Review

Theory of particle transport phenomena during fatigue and time-dependent fracture of materials based on mesoscale dynamics and their practical applications

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Abstract: In this work, the mesoscale mechanics of metals, which links their microscopic physics and macroscopic mechanics, was established. For practical applications, the laws for quantitatively predicting life of cycle and time-dependent fracture behavior such as fatigue, hydrogen embrittlement, and high-temperature creep were derived using particle transport phenomena theories such as dislocation group dynamics, hydrogen diffusion, and vacancy diffusion. Furthermore, these concepts were also applied for estimating the degree of viscoelastic deterioration of blood vessel walls, which is dominated by a time-dependent mechanism, and for the diagnosis of aneurysm accompanied by the viscoelastic deterioration of the blood vessel wall. In these theories, new mechanical indexes were derived as dominant factors for predicting the life of fatigue crack growth and the time-dependent fracture of notched specimens of materials such as hydrogen embrittlement and high-temperature creep. Furthermore, as an example of a practical application, these theories were applied to estimate the degree of viscoelastic deterioration and chaotic motions of blood vessel walls, which are closely related to blood vessel diseases such as atherosclerosis and aneurysm. Moreover, new indexes to diagnose them were also proposed for clinical applications.

Keywords: dynamic factor, α method, HE parameter, $Q^*$ parameter, $I^*$ parameter, chaos theory

1. Introduction

Research on strength of materials has been performed in the fields of materials science and applied mechanics. This research was targeted toward the development of the mechanical properties of materials and their practical application to the security of structures.

The former field involves clarification of the fracture mechanism,1)–7) while the latter involves establishment of laws for predicting fracture life (for example, fatigue and creep crack growth life),8)–10) standardization of mechanical testing methods such as those recommended by ASTM and ISO standards,11),12) and proposing nondestructive inspection methods for establishing fracture prevention technology for structures as a final goal. Because advanced materials such as nickel-based super alloys and high-strength steels used in high-temperature, hydrogen-rich, or corrosive environments, exhibit creep brittle or hydrogen embrittlement properties, there are many cases wherein final fracture is dominated by micro damage formation, crack initiation, and initial growth at the early stage of fracture life.13) Therefore, for practical applications of these materials to engineering structures, it is necessary to clarify behaviors of early-stage crack growth and micro damage formation and establish the law for fracture life.13) For this purpose, it is necessary to derive the theory and law for predicting fracture life by establishing mesoscale mechanics that links metal physics with macroscale mechanics.

Previously, the theory of multiscale fracture mechanics, i.e., the combined micro and macro fracture mechanics, has been established and applied

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to brittle fracture and fatigue crack growth under static and atmospheric conditions.\(^3,4\)

In this paper, not only the researches under static and atmospheric conditions but also those under dynamic and environmental conditions dominated by cyclic and time-dependent fracture mechanisms, such as fatigue crack growth, hydrogen embrittlement and high-temperature creep were noticed from the viewpoint of particle transport phenomena in materials,\(^1,4\) such as dislocation group dynamics, hydrogen diffusion, and vacancy diffusion. On the basis of this concept and our proposed mesoscale mechanics, new mechanical indexes are derived as dominant factors for predicting fatigue crack growth life\(^1,5,6\) and the time-dependent fracture life of notched specimens subjected to hydrogen embrittlement\(^1,4\) and high-temperature creep.\(^1,4\)

Furthermore, as the blood vessel wall is a viscoelastic material dominated by a time-dependent mechanism (creep mechanism) and is influenced by cyclic pulsatile blood pressure (fatigue effect), it is considered closely related to both fatigue and creep components. As a practical application, our theories were applied for estimating the degree of viscoelastic deterioration and chaotic motion of blood vessel walls, which are closely related to blood vessel diseases such as atherosclerosis and aneurysm, and new indexes to diagnose them were proposed for clinical application.\(^1,7\)

2. Characterization of fatigue crack growth rate (FCGR) based on dislocation group dynamics and theory of thermally activated process

In this section, the theory of particle transport phenomena in non-steady dislocation group emission and motion, which is related to fatigue crack growth, is discussed. The formulation of FCGR is important for predicting fatigue fracture life in transportation equipment such as aircraft, automobiles, and railway and engineering structures.

Under this situation, research on fatigue crack growth was promoted and Paris’ law, which is the experimental law for FCGR \((da/dN)\) during the stage of stable crack growth, was proposed\(^6\) to predict crack growth life as given by Eq. [1] and shown in Fig. 1.\(^1,8\)

The algorithm for predicting of fatigue crack growth life based on Paris’ law is as follows:

\[
\frac{da}{dN} = A\Delta K^n. \tag{1}
\]

By integrating Eq. [1], the crack growth life in the region of stable FCGR \((da/dN)\), \(N_f\), is calculated, which covers the major portion of total fatigue crack growth life as given by Eq. [2].

\[
N_f = \int_0^{N_f} dN = \int_{a_0}^{a_f} \frac{da}{A(\alpha \sqrt{\pi a \Delta \sigma})^n}, \tag{2}
\]

where \(A\): constant; \(da/dN\): fatigue crack growth rate (FCGR), \(N_f\): life of stable fatigue crack growth, \(N\): fatigue load cycle, \(a\): crack length, \(\Delta \sigma\): normal stress amplitude, \(n\): experimentally obtained material constant, \(\Delta K = \alpha \sqrt{\pi a \Delta \sigma}\): value of stress intensity factor amplitude, \(a_0\): initial crack length, \(a_f\): final crack length, and \(\alpha\): coefficient dominated by the dimension of specimen.

However, on the basis of Eqs. [1] and [2], fatigue failure life can be experimentally predicted by deriving an experimental equation for FCGR given by Eq. [1], many theoretical studies also have been conducted to clarify the mechanism of fatigue crack growth.\(^9\)–\(^21\) These studies have focused on the striation mechanism of FCGR\(^22\) in the region of stable crack growth, which accounts for the major part of crack growth life. Because striation mechanism of FCGR is considered to be dominated by local plastic deformation near a crack tip, it is necessary to conduct an analysis for deriving the law of the number of dislocations emitted from a crack tip during the fatigue loading process based on the dislocation group dynamics emitted from a stressed source.
In the early stage of dislocation group mechanics research, static mechanical analyses were mainly conducted, and research on dislocation dynamics only considered an isolated dislocation (single dislocation). In the 1970s, pioneering studies on dislocation group mechanics were conducted using the model shown in Fig. 2. The basic equation of dislocation motion is given by Eq. [4]:

\[ v_{iso} = v_0 \left( \frac{\tau}{\tau_0^0} \right)^m = M \tau^m, \]  

\[ V_i = \frac{dx_i}{dt} = M^* \tau_{eff,i}^m, \]

where \( \tau_{eff,i}^m \) is the effective shear stress exerted on the dislocation at position \( x_i \), \( \tau \): applied shear stress rate, \( \tau_0^0 \): shear stress which moves an isolated dislocation with the velocity of 1.0 cm/s, \( M, M^* \): material constants, \( v_0 \): 1 cm/s, \( x_i \), \( v_i \): position and velocity, respectively, of an individual dislocation in the dislocation group array, \( t \): time of stress application, and \( x_{iso}, v_{iso} \): position and velocity of an isolated dislocation (single dislocation), respectively.

By conducting these analyses, a similarity law for the number of dislocations and the positions of the individual dislocations in the dislocation array emitted from a stressed source was determined, which is dominated by the similarity law parameter \( \eta \) as given by Eqs. [5a], [5b], and [5c]:

\[ n = A^* \eta^{n+1}, \]  

\[ \frac{x_i}{x_{iso}} = f_i(\eta), \]  

\[ \eta = \tau^m / \tau_{eff,i} \]

where \( n \) is the number of dislocations emitted from a stressed source, \( \eta \) is the similarity law parameter composed of shear stress rate \( \tau \), stress application time \( t \), and material constant \( m \), which is named the “dynamic factor” by authors.

Combining the equation for the number of dislocations emitted from a stressed source given by Eq. [5a] with the experimental and theoretical velocity equations of an isolated dislocation given by Eqs. [6]–[8], the number of dislocations emitted from a stressed source is given by Eq. [9a] and the theoretical equation of FCGR dominated by a stress-dependent thermally activated process given by Eq. [9b] were derived. Equation [9b] is the theoretical dynamic formulation of FCGR, which corresponds to Paris’ law. Furthermore, Eq. [9b] includes the dependence not only on the amplitude of the stress intensity factor \( \Delta K \) but also on the temperature \( T \) and load frequency \( f \) of FCGR:

\[ v = v_0 \left( \frac{\tau}{\tau_0^0} \right)^m = v_0 \exp \left( -\frac{H(\tau)}{kT} \right), \]

\[ H(\tau) = H_k \left( 1 + \frac{1}{4} \log \frac{16\tau_p}{\pi \tau} \right), \]

\[ \tau_0 \approx \tau_p \exp \left( \frac{16}{\pi} \right) \]

\[ n = A(4f)^{\frac{m+1}{2}} \frac{(\Delta K)}{\sqrt{\pi \mu}} \left( \frac{\Delta K}{\Delta K_1} \right)^{\frac{n+1}{2}} \]

\[ \frac{da}{dN} = bA(4f)^{\frac{m+1}{2}} \frac{(\Delta K)}{\sqrt{\pi \mu}} \left( \frac{\Delta K}{\Delta K_1} \right)^{\frac{n+1}{2}} \]

where \( A \) is a constant, and \( b \) is the Burgers vector, \( \tau_p \) and \( H_k \) denote the Peierls stress and kink energy that are related to the motion of an isolated dislocation, respectively, \( \varepsilon \) is the local size of the region on...
which local stress is exerted around a tip of the crack. Here, $\tau$ is replaced by a local normal stress averaged by $\varepsilon$ given by $\Delta K_1/\sqrt{\varepsilon}$, where $\Delta K_1 = \sqrt{(\pi \alpha)} \times \sigma$.

Equation [9b] is formally written as Eq. [10],15),16) which gives the FCGR derived from dislocation group dynamics with emission from a stressed source (striation mechanism).15),16) Moreover, the FCGR dominated by nucleation mechanism of voids caused by vacancy diffusion was previously derived using Eq. [11] based on nucleation theory.30)

The experimental relationship between logarithmic values of FCGR in the second region of stable fatigue crack growth and the inverse of temperature for increasing values of $\Delta K$ (increase in crack length) in the second region of stable FCGR for aluminum alloys,31) is thermally activated and can be described by Eqs. [10] and [11]. The dynamic factor $\eta$ is a key parameter in the formulation of FCGR.

Furthermore, the threshold stress intensity factor based on dislocation group dynamics,34) which is the critical condition for the start of fatigue crack growth, has also been formulated named as Yokobori’s model.35)

3. Quantitative characterization of the sensitivity of hydrogen embrittlement based on
hydrogen diffusion

In this section, the theory of particle transport phenomena of local stress-driven hydrogen diffusion, which is related to hydrogen embrittlement, is discussed.

Concerning the development of hydrogen energy utilization technologies such as fuel cell vehicles and hydrogen stand infrastructure, the establishment of sophisticated prevention technology against hydrogen embrittlement is needed, and owing to the difficulty of experimental observation of hydrogen diffusion behaviors, it is essential to analyze the diffusion behaviors of hydrogen that flows into structures and concentrates under stress application. For this problem, the introduction of a coefficient $\alpha$ in the stress-driven term of hydrogen diffusion equation, which actualizes the effect of stress on hydrogen diffusion behavior, that is, the $\alpha$ multiplication method14),36) has been proposed. Furthermore, authors have proposed the FEM–FDM method, in which stress analysis is conducted by the finite element method (FEM) and hydrogen diffusion analysis is conducted by the finite difference method (FDM), which are suitable for each analysis.14),37)

Based on these methods, the theoretical equation for predicting the site of hydrogen concentration and the sensitivity of hydrogen embrittlement in engineering structures was proposed for the first time, which is a meso and macroscale analysis.14),36),37) Other studies on the analysis of hydrogen diffusion were mainly concerned with the clarification
of the hydrogen embrittlement mechanism, which is a microscale problem in materials science. Therefore, our proposed method is distinguished from others in terms of the research of responsible safety maintenance of hydrogen energy infrastructure.

Stress-induced hydrogen diffusion is a mechanism elucidated by the model that the lattice distance was expanded by the hydrostatic stress at the site of maximum hydrostatic stress, such as around the notch tips, caused by the existence of the site of stress concentration. Hydrogen enters such sites via the mechanism shown in Fig. 4. A governing equation has been previously proposed as a basic equation. However, at that time, non-numerical methods for the derivation of an analytical solution were proposed, and the specification of \( \alpha_i \) in the linear combination of each term of the hydrogen flux \( J \) and hydrogen diffusion equations, given by Eqs. [12] and [13], respectively, were taken as unity. However, as shown in Eqs. [14] and [15], essentially, diffusion-driven forces due to the gradient of a potential such as those of particle diffusion concentration and local stress, \( \nabla \phi_1 \) and \( \nabla \phi_2 \), are different from each other. Therefore, the diffusion coefficient is not given by Eq. [16] but by Eq. [17], which includes the effect of entropy \( S(\nabla \phi_1, \nabla \phi_2) \), on the diffusion coefficient. Therefore, flux \( J \) and the hydrogen diffusion equations should be given as a linear combination of \( \alpha_i \), that is, a coefficient of the entropy for each term, as given by Eqs. [12] and [13], respectively. This method is called the \( \alpha \) multiplication method. In an analysis using this method, typical time-sequential increasing behaviors of hydrogen concentration at the elastic–plastic boundary, where the hydrostatic stress takes the maximum value, were found to appear for high-strength steel with a yield stress of 1442 MPa, which is in good agreement with predicted experimental results. This method can also predict the site of hydrogen concentration, shown under the mechanical model in Fig. 5(a) and 5(b); in particular, hydrogen concentration appears as shown in Fig. 5(c); in particular, hydrogen concentration at the site of maximum hydrostatic stress does not occur even for a yield stress of 1442 MPa. These results confirm the validity of \( \alpha_i \) in hydrogen concentration analyses.

\[
J = -\alpha_1 \frac{D}{RT} C\nabla \phi_1 - \alpha_2 \frac{D}{RT} C\nabla \phi_2, \quad [12]
\]

\[
\frac{\partial C}{\partial t} = -\nabla J = \alpha_1 D\nabla^2 C - \alpha_2 \frac{D\Delta V}{RT} \nabla(C\nabla \sigma_P), \quad [13]
\]

\[
\phi_1 = RT \ln C, \quad [14]
\]

\[
\phi_2 = -\sigma_P \Delta \sigma_P, \quad [15]
\]

where \( C \) is the hydrogen concentration; \( R \) is the gas constant; \( T \) is the absolute temperature; \( \sigma_P(=\frac{\sigma_{xx}+\sigma_{yy}+\sigma_{zz}}{3}) \) is the hydrostatic stress; \( \sigma_{xx}, \sigma_{yy}, \) and \( \sigma_{zz} \) are the normal stress components along the \( x, y, \) and \( z \) axes, respectively; \( \phi_1 \) is the chemical potential induced by the hydrogen concentration; \( \phi_2 \) is the potential induced by local hydrostatic stress; \( \Delta V \) is the volume change of the crystal lattice due to hydrogen entry under \( \sigma_P \); \( H, G, \) and \( S \) denote the enthalpy, Gibbs free energy, and entropy, respectively, and \( D_0 \) is a constant with the dimensions of diffusion coefficient. The diffusion coefficients \( D \) and \( D_1 \) are respectively given by

\[
D = D_0 e^{-\frac{H}{RT}}, \quad [16]
\]

\[
D_1 = D_0 e^{-\frac{G}{RT}} = D_0 e^{-\frac{H-TS}{RT}}, \quad [17]
\]

where \( \nabla \phi_1 \) is a driving force induced by the gradient of a chemical potential related to hydrogen concentration, \( \nabla \phi_2 \) is a driving force induced by the gradient of local hydrostatic stress, and \( \alpha_i \) (\( i = 1, 2 \)) is a coefficient of \( \alpha \) multiplication concept.
Using this method, the indicator of the hydrogen concentration rate of steel, that is, the HE parameter, which corresponds to the inverse of time when the hydrogen concentration at the site of maximum hydrostatic stress reaches some critical value, is given by Eq. [18]. This result revealed and specified the temperature range for the peak hydrogen concentration. This result is shown in Fig. 6. It means hydrogen embrittlement is typical at this peak region. It predicts that the degree of hydrogen embrittlement occurs at a peak temperature value of around 190 K, which is in good agreement with the experimental results of SUS304L (190 K). This result shows that the HE parameter given by Eq. [18] well predicts the experimental sensitivity of hydrogen embrittlement:

$$HE = \left( \frac{D}{b^2} \right) \left( \frac{\sigma_{ys}}{E} \right)^6 \left( \frac{K}{E \sqrt{b}} \right)^{-4.5}$$

where $\sigma_{ys}$ is the yield stress, $b$ is Burger’s vector (3 $\times$ 10^{-10} m), $E$ is Young’s modulus, $D$ is the diffusion coefficient, and $K$ is the stress intensity factor.

This method can be used to determine the conditions under which hydrogen can be mechanically locked and concentrated around a notch tip. Based on this method, a conventional hydrogen embrittlement FCGR test method using mechanically locked hydrogen-charged small specimens under atmospheric conditions was proposed, which is important in FCGR experiments. In this method, hydrogen charging is conducted under a sustained load application to a notched specimen, and using this hydrogen-charged small specimen, FCGR tests are conducted under atmospheric conditions. As hydrogen molecules are very small, they are easily diffused and released from metals into the...
environment. Therefore, to maintain the hydrogen in the metal after hydrogen charging, a large-C(T) (Compact tension) specimen has been used or hydrogen-charging condition has been maintained during FCGR experiments, which present difficulties in conducting FCGR experiments under hydrogen environment conditions. The size of a large-C(T) specimen (width = 222.25 mm and thickness = 88.9 mm) is 3.5 times that of a standard-C(T) specimen. In our proposed method with mechanical locking, the hydrogen-charged specimen is only 20.8-mm width and 0.8-mm thick. This method was verified as a convenient method for the quantitative estimation of the sensitivity of hydrogen embrittlement. Furthermore, experiments based on this method were successfully conducted. Using this test method, FCGR experiments using hydrogen-charged small specimens for electromagnetic stainless steels were conducted under atmospheric conditions. The algorithm for calculating the acceleration ratio of FCGR under the hydrogen-charged condition to that under atmospheric conditions is shown in Fig. 7:

\[
D^* = \frac{da/dN|_{Active} - da/dN|_{Inactive}}{da/dN|_{Inactive}}
\]

where \(D^*\) is the acceleration ratio of FCGR under hydrogen-charged conditions to that under atmospheric conditions.

As shown in Fig. 7, the area enclosed by \(\log D^*\), \(\log \Delta K_1\), and \(\log \Delta K_2\) is the average acceleration ratio \(\eta_1\):

\[
\eta_1 = \sum_{i=1}^{n} (\log D_i^* - \log D_{0i}) / \log (\Delta K_2/\Delta K_1)
\]

The ratio of stress increasing time to stress decreasing time is 1:4, that is, fast–slow fatigue stress wave form.
where \( \log D_i^* \) is the value of \( \log D^* \) at the segment, \( i \) of \( n \) between \( \log \Delta K_1 \) and \( \log \Delta K_2 \). \( \log D_0 \) is the lowest limit of \( \log D^* \) with a value of \( \log 10^{-2} \). For the case of \( \log D^* \) being less than \( \log 10^{-2} \), the FCGR under the environmental conditions is considered to be almost equal to that under atmospheric conditions.

Furthermore, the initial acceleration behavior of FCGR was also a dominant factor and was evaluated under hydrogen-charged conditions. The initial acceleration behavior decreases with increasing \( \Delta K \), as shown in Fig. 7. Therefore, the initial acceleration degree \( \eta_2 \) is given by

\[
\eta_2 = \frac{d\{\log(D^*)\}}{d\{\log(\Delta K)\}}. \tag{21}
\]

By taking account of both acceleration effects on FCGR, i.e., \( \eta_1 \) and \( \eta_2 \), the acceleration factor \( \overline{|r_{AH}|} \) of FCGR under hydrogen-charged conditions is given by\(^{46,47}\)

\[
\overline{|r_{AH}|} = \sqrt{\eta_1^2 + \eta_2^2}. \tag{22}
\]

Figure 8 shows the relationship between \( \overline{|r_{AH}|} \) and \( HE \)\(^{46,47,49}\) for various electromagnetic stainless steels. In materials with 13% Cr addition, \( \overline{|r_{AH}|} \) increased with \( HE \), but in materials containing more than 18% Cr or 0.2% Ti as an additive, \( \overline{|r_{AH}|} \) typically decreased with increasing \( HE \) (see Fig. 8 and Table 1), which was effective in improving hydrogen embrittlement resistance against FCGR. From the viewpoint of materials engineering, rich Cr content results in the formation of an oxide film on the metal surface, which prevents the inflow of hydrogen from the outer part,\(^{50}\) and 0.2% Ti additive has been considered to trap hydrogen and suppress its diffusion.\(^{51,52}\) Our results verified these considerations of materials engineers and confirmed the practical application of these materials to engineering devices.\(^ {47}\) In particular, in addition to the merits of specification of the site and conditions of hydrogen concentration in a structure, our method is unique in its ability to mechanically control hydrogen diffusion and concentration, i.e., estimating hydrogen embrittlement FCGR using a hydrogen-charged small specimen by mechanical locking of hydrogen under atmospheric conditions.

The \( \alpha_i \) coefficients and \( HE \) are key indicators for characterizing hydrogen concentration and diffusion behaviors related to hydrogen embrittlement.

### 4. Characterization of high-temperature creep deformation and crack growth based on the analysis of vacancy diffusion and the theory of thermally activated process

In this section, the theory of particle transport phenomena of vacancy diffusion, which are related to creep deformation and thermally activated process of crack creep growth, is discussed.

#### 4.1. Multiscale creep damage at stress concentration sites

Our research groups have conducted systematic studies on the crack initiation and growth under creep conditions.\(^ {53}\) These studies revealed that the mechanism of creep damage formation that contributes to the initiation and

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**Table 1. Mechanical properties and chemical compositions of electromagnetic stainless steels**\(^ {49}\)

| Materials | Chemical composition | 0.2% Y.S (MPa) | T.S. (MPa) |
|-----------|----------------------|---------------|-----------|
| K-M31\(^ {47}\) | 13Cr–2Si | 400 | 570 |
| K-M35FL\(^ {47}\) | 13Cr–1Si–0.3Al–1Pb | 245 | 370 |
| K-M38 | 18Cr–2Si–1Mo | 450 | 590 |
| K-M38CS | 18Cr–2Si–1Mo–0.2Ti | 450 | 590 |
| K-M45 | 20Cr–2Si–2Mo | 590 | 680 |
| K-M62F | 13Cr–0.9Al–0.3Mo–0.1Ti | 235 | 400 |

---

**Fig. 8.** Relationship between \( HE \) and \( \overline{|r_{AH}|} \). In materials with 18%, 20% Cr (K-M38, K-M45) or 0.2% Ti (K-M38CS) additives, the \( HE \) values are lower than in other materials, implying high \( HE \) resistance.\(^ {49}\)
growth of creep cracks is important from an engineering viewpoint. Furthermore, to conduct this research, it is necessary to conduct analyses and experiments based on multiscale mechanics, which link the nanoscale vacancies to macroscale cracks through voids and micro-cracks on the mesoscale. The relationships among the three scales are shown in Fig. 9. The nanoscale vacancies (atomic defects) are concentrated by the diffusion mechanism, which results in the formation of voids on the scale of microns at the site of the vacancy-concentrated region based on the nucleation mechanism. These voids originate along the crystal grain boundaries and coalesce, which results in grain boundary micro-cracks with a grain size of 10–100 nm. These cracks are formed mainly at the grain boundaries. An analysis of macroscopic damage mechanics is adopted in this region. The grain boundary micro-cracks coalesce and macroscale (mm-sized) cracks are formed on a scale of millimeters; these are treated by macroscopic mechanics such as fracture mechanics.

As the progression degree and scale of the damage region at the final fracture differ depending on creep ductility or brittleness of materials such as Ni-based super alloys and heat-resistant steels, multiscale damage mechanics is needed to establish a prediction method for creep fracture life in such materials. For example, for γ’ strengthening Ni-based super alloys, which are creep-brittleness materials, the scale of creep damage is of the order of microns at the final stage of fracture, and for Cr–Mo–V steel, which is a creep-ductile material, the scale of creep damage reached 500–800 μm in experiments using the same scale double-edge notched specimen with a width of 4 mm and a notch depth of 0.25 mm.

Furthermore, as nanoscale vacancy diffusion of creep is driven by hydrostatic stress, σ_P similar to the hydrogen diffusion of hydrogen embrittlement, the vacancy diffusion equation is also given by Eq. [13]. However, the material constants and values of α_i differ from those of hydrogen diffusion. Because the local stress triaxiality TF (given by Eq. [23]), that is, a macroscopic mechanical factor reflecting the effect of specimen dimensions on creep ductility, is dominated by hydrostatic stress σ_P which promotes the concentration of vacancies at the site of maximum hydrostatic stress, vacancy diffusion is also dominated by TF. Therefore, in the same material, a higher value of TF owing to specimen dimension causes creep brittle property by local vacancy concentration. This property is defined as structural brittleness.

\[ TF = \frac{3\sigma_P}{\sigma} \]  

where σ_P is the hydrostatic stress and \( \dot{\sigma} \) is the equivalent stress related to plastic deformation. \( \tau_{xy}, \tau_{yz}, \) and \( \tau_{zx} \) are the shear stress components respectively.

4.2 Analysis of micro-creep damage related to creep deformation. Figure 10 shows the results of the comparison between the area of vacancy concentration around the notch tip obtained by analyzing the vacancy diffusion given by Eq. [13] and that of creep damage composed of grain...
boundary micro-cracks observed by the proposed in situ creep-fatigue testing machine.\textsuperscript{64} The material was a TiAl intermetallic compound.\textsuperscript{62} Although the vacancies and boundary cracks were analyzed on the nanoscale and micron scale, which is the difference in scale between them as shown in Fig. 9, both concentrated areas showed similar behaviors, that is, behaviors of the path-selective growths and local concentrations around the notch tip as shown in Fig. 10.\textsuperscript{62} Therefore, the creep damage composed of micron scale grain boundary cracks were considered to originate from nanoscale vacancy concentrations, as shown in Fig. 9.

The vacancy concentrations obtained in the vacancy diffusion analysis were combined into a parameter of the damage progression degree, defined by $D_i$ and calculated by Eq. [24].\textsuperscript{61} The damage function $\omega$ defined in damage mechanics was newly formulated as Eq. [25].\textsuperscript{61} In previous damage mechanics,\textsuperscript{10,65} the damage function was calculated as the ratio of fracture life $t_f/t_0$ (where $t$ is the time of stress application and $t_f$ is the creep fracture life). However, damage function, $\omega$ is a dominant factor to derive creep strain as shown in Eqs. [27] and [28] which results in creep fracture life. In this expression, $t_f$ is unknown and it is included in the damage function. Therefore, it will not be possible to obtain creep strain by Eq. [28] and fracture life explicitly. On the other hand, using Eq. [25], $\omega$ can be theoretically derived based on the vacancy diffusion analysis and creep strain can be theoretically predicted using our proposed $\omega$ by Eqs. [27] and [28]. Creep fracture life can be predicted by creep strain, since creep fracture usually closely concerns the critical creep strain. In Eqs. [24] and [25], $C_i$ denotes the average vacancy concentration in element $i$ obtained by averaging the concentrations at the surrounding finite difference grids ($l$, $m$, $n$, and $o$ in Fig. 11\textsuperscript{13,61}), and $C_r$ is the critical vacancy concentration. The absolute value $|C_i - C_r|$ is considered an effective component of creep damage. $A_i$ is the area of element $i$ surrounded by finite difference grids, $l$, $m$, $n$, and $o$, which concerns $C_r$ as shown in Fig. 11. $D_i(t)$ is the value of $D_i$ at time $t$ (Fig. 12\textsuperscript{13,61}), and $D_f$ is the value of $D_i(t)$ when the time-sequential change in total damage area, $\Sigma A_i$, is saturated. In the vacancy diffusion analysis, the saturation time is detected as the time of fracture (Fig. 12). Substituting the newly defined parameter $D_i$ into the newly proposed micro-damage function (Eq. [25])\textsuperscript{13,61} and Eqs. [25] and [26] into Eq. [27], the creep rate is obtained as Eq. [28]:

\[
D_i(t) = \sum_i |C_i - C_r| A_i, \quad [24] \\
\omega = \frac{D_i}{D_f}, \quad [25] \\
X = \frac{\sigma_p}{\sigma}, \quad [26] \\
\delta^{C}_{ij} = \frac{\sigma}{\sigma_0(1 - \omega)} \left\{ \frac{1}{2} \left( 1 - \omega + \omega X^2 \right)^{(n-1)/2} \right\} \\
\times \left\{ \frac{3}{2} \left( 1 - \omega \right) S_{ij} + \omega X \delta_{ij} \right\}, \quad [27] \\
\varepsilon^{m+1} = \varepsilon^m + (\delta^{C}_{ij})^m \Delta t. \quad [28]
\]

$S_{ij}$ is deviatoric stress component which concern plastic deformation, that is, $S_{ii} = \sigma_u - \sigma_p$, $S_{ij} = \tau_{ij}$.
where $\delta_{ij}$ is the Kronecker delta function, $X = TF/3$, $\varepsilon$ and $\dot{\varepsilon}_{ij}$ are the creep strain and creep rate, respectively, $\Delta t$ is the time increment, and $m$ is the iteration number of calculations.

Figure 13 shows the creep deformation curves of the notched specimens of the fine-grained TiAl intermetallic compound (TiAl–Fe–V–B) at three temperatures (800 °C, 850 °C, and 900 °C) and their theoretical curves described by Eqs. [25]–[28]. The theoretical curves are in good agreement with the experimental curves, except in the final accelerated region.$^{61,62}$

In the above analysis, the prediction of creep deformation and its fracture life (which are macroscopic phenomena) was linked to an analysis of the local stress-driven vacancy diffusion (a microscale phenomenon) through damage mechanics.

4.3. Laws of creep crack growth rate (CCGR) and fracture life of notched specimens based on the $Q^*$ parameter. Our research groups have also conducted creep crack growth tests using various shapes of specimens for heat-resistant steels (9–12 Cr steels and stainless steels), Ni-based super alloys, and TiAl intermetallic alloy,$^{66-68}$ which are used in high-efficiency electric power plants and jet engines. The double-edge notched (DEN), C(T), circular notched round bar specimen (CNS),$^{69}$ and weld joint specimens had different TF values. In these studies, the $Q^*$ parameter given by Eqs. [29] and [30] was proposed as a predictor of the creep crack growth life.$^{67,70}$ This parameter is related to a local stress-dependent thermally activated process and its activation energy is related to that of vacancy diffusion,$^{13}$ as mentioned in subsection 4.2:

$$\frac{da}{dt} = A \exp(Q^*),$$

$$Q^* = -\frac{(Q - f(\sigma_l))/RT}{A \exp(Q^*)},$$

where $a$ is the crack length, $t$ is time, $da/dt$ is the creep crack growth rate (CCGR) (mm/h), $Q^*$ is the proposed dominating parameter of CCGR, named $Q^*$ parameter. $Q$ is the activation energy, which is related to the experimentally determined vacancy diffusion. $\sigma_l$ is the local elastic–plastic stress near the tip of a creep crack. $R$ is the gas constant. $T$ is the absolute temperature. $a_i$, $a_f$ is the initial and final crack length, respectively. $t_f$ is the fracture life related to creep crack growth.

However, to apply the law of creep crack growth life to actual engineering structures, it is necessary to consider the effects of the dimensions of an engineering structure (characterized by TF), differences in microstructure of material strengthening caused by different manufacturing processes such as different sizes of ingots depending on the scale of the engineering structure, and aging degradation on the CCGR. In this section, W-added high-Cr steels were used to study the effects of the abovementioned factors on CCGR and life as follows.$^{13}$

![Comparison of experimental and theoretical creep curves of notched TiAl–Fe–V–B specimens at different temperatures: (a) 800 °C (1073 K), (b) 850 °C (1123 K), and (c) 900 °C (1173 K). The theoretical curves were derived from the coupled analyses of the stress-driven vacancy diffusion and damage mechanics.$^{61,62}$](image)
4.3.1. The effect of local stress multi-axiality TF.

In subsection 4.1, using specimens of various shapes and TF values, creep deformation was found to be affected by the local stress multi-axiality TF (structural brittleness).\(^{63}\) The activation energy of CCGR in the formulation of \(Q^*\) given by Eq. [29] is also affected by TF, as shown in Fig. 14.\(^{71}\)

4.3.2. Scale effect of materials.

A large engineering structure requires ingots of large size during the manufacturing of materials. For such a case, in the cooling process of the heat treatment during the materials manufacturing process, the cooling rate differs between the outer and inner parts of the materials, which results in different material structures such as grain size \(d\) or space of the strengthening structure of material \(l\) (width of lath structure) as shown in Figs. 15 and 16 because these material structures are sensitive to the cooling rate. In W-added high-Cr steel, the width of the lath structure \(l\), that is, the strengthening structure and grain size \(d\) are dominated by this effect and the activation energy of creep deformation is changed by this effect. These results are shown in Fig. 15\(^{13,72}\) as a plot of activation energy versus lath width \(l\). Furthermore, the relationship between lath width and grain size was obtained, as shown in Fig. 16.\(^{13,73}\) These results reveal that the activation energy of CCGR is also affected by grain size (i.e., by the material structure).

4.3.3. Effect of aging (time-sequential degradation of materials under operation).

Aging effect in materials was reported to induce a reduction in fracture toughness owing to the precipitate\(^{74}\); however, our research groups found that the effect of aging through almost 30 years of operation on \(Q^*\) was not sensitive in the exponential term of \(Q^*\) in Eq. [32] and sensitive to the coefficient of \(A\) in Eq. [32]\(^{13,75}\).

On the basis of the effects shown in sub-subsections 4.3.1–4.3.3, the equations of CCGR and creep crack growth life are given by Eq. [32]\(^{13}\):

\[
\frac{1}{tf} \approx \frac{da}{dt} = A(TF, d, Aging) \exp(Q^*(TF, d)) \quad \text{[32]}
\]

Concerning W-added 9–12 Cr steel, by even considering the effect (as shown in sub-subsection 4.3.2) on predictive law of the life of creep crack growth given by Eq. [32], the scattering of creep crack growth life was found to decrease from 100% to 30%.\(^{13,74}\) That is, the correlation appears as shown...
in Figs. 14–16. It makes it possible to apply the predictive law of creep crack growth life to predict the actual creep crack growth life of engineering structure which includes effects of TF, scale effect of materials and aging on creep crack growth life.

The $Q'$ parameter has been admitted as an indicator for the prediction of creep crack growth life,70 and using the Eq. [32], the prediction of the remnant life of creep crack growth is possible. Furthermore, $Q'$ has been introduced in ASTM E1457-19e11) and ISO/TTA 5:2007, “Code of practice for creep/fatigue testing of cracked components”12 as a method for evaluating CCGR. Moreover, using stainless and high-Cr steels, the superiority of the $Q'$ was mentioned from the viewpoint of the estimation of CCGR and the prediction of creep crack growth life.77,78

5. Application to the nondestructive detection method for the progression degree of viscoelasticity (time-dependent property) of blood vessel walls

In this paper, I discussed fatigue and creep phenomena, which are dominated by cyclic and time-dependent mechanisms. The blood vessel wall is a viscoelastic material similarly dominated by a time-dependent mechanism (creep mechanism) and is influenced by cyclic pulsatile blood pressure (inducing a fatigue effect). Therefore, the mechanical deterioration of blood vessel walls includes both fatigue and creep components, which are cyclic and time-dependent mechanisms.

In this section, a method for detecting the effect of fatigue and creep on the mechanical deterioration of blood vessel walls based on our proposed fatigue and creep test method. The actual test methods are detailed below.

Experimental pulsatile flow was applied to the blood vessel in vitro, and the occurrence condition of the deterioration of the blood vessel wall was determined, as shown in Fig. 18.80

Condition (a) in Fig. 18 represents a constant pulsatile amplitude pressure flow of 0–294 mmHg, i.e., higher-pulsatile amplitude pressure flow (fatigue condition). Total pulsatile cycles are 20000 cycles.

Condition (b) represents a constant-amplitude pulsatile pressure flow of 147–294 mmHg, i.e., lower-amplitude pulsatile pressure flow with high minimum pressure involving the effect of time-dependent creep component into the fatigue condition caused by a decreasing pulsatile pressure amplitude. The total number of pulsatile cycles is 20000.

Condition (c) represents a constant pressure with 294 mmHg (creep condition). The pressure application time is the same as that in conditions (a) and (c).

Condition (d) represents a fluctuating pulsatile pressure flow; the maximum pressure of 294 mmHg was divided into six stages of pressure increments as follows: $\Delta P_1 = 110 \text{ mmHg}$ (2290 cycles, the first stage), $\Delta P_2 = 147 \text{ mmHg}$ (260 cycles), $\Delta P_3 = 184 \text{ mmHg}$ (260 cycles), $\Delta P_4 = 220 \text{ mmHg}$ (260 cycles), $\Delta P_5 = 258 \text{ mmHg}$ (260 cycles), $\Delta P_6 = 294 \text{ mmHg}$ (3332 cycles), $\Delta P_{T1} = 110 \text{ mmHg}$ ($\Delta P_1$ of step T1, 1250 cycles), $\Delta P_{T2} = 110 \text{ mmHg}$ ($\Delta P_1$ of step T2, 1250 cycles), where $P_{\text{min}} = 0 \text{ mmHg}.$
Condition (d) was also applied to the fatigue test in the metal as the variable-amplitude loading test to investigate the effect of overload on fatigue fracture life. However, condition (d) was modified to detect the effect of increasing blood pressure amplitude on the mechanical deterioration of the blood vessel wall. This condition is named the fluctuating pulsatile pressure flow condition (Fig. 18).

In Fig. 18, \( R \) is diameter of the blood vessel wall during the stage of maximum pulsatile blood pressure (294 mmHg) in the specified pulsatile blood pressure cycle and \( R_0 \) is initial diameter of the blood vessel wall. \( R/R_0 \) is the non-dimensional diameter of the blood vessel wall normalized by \( R_0 \). \( (R/R_0)_{\text{initial}} \) is the non-dimensional diameter during the first pulsatile cycle.

Values of \( R \) were obtained from the \( P-R \) curve obtained from static pressure increment \( \Delta P \) (ranging from 0–294 mmHg) by interruption of pulsatile test at each specified time of the pulsatile pressure application. After measuring \( P-R \) curve, pulsatile test was continued again.

Figure 18 shows the pulsatile cycle sequential characteristics of the deformation behavior of the blood vessel wall, including inelastic deformation, which is related to the viscoelastic property of the blood vessel wall.

Under condition (a) (i.e., the fatigue condition), the diameter of the blood vessel wall did not increase but slightly shrank against the pulsatile pressure cycles. Under condition (b), the diameter of the blood vessel wall increased against the pulsatile pressure cycles, implying that inelastic deformation is caused by low-amplitude pulsatile pressure related to the creep effect. Condition (c) (i.e., the pure creep condition (constant pressure)) showed pulsatile sequential characteristics of inelastic deformation of the diameter of the blood vessel wall similar to that under condition (b); however, the value itself is not so large compared with that under the condition (b). Under condition (d) (i.e., the fluctuating pulsatile pressure flow), typical inelastic deformation of the blood vessel wall was caused. This behavior typically appears during the first increasing process of maximum pulsatile pressure (stages of \( \Delta P_2-\Delta P_3 \)). In this process, a time-dependent mechanism, i.e., the creep effect, is considered to be involved in each pressure-increase step.

We have previously conducted similar pulsatile pressure tests with the pulsatile pressure conditions of Fig. 18 and measured the tensile strength of blood vessel walls after pulsation tests. The corresponding results also showed that the strength of blood vessel walls slightly increased compared with that of the control under condition (a), i.e., high-amplitude pulsatile pressure (fatigue), and it was also found to typically decrease under condition (d), i.e., under the fluctuating pulsatile pressure flow condition.

From the abovementioned results, the strength of the blood vessel wall was found not to decrease...
under the constant high-amplitude pulsatile flow (fatigue), i.e., the effect of pulsatile pressure amplitude on the blood vessel wall prevented the progress of inelastic deformation of the blood vessel wall, which is related to the viscoelastic property of the blood vessel wall, even though the maximum pressure is higher (e.g., 294 mmHg).

This behavior is very different from that of metal materials, since as shown in Fig. 1, FCGR is accelerated by application of stress amplitude in metals. This property distinguishes blood vessel walls from metal materials.

Inelastic deformations of the blood vessel wall were observed under a low-amplitude pulsatile flow (condition (b) of Fig. 18) and a fluctuating-pressure pulsatile flow (condition (d) of Fig. 18). These promoted the viscoelastic property of the blood vessel wall by creep and pressure increases. 

For the case of metals, many studies on fatigue under variable amplitude loading have been conducted; however, it has not yet been clarified whether this effect shortens fatigue life, especially the effect of overload on fatigue life. This is because fatigue life is not only dominated by plastic deformation (inelastic deformation) but also by fatigue crack growth in metals. Plastic deformation sometimes retards the fatigue crack growth, implying a competitive interplay between crack growth and plastic deformation.

However, unlike metals, blood vessel walls can be mechanically deteriorated by viscoelastic deformations alone.

According to the abovementioned results, preventing the viscoelastic deformation of blood vessels caused by creep and increasing pressure would dominantly prevent mechanical deterioration of blood vessel walls.

5.2. A noninvasive diagnostic method for blood vessel disease.

Based on the results mentioned in the previous subsection, we derived an estimation index of the progression degree of viscoelasticity named the $I'$ parameter that enables the noninvasive measurement of the acceleration response of the blood vessel wall during the systolic phase of the pulsatile period. We also proposed an algorithm for the noninvasive diagnosis of atherosclerosis from the quantified estimation of the viscoelasticity of the blood vessel wall including a detection method for aneurysm based on chaos theory using Doppler effect sensor.

Various methods for noninvasive diagnosis of atherosclerosis have been proposed. These methods are concerned with the elastic rigidity of the blood vessel wall (PWV, CAVI), the blood pressure drop across a length of the blood vessel (ABI), and shape of the blood vessel wall (IMT). However, these methods do not detect the progression degree of inelasticity of the blood vessel wall. Furthermore, the irregular pulsatile motions of blood vessel wall caused by an aneurysm have not been clarified. We have analyzed the accelerated response of blood vessel walls during the systolic period. We have also observed the attractors of the periodic trajectories of pulsatile velocity of blood vessel walls. From the obtained results, the progression degree of viscoelasticity (mechanical deterioration) of the blood vessel wall by $I'$ and the irregularity degree of the pulsatile trajectory of velocity of blood vessel walls were quantified by $I'$ and entropy $S$, respectively.

Examples of the trajectory of pulsatile deformation rate of blood vessel walls obtained by the analysis of the attractor based on chaos theory are shown in Figs. 19(a)–(c). Because the trajectory of pulsatile deformation rate of blood vessel walls during the diastolic process is not enough to be well detected by a Doppler effect sensor, the trajectory during the systolic process is mainly reflected in the whole trajectories as shown in Figs. 19(a)–(c).

In this research, $X(t)$ is the measured deformation rate of the blood vessel wall at time $t$, and the time increment $\Delta t$ was 30 ms.

The trajectory in the normal blood vessel followed elliptical cycles (limit cycles), as shown in Fig. 19(a), which corresponds to the sharp response of the pulsating velocity waveform of the blood vessel wall. When the viscoelasticity of blood vessel wall which concerns with the atherosclerosis appears, its deformation response to the pulsatile pressure delays, which results in trajectories became distorted with a more large aspect ratio, that is, more flat trajectory, reflecting the blunt response waveform of the pulsating velocity of the blood vessel wall (Fig. 19(b)). Furthermore, in a blood vessel with an aneurysm, the peak region of the pulsatile waveform of the blood vessel wall exhibited two-phase behavior; correspondingly, small local circuit loops originated in the trajectory of pulsatile pressure rate of the blood vessel wall, as shown in Fig. 19(c).

Based on these results, as an indicator that represents the degree of disturbance of the trajectory of the pulsatile blood vessel waveform, the concept of entropy, which quantifies the disturbance of the
trajectory, was introduced. The results of clinical measurements represented by the two-dimensional map on the relationship between $I^*$ and entropy $S$ is shown in Fig. 20.

Entropy is defined by Eq. [33]:

$$ S = - \sum_{i=1}^{N} P_i \log P_i, $$

where $N$ is the number of divisions of the area that encloses the trajectories of the pulsating rate of the blood vessel wall and $P_i$ is the probability of dropping in the $i$th area of the trajectory of pulsating rate of the blood vessel wall.

These results show that the deterioration of elastic property of the blood vessel wall (the progression degree of viscoelasticity of the blood vessel wall) is related to atherosclerosis, and the disturbance of the pulsatile motion of the blood vessel wall related to the existence of aneurysm was detected by the two-dimensional map of $I^*$ and entropy $S$ in 15 s by a noninvasive diagnostic apparatus shown in Fig. 20.

This apparatus was approved as a medical device from September 2004 until March 2009 by the Ministry of Health, Labour, and Welfare of Japan.
(TRY-1, Taiyo Denshi Co. Ltd., Miyagi, Japan, Medical device manufacturing approval number: 21600BZX00440000). \( I^* \), which is an index of viscoelastic deterioration of the vascular wall, is measured by a simple, noninvasive, and feasible technique in clinical settings and is an effective predictor of coronary artery disease, as reported in a clinical paper in 2015. Furthermore, based on chaos theory, a noninvasive diagnostic tool for aneurysm was proposed and validated in artificial blood vessels by analysis of aneurysm characteristics, computational fluid dynamics, and clinical measurements.

6. Summary

Based on the theory of particle transport phenomena such as dislocation group dynamics, hydrogen diffusion, and vacancy diffusion, in this study, mesoscale mechanics that links the microscopic physics and the macroscopic mechanics of metals was established and the theory of quantitative prediction of the life of the cycle and time-dependent fracture such as fatigue, hydrogen embrittlement, and high-temperature creep were derived for practical applications.

Furthermore, this concept was applied to estimate the progression degree of the time-dependent viscoelastic property of blood vessel walls and existence of aneurysm. The dominant indicators for predicting the deterioration of strength of materials and their fracture life, which are relevant to both the safety maintenance of engineering structures and the health of blood vessels, were proposed as follows:

1) By analyzing dislocation group dynamics, the dynamic factor \( \eta \) was derived as an indicator of the number of dislocations emitted from a stressed source. Using this indicator, a theoretical formulation of FCGR was established.

2) By analyzing hydrogen diffusion, the multiplication coefficient \( \alpha \) and \( HE \) parameter were derived as indicators of the degree of hydrogen embrittlement. Using these indicators, the hydrogen concentration distribution around the site of stress concentration was estimated, and the sensitivity of hydrogen embrittlement of materials was quantified.

3) Through the \( \alpha \) multiplication method, the vacancy diffusion behavior was linked to creep deformation. Furthermore, the \( Q^* \) parameter, which characterizes the CCGR rate and predicts creep crack growth life, was proposed for practical applications. \( Q^* \) is controlled by a thermally activated process and its activation energy is dominated by vacancy diffusion. Furthermore, it is affected by local stress multi-axiality \( TF \), scale effect of materials, and aging.

4) The effect of fatigue and creep on the mechanical deterioration of blood vessel walls was estimated in fatigue and creep tests. Preventing the viscoelastic deformation of blood vessel caused by creep and increase in maximum blood pressure is a dominant factor in the prevention of mechanical deterioration of the blood vessel wall. Furthermore, a clinical application of these results in the noninvasive diagnostic method for atherosclerosis and aneurysm was developed and the \( I^* \) parameter and entropy \( S \) were derived for estimating the viscoelasticity progression degree of blood vessel walls and for predicting aneurysms.

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Profile

A. Toshimitsu Yokobori, Jr. was born in Tokyo in 1951 and graduated from Tohoku University’s Faculty of Engineering in 1973. He majored in research on strength of materials, especially in cycle- and time-dependent fracture mechanisms of materials. He received the Doctor of Engineering degree from Tohoku University in 1978. He worked as a research associate and an associate professor at Tohoku University and became Professor of the Faculty of Engineering at Tohoku University in 2001. He became Emeritus Professor at Tohoku University and Guest Professor at Teikyo University in 2016. In 2017, He became Specially-appointed Professor of Strategic Innovation and Research Center at Teikyo University. He was elected as Honorary Fellow of the International Congress on Fracture in 2013 and as the honorary member of The Japan Society of Mechanical Engineers. He was appointed the section chairperson of Mechanical Test Method of Cardio Vascular Implant and Materials, ASTM F4.04 10, from 1987 to 1993. He has conducted innovative research on the mechanism of cycle- and time-dependent fractures, such as those caused by fatigue, hydrogen embrittlement, corrosion fatigue, creep, and creep–fatigue interaction in metals. As a practical application of this work, he has established a technology that protects against fracture. He also applied his own theory to research on biomaterials and tissue materials such as blood vessels. Furthermore, based on viscoelasticity and chaos theory, he has proposed an innovative noninvasive diagnostic method for the progression degree of atherosclerosis and aneurysm. For his accomplishment, he received awards such as the Achievement Award of the Department of Material Mechanics in Japan Society of Mechanical Engineers, the Japan Academy Prize, and the ICF Takeo Yokobori Gold Medal (International Congress on Fracture).