weld, including heat affected zones. TEM investigations were carried out using carbon extraction replicas. Both EDX microanalyses and electron diffraction studies were used for identification of minor phases. The most questionable area of the weld investigated represents the X14CrMoVNbN 10 1/ P24–IG fusion zone. Due to pronounced differences in carbon activity in both materials redistribution of interstitial elements from the P24-IG overlay into the X14CrMoVNbN 10 1 steel is to be expected. In the area of the fusion zone intensive precipitation of M23C6 and chromium rich M2X phase was accompanied by a significant precipitation of M6X particles. This phase exhibits a high rate of coarsening. In the X14CrMoVNbN 10 1base material this minor phase was not present.

Fig. 1. Macroetch of the heterogeneous weld, WM = NiCrMo2.5-IG filler material.
Precipitation in the multi-bead overlay of the P24-IG steel was very heterogeneous. The applied PWHT regimes resulted in the intensive precipitation of $\text{M}_2\text{C}_6$, $\text{M}_7\text{C}_3$ and MX phases. PWHT of the NiCrMo2.5-IG weld metal at 590°C resulted in precipitation of cementite ($\text{M}_3\text{C}$). In the quenched and tempered matrix of the 27NiCrMoV 15 6 steel the following minor phases were identified: $\text{M}_3\text{C}$, $\text{M}_2\text{C}_6$ and $\text{M}_7\text{C}_3$. Precipitation in the HAZ of the 27NiCrMoV 15 6 steel was qualitatively identical as that in the base material, because PWHT was carried out at the same temperature as tempering of the 27NiCrMoV 15 6 steel during the quality heat treatment (590°C). Figure 3 shows a typical precipitation in the HAZ of the 27NiCrMoV 15 6 steel.

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**REFERENCES**

THE STABILITY OF RETAINED AUSTENITE AND TRANSFORMATION BEHAVIOUR IN TRIP STEELS AT LOW TEMPERATURE

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KEY WORDS: TRIP steel, retained austenite, deformation induced transformation, EBSD

Low alloy multiphase steels associated with transformation-induced plasticity (TRIP) of retained austenite (γR) shows a good balance of tensile strength and ductility, since the TRIP causes to maintain higher work-hardening rate in the high strain regime.[1] Then the stability of γR is a key factor to control the TRIP effect.[2],[3] Although the stability of γR depends on various factors such as chemical compositions, morphology and size of austenite, the transformation behaviour of individual γR has not been clear. In the present study, the influence of stability of γR on work-hardening and transformation behavior at low temperature has been investigated for TRIP steels with different chemical compositions.

Two types of low alloy TRIP steel sheets with different volume fraction of γR (low-γ: 9.4% and High-γ: 17.2%) containing about 1.3 mass% C were used. The steels were cold-rolled and annealed at 1063 K in α+γ region, and then cooled to 673 K for austempering. Tensile tests were carried out at 77 K, 193 K, 233 K and 293 K. The microstructure and crystal orientation of TRIP steels were analyzed using electron backscattered diffraction (EBSD). The microstructure consists of ferrite matrix (αf), bainite (αb) and γR. The γR grains were distributed in αf grains and at the grain boundaries between αf and αb.

Figure 1 represents the stress - strain curves for the High-γ steel. The steels showed high ductility and tensile strength at 193 K, 233 K and 293 K. The tensile strength of steels at 77 K was much higher than that of other test temperatures, although the steels exhibited early fracture. Figure 2 shows the relationship between work-hardening rate and true strain curve for the High-γ steel. The work-hardening rate was increased with decreasing test temperature. At 193 K and 233 K, the rate was kept with high value in the high strain regime. The stability of γR at 293 K was much higher than that of other test temperatures. The γR at 233 K and 193 K at 10% strain was mostly transformed. Then the difference of the work-hardening rates among the temperatures may depend on the stability of γR and the strengthening of αf at lower temperature, although the influence of TRIP on ductility has not been clear yet.

The most of γR grains after tensile test at 193 K, 233 K and 293 K revealed the orientation near <111> parallel to the tensile direction, in which their Schmid factors were low as shown in Table 1. It suggested that the deformability of γR commonly affected to the martensitic (α’) transformation in the steels at 293 K. The volume fraction of γR at 77 K was less than 1% at 3% strain, the most of γR were transformed to α’ under stress concentration. The work hardening rate at 193 K, 233 K and 77 K in the steels may depend on their volume fraction of γR.
Fig. 1. Stress-strain curves at each test temperature for the High-γ. [4].

Fig. 2. Work hardening rate-true strains curves at each test temperature for the High-γ.

Table 1 Volume fraction of γ_R and ratio of near* <111>γ in γ_R after tensile test at each test temperature for the High-γ(* Tolerance angle is 15 degree). [4].

<table>
<thead>
<tr>
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<th>293 K</th>
<th>233 K</th>
<th>193 K</th>
<th>77 K</th>
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<tr>
<td>γ_R (vol%)</td>
<td>6.1</td>
<td>0.6</td>
<td>0.4</td>
<td>1.0</td>
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<tr>
<td>Near* &lt;111&gt;γ</td>
<td>25.8%</td>
<td>43.1%</td>
<td>45.2%</td>
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REFERENCES
ANALYSIS ON CRITICAL CTOD OF LONG-TERM USED PENSTOCK

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KEY WORDS: CTOD, fracture toughness, chemical compositions, carbon steel, long-term used

A number of studies have been conducted to investigate fracture toughness of structural steels. However, there are a limited number of studies for long-term used structural steels. Given the recent engineering practice of life extension for the existent steel infrastructures, it is very important to develop reliable methodologies for proper evaluation of infrastructure integrity. Correspondingly, as an extension of previous study, fracture toughness of a carbon steel (JIS SS400) that has been used for the penstock of a hydroelectric power plant for about 50 years was examined. By measuring the critical crack tip opening displacement (CTOD) in conjunction with analysis for chemical compositions, the characteristics of fracture toughness in terms of critical CTOD was investigated.

The specimens were cut out from the penstock that has been in service for a domestic hydroelectric power plant for about 50 years, and their thickness were 10, 15, 20 and 25 mm. The critical CTODs were measured by changing the notched position (at base metal, heat affected zone, bond, and weld metal, respectively). Based on fracture behaviour, the critical CTOD can be classified as:

- \( \delta_c \): CTOD at the onset of brittle fracture
- \( \delta_u \): CTOD at the onset of brittle fracture after stable crack propagation
- \( \delta_m \): CTOD at the onset of ductile fracture

Fig. 1 shows the variations of critical CTOD with different notched positions. As shown, \( \delta_c \) exhibited smaller value than \( \delta_u \) and \( \delta_m \). Further, \( \delta_c \) shows the minimum value when the
notched position is at weld metal. It is noted that the effect of testing temperature ($T = 0$ and $-10^\circ$C) was negligibly small. The previous study showed that measuring the chemical compositions is an effective mean to estimate $0.2\%$ proof strength and ultimate tensile strength. Therefore, the correlation of $\delta$ to the carbon (C) and the carbon equivalent (Ceq) contents were examined for the base metal. As shown by Fig. 2, $\delta$ decreases in conjunction with increase of C and Ceq, and good correlations (C: -0.894 and Ceq: -0.921) are observed. It is well known that the presence of C causes the degradation in toughness of steel because of nucleation of micro-crack at carbide. Concerning Phosphorus (P) and Sulphide (S), their effects were not clearly observed. In general, toughness of material is dependent on the microstructure of material, thus the use of chemical composition to estimate toughness may not be convincing. However, all the specimens were composed of ferrite and pearlite and significant difference in grain size was not observed. Because of this circumstance, the chemical compositions can serve as an effective mean to evaluate fracture toughness.

Currently, it is common to use the cut-out specimens to perform the tests and characterize the material properties and performance. However, such approach causes temporal malfunction of the operating infrastructure. This study shows that non-destructive evaluation on fracture toughness of long-term used steel is possible, but further studies are necessary to assess the applicability and reliability.

REFERENCES


EFFECT OF PRE-CRASH PHASE ON OCCUPANT PROTECTION WITH SEATBELT

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KEY WORDS: occupant protection, pre-crash, restrain system, driver behaviour

In general, crash safety performance of frontal collision for automobiles is assessed with a crash test dummy, and the posture of the dummy on a seat is supposed with standard driver’s position. According to reports of traffic accident analysis [1], the drivers of 60% in the frontal collision did avoidance maneuver for the accident such as a braking and handle operation. The posture of the driver at the collision might be deferent from the standard driver’s position applied to the crash safety assessment, and occurrence ratio of chest injury in the frontal collision was influenced with and without the braking operation according to the traffic accident analysis report [1] in the real world. Therefore, it is necessary to investigate effect of driver’s posture at the collision on injury pattern experimentally for the consideration of strategy for occupant protection, because it is difficult to clear its effect only by the traffic accident analysis.

Recently, the pre-crash safety concept was focused in terms of occupant protection in an automobile. The pre-crash means the phase just before the collision. A pre-crash seatbelt have been applied to part of automobiles. Its seatbelt pulls back the occupant to the standard driver’s position during the braking at the pre-crash phase. In terms of performance assessment of the pre-crash seatbelt to the occupant protection, experimental verification concerning the avoidance maneuver of the driver is desired.

This study was performed in Japan automobile research institute (JARI) as the previous work of author. The contents of this report already have been presented and published in a journal [2].

Figure 1 shows the pre-crash trolley used in the crash test. The trolley consisted of two rigid seats, foot plates and 3-point seat belts with an emergency limitation retractor (ELR). The trolley pulled with traction system and moved to a crash barrier. Deceleration of the trolley in the pre-crash phase was simulated with disk brake system, and the trolley was collided to the barrier. Impact from the barrier to the trolley was simulated with turning device on the trolley. Figure 2 shows time history of the trolley decelerations during the braking and after the collision. Vertical axis of left hand side is deceleration before the collision, and the one of right hand side is deceleration after the collision. 0ms means time at the collision to the barrier.

Dotted line means deceleration corridor of the frontal collision defined with ECE-R17 regulation. In the present test trolley, the deceleration was almost constant before the collision.
400ms of the collision and simulated the one of ECE-R17 regulation after the collision.

In this experiment, behaviour of the occupant posture in the pre-crash phase should be simulated with the crash test dummy (Hybrid-III). Before the experiment, the bending characteristics of the dummy for frontal collision was checked and improved by using data obtained with volunteers test under low frontal impact condition. Figure 3 shows the comparison of flexion angle between volunteers and the dummy. The flexion angle was defined as angle between two lines (Hip-T1, Hip-Knee). The averaged flexion angle obtained with 4 volunteers is grey solid line. The standard deviation of the volunteers shows grey dotted line. The flexion angle of the original test dummy which is depicted with thin solid line was far different from that of the volunteers. From this result, the bending characteristics of lumber spine in the test dummy was improved in order to simulate the flexion of the volunteers. The flexion angle of the improved dummy shown with thick line became closer to the volunteers than the original dummy. Therefore, it seem that behaviour of the occupant in the pre-crash phase could be simulated by using the improved one as compared with the original one.

In the traffic accident of frontal collision, the injury of the chest occurred much higher than that of the torso according to the traffic accident analysis report[1]. The effect of the braking on the chest injury was investigated based on the chest acceleration as chest injury criterion. In this experiment, the twice crash tests were carried out for each test condition which was with and without the braking. In the case without braking, the trolley was collided to the barrier with 48km/h. In the case with the braking, the trolley moved with 67km/h, and then it was decelerated from 67km/h to 48km/h by the braking and was collided to the barrier. Thus, collision speed of the trolley (48km/h) in the both cases was the same. Figure 4 shows time history of the chest acceleration with and without the braking in the pre-crash phase. From the result, the peak chest acceleration in the case of the braking was higher than that in the case of the non-braking under the same collision speed of the trolley. It seems that it was caused by forward posture of upper torso by the braking.

REFERENCES
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