Long-Term Stability of Residual Stress Improvement by Water Jet Peening Considering Working Processes

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To prevent primary water stress corrosion cracking (PWSCC), water jet peening (WJP) has been used on the welds of Ni-based alloys in pressurized water reactors (PWRs). Before WJP, the welds are machined and buffed in order to conduct a penetrant test (PT) to verify the weld qualities to access, and microstructure evolution takes place in the target area due to the severe plastic deformation. The compressive residual stresses induced by WJP might be unstable under elevated temperatures because of the high dislocation density in the compressive stress layer. Therefore, the stability of the compressive residual stresses caused by WJP was investigated during long-term operation by considering the microstructure evolution due to the working processes. The following conclusions were made: The compressive residual stresses were slightly relaxed in the surface layers of the thermally aged specimens. There were no differences in the magnitude of the relaxation based on temperature or time. The compressive residual stresses induced by WJP were confirmed to remain stable under elevated temperatures. The stress relaxation at the surface followed the Johnson–Mehl equation, which states that stress relaxation can occur due to the recovery of severe plastic strain, since the estimated activation energy agrees very well with the self-diffusion energy for Ni. By utilizing the additivity rule, it was indicated that stress relaxation due to recovery is completed during the startup process. It was proposed that the long-term stability of WJP under elevated temperatures must be assessed based on compressive stresses with respect to the yield stress. Thermal elastic–plastic creep analysis was performed to predict the effect of creep strain. After 100 yr of simulated continuous operation at 80% capacity, there was little change in the WJP compressive stresses under an actual operating temperature of 623 K. Therefore, the long-term stability of WJP during actual operation was analytically predicted.

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Introduction

In recent years, the demand for nuclear power generation that can reduce CO₂ emissions has been growing. However, PWSCC, which occurs in the weld metal of Alloy 600, causes degradation in PWRs [1,2]. The environment, the material, and especially the tensile residual stresses due to welding are prerequisite to the cracking. To prevent PWSCC, WJP, in which tensile residual stresses due to welding can be changed into compressive stresses, is used on the welds of Alloy 600 in the actual equipment [2]. In order to conduct a PT to verify the weld qualities before WJP, the welds were machined and buffed to produce a smooth surface. Microstructure evolution then takes place in the worked area due to the severe plastic deformations caused by these processes. After that, WJP can be used on the worked area, which retains severe plastic strains. However, the compressive residual stresses induced by WJP could be unstable under elevated temperatures because of the high dislocation density in the compressive stress layer. In contrast, long-term operation has a risk of stress relaxa-

tion through the development of creep strains since PWRs are regularly operated at temperatures of about 573 K, and the maximum operating temperature is approximately 623 K. To evaluate the stability of the compressive stress layer in the actual equipment, the behavior under elevated temperatures must be clarified by considering the superposition of the working processes (i.e., welding, machining, buffing, and WJP). In this work, the long-term stability of the residual stress improvement by WJP was investigated by considering the microstructure evolution due to the working processes.

Experimental Procedure

Materials. The chemical composition of the type 600 Ni-based alloy tested is (in weight %) 0.009C-0.16Si-0.14Mn-73.98Ni-16.59Cr-6.45Fe-0.02Cu-2.34Nb-0.05Co. As shown in Fig. 1, C-shaped ring specimens were employed to simulate the welding residual stress. The specimens had a diameter of 40 mm and a thickness of 5 mm. To reproduce the tensile residual stresses due to welding, the specimens were swaged and were continuously welded at the following conditions: a welding current of 60 A, an arc length of 7 mm, a welding speed of 0.2 m/min, and a wire feed rate of 1 mm/min. Argon was used as the shielding gas at a flow rate of 10 l/min. To generate an alteration layer by severe

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Fig. 1 Schematics of C-shaped ring specimen proposed to simulate weld residual stress

plastic deformation, the surface of the specimen was turned by a lathe instead of buffed. The working processes for the actual equipment were recreated in the two following steps.

Water Jet Peening. WJP was performed at standard conditions to reproduce the controlling procedure of welding residual stress, as expressed in Fig. 2. WJP is a residual stress reduction method that utilizes the collapse of cavitation bubbles on the surface [2]. As the high-speed water jet is ejected, a strong shear force is produced due to the strong vortex flow and turbulence. The pressure in that region drops to the level of the vapor pressure, which causes cavitation bubbles. When they move downstream of the jet, the pressure recovers to the normal level, and the cavitation bubbles shrink rapidly. This phenomenon is referred to as the collapse of cavitation bubbles. The collapse generates an extremely large shockwave of about 1 GPa and can change the residual stresses in the welds by creating compressive plastic deformation.

Thermal Aging. The thermal aging conditions are shown in Table 1. In order to evaluate the stress relaxation behavior during long-term simulation at a temperature higher than the operating temperature range, accelerated thermal aging was conducted at 773 K for 800 h. Additionally, aging at 623 K for 24 h was also conducted to evaluate the stress state at the time of maximum operating temperature during startup. A pure argon atmosphere was used because of the light water environment in the pressure vessel. To reflect the behavior that is expected based on the aging results, the tests were conducted at lower temperatures and for shorter durations using a liquid tin bath.

Fig. 2 Mechanism of residual stress improvement by water jet peening

X-Ray Stress Measurement. Since materials such as Ni-based alloys have large grains, the measurement of residual stress by X-ray diffraction is known to be difficult. Due to the influence of the large grains, Laue spots corresponding to the equator are not always detected with zero- or one-dimensional detectors. Therefore, it is necessary to establish a measurement method that can accurately evaluate the scattering angle (i.e., elastic strain). To exclude the influence of the large grains, the authors have proposed a method of recording the diffraction peaks using a two-dimensional detector in combination with multi-axial rocking prior to stress analysis [3]. A schematic of the diffraction system geometry is shown in Fig. 3. Rocking of the incident angle, ω , by $\pm 4^\circ$ was performed at the junction of the reciprocal lattice points and the Ewald sphere because the Laue condition is satisfied by the specific energy, as shown by either $K\alpha_1$ or $K\alpha_2$ due to the preferred orientation and grain growth. Additionally, in the axis direction of the ring specimen, X , rocking of $\pm 7^\circ$ was also performed by taking advantage of its characteristic feature: a constant distribution. Moreover, the γ -direction of the two-dimensional detector corresponds to the group of Laue spots around the incident beam (i.e., multi-axial rocking).

The stress measurement conditions are shown in Table 2. Additionally, in order to keep the gauge volumes constant during the measurement, the footprint of the specimen surface was limited to 7 mm in the axis direction and 1 mm in the hoop direction using a mask. The measured strain was converted into residual stress by using the stress model proposed by He [4]. Since the X-ray penetration depth in this system is very shallow, the analysis was performed in the plane stress state

$$\left(\frac{1}{2}S_2h_1^2 + S_1\right)\sigma_{11} + \left(\frac{1}{2}S_22h_1h_2\right)\sigma_{12} + \left(\frac{1}{2}S_2h_2^2 + S_1\right)\sigma_{22} + \frac{1-2\nu}{E}\sigma_{ph} = \ln\left(\frac{\sin\theta_0}{\sin\theta}\right) \quad (1)$$

Table 1 Conditions of thermal aging

Temperature/K	Aging time/h	Heating medium
623	24	Ar
773	800	Ar
Temperature/K	Aging time/min	Heating medium
553	1, 10, 20, 60	Liquid-tin
583	1, 10, 20, 60	Liquid-tin
623	1, 3, 5, 30	Liquid-tin

Fig. 3 Geometry of diffraction system for stress measurement

Here, h_1 and h_2 are

$$h_1 = \sin \theta (\sin \omega \sin \psi \sin \varphi - \cos \omega \cos \varphi) - \cos \theta \sin \gamma (\cos \omega \sin \psi \sin \varphi - \sin \omega \cos \varphi) + \cos \theta \cos \gamma \cos \psi \sin \varphi \quad (2)$$

$$h_2 = -\sin \theta (\sin \omega \sin \psi \cos \varphi - \cos \omega \sin \varphi) + \cos \theta \sin \gamma (\cos \omega \sin \psi \cos \varphi + \sin \omega \sin \varphi) - \cos \theta \cos \gamma \cos \psi \cos \varphi \quad (3)$$

where $1/2S_2$ and S_1 are the X-ray elastic constants, in which $1/2S_2 = (1 + \nu_{hkl})/E_{hkl}$ and $S_1 = -\nu_{hkl}/E_{hkl}$. Stress-free d-spacing is necessary in these calculations. For the plane stress state, σ_{33} can be assumed to be zero, and the stress can be calculated with an approximation of a stress-free d_0 or $2\theta_0$. Any errors will contribute only to a pseudohydrostatic term. The true d_0 corresponds to the intersection of the σ_{ph} line and the zero stress.

Residual stress was then measured along the depth direction using the surface removal method of electrical polishing with 5% perchlorate methanol. Since the lattice strain measured by X-ray diffraction differs from the mechanical strain, the stress analysis needs to consider the dependence on the diffraction plane {220}. The X-ray elastic constants were measured using the four-point bend test according to the following fundamental equation:

$$2\vartheta = -\frac{2(1 + \nu_{hkl})}{E_{hkl}} \tan \theta_0 \cdot \sigma_A \cdot \sin^2 \psi + \frac{2\nu_{hkl}}{E_{hkl}} \tan \theta_0 \cdot \sigma_A + 2\theta_0 \quad (4)$$

The slope in the 2θ - $\sin^2\psi$ diagram, M , is defined as follows:

$$M = -\frac{2(1 + \nu_{hkl})}{E_{hkl}} \tan \theta_0 \frac{180}{\pi} \sigma_A \quad (5)$$

That is, the X-ray elastic constants, $1/2S_2$ and S_1 , can then be estimated from the change of slope, M , and intercept, $2\theta_{\psi=0}$, which both bear linear relationships with stress

Table 2 Conditions of X-ray residual stress measurement

Power	36 kV 88 mA
Wavelength	0.2291 nm (CrK α)
Beam size	φ 1 mm
Detector	2D PSPC
Irradiated area	1 × 15 mm ² (with mask)
Diffraction plane	{220}
Measuring time	30 s/frame
Rocking	$\Delta\omega$ (± 4 deg) ΔX (± 7 mm)
Apparatus	D8 DISCOVER with GADDS

$$\frac{1}{2}S_2 = \frac{(1 + \nu_{hkl})}{E_{hkl}} = -\cot \theta_0 \left(\frac{\partial M}{\partial \sigma_A} \right) \frac{\pi}{180} \quad (6)$$

$$S_1 = -\frac{\nu_{hkl}}{E_{hkl}} = -\frac{\cot \theta_0}{2} \frac{\partial (2\theta_{\psi=0})}{\partial \sigma_A} \frac{\pi}{180} \quad (7)$$

Microstructure Observation. Electron backscatter diffraction (EBSD) was performed to confirm the metallurgical characteristics that affect stress relaxation. The observation field on the cross section of the specimen was mechanically polished and then electrically polished using 5% perchlorate methanol. The orientation was mapped with a step size of 0.3 μ m. In parallel, a Micro Vickers hardness measurement was taken at a load of 49 mN.

Thermal Elastic–Plastic Creep Finite Element (FE) Analysis. To study the influence of the creep strain on the long-term stability of the residual stress improvement by WJP, thermal elastic–plastic creep FE analysis was performed using the inherent strain. The FE mesh for the analysis is shown in Fig. 4 and contains 3951 nodes and 3104 elements. A half model of the specimen was used for the analysis due to symmetry. The thermophysical and thermomechanical properties of Alloy 600 as a function of temperature are shown in Fig. 5. The working processes included swaging, welding, WJP, and aging, and the simulation operations for each process are as follows:

- (1) To simulate the swaging of the ring specimen, the gap at the area (a) in Fig. 4, is forced closed.
- (2) To investigate the influence of the welded joint on the tensile stresses at the top surface, thermal elastic–plastic analysis is performed with a narrow heated zone adjacent to the gap, shown as the area (b) in Fig. 4.

Fig. 4 FE mesh and boundary conditions

Fig. 5 Material properties used in thermal elastic-plastic creep FE analysis

Table 3 Conditions of accelerated creep test

Temperature/K	Stress/MPa	Heating medium
723	100, 150, 200	Ar
773	100, 150, 200	Ar

Fig. 6 2θ - $\sin^2\Psi$ diagram under applied stress in Ni{220}

- After the welding simulation, a compressive stress distribution at the surface due to WJP is produced by inherent strain analysis. The area (c) in Fig. 4 is given tensile inherent strain in the hoop and axial directions and is given compressive inherent strain in the radial direction.
- Finally, thermal elastic-plastic creep analysis is performed to investigate the stress relaxation behavior at the surface during thermal aging. The Norton rule [5] shown in the following equation is used in this thermal elastic-plastic creep analysis:

$$\dot{\epsilon} = A\sigma^n \exp\left\{\frac{-Q}{RT}\right\} \quad (8)$$

where $\dot{\epsilon}$ is the creep strain rate, and σ is the equivalent stress. However, during the heating and cooling processes, the temperature is changed equally, and creep behavior is not considered; creep behavior is only considered in the holding process. For actual long-term operation, the holding temperature is 573 K. However, the creep parameters are extrapolated from the values at 773 K and 723 K obtained during the accelerated creep test shown in Table 3 since the holding temperature is too low to measure creep parameters.

Fig. 7 Changes in slope M with applied stress in Ni{220}

Fig. 8 Change of intercept $2\theta_{\psi=0}$ with applied stress in Ni{220}

Results and Discussion

Effect of Alloy Elements on X-Ray Elastic Constants. Figure 6 shows the 2θ versus $\sin^2\Psi$ diagram obtained using the four-point bend test. The regression lines obtained using the least-squares method for various stresses intersect at one point. Additionally, the change of its slope, M , and intercept, $2\theta_{\psi=0}$, bear linear relationships with stress, as shown in Figs. 7 and 8, respectively. The measured X-ray elastic constants are $1/2S_2 = 5.51 \times 10^{-3} \text{ GPa}^{-1}$ and $S_1 = -1.21 \times 10^{-3} \text{ GPa}^{-1}$. At the same time, the effect of the chemical composition on the X-ray elastic constants must be evaluated theoretically using the Kröner model [6] since the Ni-based Alloy 600 has a great compositional range. This model assumes that, in any anisotropic single crystal placed in a homogeneous matrix, the stress continuity is a boundary condition. In addition, the average stress and strain in each grain equals those of the surrounding homogeneous matrix. From the elastic stiffness, C_{ij} , of a single crystal of pure Ni, the X-ray elastic constants were estimated using the Kröner model, which was extended by Wit [7]. The elastic stiffnesses of the single crystal used [8] are $C_{11} = 250.8 \text{ GPa}$, $C_{12} = 150.0 \text{ GPa}$, and $C_{44} = 123.5 \text{ GPa}$. The isotropic shear modulus, G , is estimated by the following:

$$G^3 + \alpha G^2 + \beta G + \gamma = 0 \quad (9)$$

where the Kröner constants are given by

$$\alpha = \frac{9\kappa + 12[\mu_2 + 3(\mu_1 - \mu_2)\Gamma]}{8} - \frac{2\mu_1 + 3\mu_2}{5} \quad (10)$$

$$\beta = \frac{3\kappa[\mu_2 + 3(\mu_1 - \mu_2)\Gamma]}{4} - \frac{3(6\kappa\mu_1 + 9\kappa\mu_2 + 20\mu_1\mu_2)}{40} \quad (11)$$

Fig. 9 X-ray elastic constants calculated using the Kröner model with single-crystal constants and from experiment

Fig. 10 Residual stress improvement by WJP

$$\gamma = -\frac{3\kappa\mu_1\mu_2}{4} \quad (12)$$

and the bulk modulus, κ , and shear moduli are

$$\kappa = \frac{1}{3}(C_{11} + C_{12}) \quad (13)$$

$$\mu_1 = \frac{1}{2}(C_{11} + C_{12}) \quad (14)$$

$$\mu_2 = C_{44} \quad (15)$$

The orientation parameter, Γ , introduced by Bollenrath et al. [9] to describe the diffraction effects of a cubic polycrystal is determined by

$$\Gamma = \frac{h^2k^2 + k^2l^2 + l^2h^2}{(h^2 + k^2 + l^2)^2} \quad (16)$$

The X-ray elastic constants are calculated by

$$-S_1 = \frac{1}{6G} - \frac{1}{9\kappa} \quad (17)$$

$$\frac{1}{2}S_2 = \frac{1}{2G} \quad (18)$$

Figure 9 shows the X-ray elastic constants estimated using the Kröner model. The horizontal axis denotes the orientation parameter. In other words, the orientation parameter corresponds to the diffraction planes, and $\{220\}$ is at $\Gamma = 0.25$. The calculated X-ray elastic constants are $1/2S_2 = 5.38 \times 10^{-3} \text{ GPa}^{-1}$ and

Fig. 11 Comparison of residual stress distributions with correction

$S_1 = -1.19 \times 10^{-3} \text{ GPa}^{-1}$. That is, it was found that the X-ray elastic constants must be defined by a combination of the major element and the crystal structure if the material has low contents of light elements such as carbon. The following stress analyses were performed using measured the elastic constants.

Advantage of WJP Considering the Superposition of Working Processes. To evaluate the influence of the superpositioned working processes on residual stress, X-ray residual stress measurements were taken on an unpeened specimen (i.e., combination of buffing and welding) and a WJP specimen (i.e., combination of buffing, welding, and WJP), as shown in Fig. 10. Since the stress redistribution due to the removal of layers must be considered, the following corrections [10] were applied:

$$\sigma_\theta(z_1) = \sigma_{\theta_m}(z_1) + \left(\frac{(R - z_1)^2 + R_1^2}{(R - z_1)^2 - R_1^2} \right) \times \left\{ - \left(1 - \frac{R_1^2}{(R - z_1)^2} \right) \int_0^{z_1} \left(\frac{(R - z_1)^2}{(R - z_1)^2 - R_1^2} \right) \frac{\sigma_{\theta_m}(z)}{(R - z_1)} dz \right\} \quad (19)$$

where $\sigma_\theta(z_1)$ is the original stress in the hoop direction, $\sigma_{\theta_m}(z_1)$ is the measured stress in the hoop direction, R_1 is the inner radius of the ring specimen, R is the outer radius of the ring specimen, and z_1 is the removed depth. Figure 11 shows the corrected residual stress distribution. A maximum tensile stress of 550 MPa existed at the surface of the unpeened specimen, and the tensile stresses declined sharply until a depth of 30 μm ; after that, they gradually decreased toward the neutral axis. In contrast, compressive residual stresses were induced by WJP to a depth of over 800 μm , and the minimum value was -550 MPa . The residual stress distribution of the WJP and unpeened specimens lie symmetrically on either side of the neutral axis. It is thought that the compressive stresses near the surface were much higher than the internal stresses because the severe plastic deformation due to machining causes work hardening at the surface layer. WJP is more effective by considering the working processes, since the stiffness for shock wave is enhanced due to the increased yield stress.

Change of Residual Stress Under Elevated Temperatures. Figure 12 shows the change of residual stress under elevated temperatures. Thermal aging at 623 K for 24 h (623 K–24 h) was conducted to verify the change in residual stress just after the startup of the actual equipment. It was found that the compressive residual stresses were slightly relaxed in the surface layer. Additionally, an accelerated thermal aging test at 773 K for 800 h (773 K–800 h) was conducted to evaluate the long-term stability of the compressive residual stresses induced by WJP under

Fig. 12 Thermal stability of compressive residual stress with surface depth

Fig. 14 Residual stress changes caused by thermal aging in liquid tin bath

Fig. 13 Relationship between stress relaxation and microstructure evolution during thermal aging

elevated temperatures. There were no differences between the 623 K–24 h and 773 K–800 h results. The compressive residual stresses induced by WJP were confirmed to remain stable under elevated temperatures since subsequent changes did not develop. That is, there is very little chance that compressive stresses will relax due to creep because the compressive stresses at the surface layer approaches a finite value of -280 MPa.

In order to clarify the contributing factor for slight stress relaxation at the surface layer, the plastic strain was also visualized. Figure 13 shows the distribution of the kernel average misorientation (KAM) for the various processes, which corresponds to the dislocation density [11]. The plastic strain was increased by machining the surface layer of the unpeened specimen, while thermal aging at 623 K–24 h and 773 K–800 h both relieved the plastic strain. That is, initial stress relaxation occurred since the yield stress of the surface layer was reduced by the removal of the work-hardened material due to machining. This mechanism is further explained below.

Kinetics of Stress Relaxation at Surface. The primary factor of the slight stress relaxation at the surface layer must be investigated. Figure 14 shows the results of the aging tests at lower temperatures and shorter times using the liquid tin bath. The compressive stresses approach a finite value of -280 MPa at each temperature. The changes depend on both temperature and time

Fig. 15 Johnson-Mehl plots for stress relaxation under each aging condition

Fig. 16 Accession of activation energy for stress relaxation

and may be explained by the kinetics. Therefore, a stress relaxation ratio, X , is defined as follows:

$$X = \frac{S_0 - S_V}{S_0 - S_a} \quad (20)$$

where S_0 is the initial residual stress, S_V is the residual stress after aging, and S_a is the saturated residual stress (i.e., -280 MPa). The kinetics proposed by Johnson and Mehl [12] are used

$$X = 1 - \exp\{-(Dt)^n\} \quad (21)$$

Fig. 17 Prediction of stress relaxation during actual plant startup

$$D = D_0 \exp\left(-\frac{Q}{RT}\right) \quad (22)$$

where n is the Avrami exponent, D_0 is the rate constant, R is the gas constant, and Q is the activation energy. These equations can be rewritten as follows:

$$\ln \ln \frac{1}{1-X} = n \ln t + n \ln D \quad (23)$$

$$\ln D = -\frac{Q}{R} \cdot \frac{1}{T} + \ln D_0 \quad (24)$$

From the slope in Fig. 15 and the slope and intercept in Fig. 16, it was determined that $n = 0.586$, $Q = 159.3$ kJ/mol, and

(a)

(b)

Fig. 18 Creep strain versus time in accelerated conditions for Alloy 600. (a) at 773 K and (b) at 723 K.

Fig. 19 Temperature dependence of steady creep rate for Alloy 600

$D_0 = 7.07 \times 10^{11}$ 1/s. The estimated activation energy agrees very well with the self-diffusion energy for Ni [13]. Then, it is shown that initial stress relaxation can occur due to the recovery of severe plastic strain by considering the working processes in conjunction with the change of KAM maps, as shown in Fig. 12. Additionally, the compressive stress relaxation during the actual startup can be theoretically predicted using the additivity rule proposed by Ikawa et al. [14]. In view of the actual operation, the prediction was performed at a positive temperature rate of 0.1 K/min. The results indicate that residual stress relaxation due to recovery will complete during the startup, as shown in Fig. 17. This behavior corresponds to the aging tests at 623 K for 24 h. That is, the long-term stability of WJP under elevated temperatures must be assessed based on compressive stresses with respect to the yield stress

$$X = 1 - \exp\{-\langle \overline{D}t \rangle^n\} \quad (25)$$

$$\overline{D}t = D_0 \int_{t_0}^t \exp\left(\frac{-Q}{RT}\right) dt \quad (26)$$

Long-Term Stability of Stress Improvement by WJP. Since aging cannot be continued for several decades at actual operating temperatures of about 573 K, creep tests were conducted under accelerated conditions, and the results are shown in Fig. 18. At the elapsed time of 100 h, various conditions transferred from transient creep to steady-state creep. Therefore, Eq. (8) can be rewritten as follows:

$$\ln \dot{\epsilon} = n \ln \sigma + \ln A + \left\{ \frac{-Q}{RT} \right\} \quad (27)$$

That is, the creep parameters at actual operating temperatures were estimated using the extrapolation method, as shown in Figs. 19 and 20. In order to predict the long-term stability of

Fig. 20 Stress dependence of steady creep rate for Alloy 600

Fig. 21 Change of residual stress distribution by working processes

Fig. 22 Evaluation of long-term stability over 100 yr at a temperature of 623 K

compressive stresses by WJP, thermal elastic–plastic creep analysis was performed using the given parameters. A value of -280 MPa was chosen as the initial compressive stress based on the above results. Figures 21 and 22 show the prediction for the long-term stability of the compressive stresses. After the equivalent of 100 yr of continuous operation at 80% capacity, there was little change in the WJP compressive stresses under an actual operating temperature of 623 K. Therefore, the long-term stability of WJP during actual operation is predicted.

Conclusions

The long-term stability of compressive residual stresses induced in Alloy 600 by WJP was examined in an elevated-temperature environment. The following conclusions were made:

- (1) The compressive residual stresses near the surface are thought to be much higher than in the interior because the severe plastic deformation due to machining causes work hardening at the surface layer. WJP is more effective by considering the working processes, since the stiffness for shock wave is enhanced due to the increased yield stress, compared to nonstrained materials on the surface.

- (2) The compressive residual stresses slightly relaxed in the surface layer with thermal aging. Additionally, there were no differences between the 623 K–24 h and 773 K–800 h results. The compressive residual stresses induced by WJP were confirmed to remain stable under elevated temperatures. That is, there is very little chance that compressive stresses will relax due to creep.
- (3) Stress relaxation at the surface follows the Johnson–Mehl equation and gives $n=0.586$, $Q=159.3$ kJ/mol, and $D_0=7.07 \times 10^{11}$ 1/s. It was shown that stress relaxation at the surface can occur due to the recovery of severe plastic strain since the estimated activation energy agrees very well with the self-diffusion energy for Ni.
- (4) Stress relaxation at the surface during actual startup was theoretically predicted using the additivity rule. The results indicate that stress relaxation due to recovery will be completed during startup. Therefore, it is proposed that the long-term stability of WJP under elevated temperatures must be assessed based on compressive stresses with respect to the yield stress, to evaluate on the side of prudence.
- (5) Since aging cannot be continued for several decades at actual operating temperatures, thermal elastic–plastic creep analysis was performed to predict the long-term stability. From these results, after equivalent of 100 yr of continuous operation at 80% capacity, there was little change in the WJP compressive stresses under an actual operating temperature of 623 K. Therefore, the long-term stability of WJP during actual operation was confirmed.

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