THERMAL STRESS RELIEVING OF DILUTE URANIUM ALLOYS*

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Abstract

The kinetics of thermal stress relieving of uranium - 2.3 wt. % niobium, uranium - 2.0 wt. % molybdenum, and uranium - 0.75 wt.% titanium are reported and discussed. Two temperature regimes of stress relieving are observed. In the low temperature regime ($T < 300°C$) the process appears to be controlled by an athermal microplasticity mechanism which can be completely suppressed by prior age hardening. In the high temperature regime ($300°C < T < 550°C$) the process appears to be controlled by a classical diffusional creep mechanism which is strongly dependent on temperature and time. Stress relieving is accelerated in cases where it occurs simultaneously with age hardening.

The potential danger of residual stress induced stress corrosion cracking of uranium alloys is discussed. It is shown that the residual stress relief which accompanies age hardening of uranium - 0.75% titanium more than compensates for the reduction in K_{ISCC} caused by aging. As a result, age hardening actually decreases the susceptibility of this alloy to residual stress induced stress corrosion cracking.

Introduction

Dilute uranium alloys are used in a variety of applications because of their high densities (> 18 g/cm^3) and/or their special nuclear properties. Elements such as molybdenum, niobium, titanium, and zirconium are extensively soluble in the high temperature gamma-phase of uranium, but are much less soluble in the intermediate temperature beta-phase, and are virtually insoluble in the low temperature alpha-phase. Uranium alloys are annealed by slow cooling from the gamma-phase field. This allows the gamma solid solution to decompose via a series of diffusional phase transformations, and results in a room temperature structure consisting of essentially pure alpha-uranium plus an alloy rich second phase (1). The annealed condition is usually typified by fairly low yield strength (400-500 MPa) and very limited corrosion resistance. Improved properties are frequently obtained by quenching from the gamma-phase field. Quenching suppresses diffusional decomposition and forces the gamma-phase to transform martensitically to one of several supersaturated variants

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of the alpha-phase (1). Corrosion resistance is greatly improved by virtue of the alloying elements being retained in supersaturated solution. Yield strength can also be increased to $\sim$ 1200 MPa by subsequent age hardening of these supersaturated martensites. An alternative method of increasing yield strength is by deformation hardening. Preliminary investigations suggest that this method can produce yield strengths in the vicinity of 900 MPa when applied to previously annealed material or 1200 MPa when applied to previously quenched material (2).

Unfortunately, potentially dangerous residual stresses are produced by either quenching or deformation hardening. These residual stresses have been shown to be large enough to cause substantial distortion during subsequent machining and, in some cases, to cause fracture during quenching (3,4). Perhaps the most insidious danger associated with residual stresses, however, is that they can result in stress corrosion cracking days, months, or even years after a completed part has been placed into service. Stress corrosion cracking is a time dependent type of fracture that results from a combination of stress and an aggressive environment. Uranium alloys have been shown to be particularly susceptible to this type of failure, stress corrosion having occurred in such apparently benign environments as oxygen and moist air (4-7). In view of the sensitivity of these materials to stress corrosion, it would seem highly desirable to minimize or eliminate all long term stresses from uranium alloy parts.

Residual stresses are most commonly removed from metals by relatively low temperature stress relieving heat treatments performed prior to final machining. Before stress relieving, a previously quenched or deformation hardened part contains a combination of tensile and compressive residual stresses which are balanced against one another such that the part is in static equilibrium. Thermal stress relieving consists of heating the part to a high enough temperature to permit a small amount of time dependent plastic deformation (creep) to occur. Portions that contain tensile residual stresses elongate slightly, while portions that contain compressive residual stresses compress slightly. As a result of these small amounts of deformation the residual stresses are dissipated. Since the amounts of deformation are very small* the part undergoes little macroscopic dimensional change.

Very little is known about thermal stress relieving of dilute uranium alloys. It has been shown that recrystallization of previously cold worked pure uranium begins at temperatures between 400°C and 525°C depending on the extent of prior deformation (8), and that recrystallization is preceded by microstructural changes similar to those which accompany stress relief in other metals (9-11). No previous studies have been performed, however to systemmatically determine the effects of stress magnitude, stress relieving temperature, or stress relieving time on residual stresses in uranium or its dilute alloys.

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*The amount of strain, ε, required to completely remove a residual stress, σ, is given by $\varepsilon = \sigma/E$, where E is the elastic modulus. For dilute uranium alloys $E \sim 1.5 \times 10^5$ MPa, hence a relatively large residual stress of 500 MPa would be removed by a strain of only 0.0033.

The goal of this program was to obtain this information and to provide a basis for understanding thermal stress relieving of dilute uranium alloys.

The alloys chosen for this investigation were uranium - 0.75 wt. % titanium, uranium - 2.0 wt. % molybdenum, and uranium - 2.3 wt. % niobium. These alloys were selected because of their potential for use in a variety of nuclear and non-nuclear weapon components.

Experimental

Stress Relaxation Tests

Stress relaxation tests were performed using the four-point-bending apparatus shown in Figure 1. A thin specimen was placed in the fixture and deflected to produce bending stresses of predetermined magnitude. The bending fixture and bent specimen were then given a vacuum heat treatment during which some of the deflection induced stresses were relieved. After this stress relieving treatment the specimen was unloaded and the amount of remaining deflection was measured. The percentage of stress relief was then taken to be:

$$\%R = \frac{\delta_r}{\delta_\ell} \times 100 \qquad (1)$$

where: δ_r = residual deflection after unloading
δ_ℓ = deflection as loaded.

The precision of the measurements made by this method was estimated to be $\pm$ 1%.

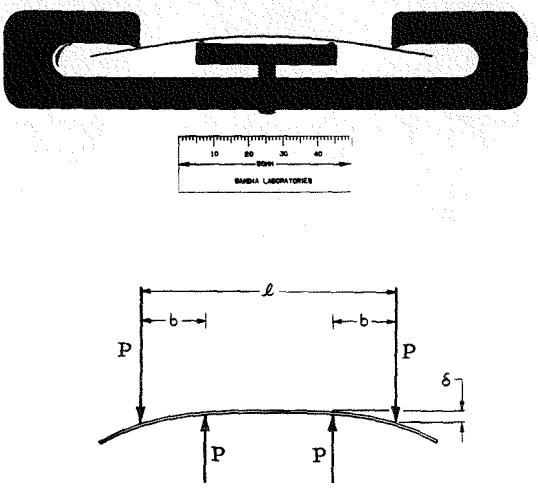

Figure 1. The four-point-bending fixture used for stress relieving experiments.

The relationship between outer fiber stress and deflection was calculated for the geometry shown in Figure 1 using the equations:

$$\delta = \frac{Pb\left(3\ell^2-4b^2\right)}{4\ Ewt^3}, \text{ and} \qquad (2)$$

$$\sigma = \frac{3Pb}{wt^2} \qquad (3)$$

where: E = elastic modulus,
w = specimen width,
t = specimen thickness, and
P, b, ℓ are defined in Figure 1.

Combining equations 2 and 3 gives:

$$\sigma = \frac{12tE\delta}{\left(3\ell^2-4b^2\right)} \qquad (4)$$

In this study $\ell = 7.62 \times 10^{-2}$ m, $b = 1.905 \times 10^{-2}$ m, and E was taken to be 1.50×10^5 MPa, giving $\sigma = 1.13 \times 10^3\ t\delta$, where σ is expressed in MPa and t and δ are expressed in meters.

Specimen Preparation

Specimens 6.35 mm wide x 101.6 mm long x either 0.64 mm or 1.27 mm thick were machined from induction melted stock that had been cross rolled to 14 mm thick plate as described in a previous report (12). The compositions of the alloys used are given in Table 1. The specimens were clamped in a copper and steel fixture to keep them straight, heat treated in vacuum for one hour at 800°C, and either water quenched or furnace cooled to room temperature. Some of the quenched specimens were subsequently age hardened prior to testing. The effects of temperature on stress relief were determined from 25°C to 550°C at a constant time of six hours. The effects of time on stress relief were determined for U-2.3% Nb from 10^2 to 10^6 seconds at 25°C and from 7.2×10^3 to 1×10^6 seconds at elevated temperatures.

Results and Discussion

Annealed Uranium - 2.3 % Niobium

Furnace cooling of uranium - 2.3 wt. % niobium from 800°C results in a two phase microstructure consisting primarily of nearly pure alpha-uranium

Table 1
Alloy Compositions

	U-2.3% Nb	U-2.0% Mo	U-0.75%Ti
Nb (wt.%)	2.22	0	0
Mo (wt.%)	0	2.04	0
Ti (wt.%)	0	0	0.76
Fe (wppm)	65	18	26
Al (wppm)	17	14	9
Cu (wppm)	10	6	4
Si (wppm)	24	28	17
C (wppm)	27	17	15
O (wppm)	28	4	64
N (wppm)	46	40	60
H (wppm)	4	2	3

plus a smaller amount of a niobium-rich gamma-phase. These phases are stable from room temperature to 647°C, so no phase changes should occur in the 25°C to 550°C range of stress relieving temperatures investigated here.

The effects of temperature and stress magnitude on stress relief in uranium - 2.3% niobium are shown in Figure 2. Two well defined temperature regimes exist. Below about 300°C a small amount of stress relief occurs whose magnitude is only slightly dependent on temperature. Above 300°C the amount of stress relief increases markedly with increasing temperature, reaching 100% at approximately 550°C. The effects of time on the low and high temperature processes are shown in Figure 3. The low temperature process exhibits very weak dependence on time, whereas the high temperature process is much more strongly dependent on time.

The fact that the low temperature stress relieving process is very weakly temperature and time dependent suggests that it is not controlled by a classical diffusional creep mechanism. Indeed, a crude estimate of the activation energy for the low temperature process obtained from the data in Figures 2 and 3 indicates an activation energy on the order of 6 kcal/mole°K*, many times

*Figure 2 indicates that a 3% increase in stress relief can be obtained by an increase in temperature from 25°C to 180°C at a constant time of six hours. Figure 3 indicates that a similar 3% increase can be obtained by an increase in time from 1×10^4 seconds to 4×10^5 seconds at a constant temperature of 25°C. Using the relation $Q = R\,[\Delta \ell nt/\Delta(1/T)]$, one obtains $Q = 1.99\,[(12.9-9.2)/(.0034-.0022)] = 6100$ cal $\approx$ 6 kcal/mole°K. It is impossible to associate any statistical confidence interval with this value since it was obtained from very limited data. This crude approximation, then, serves only to demonstrate that the apparent activation energy for low temperature stress

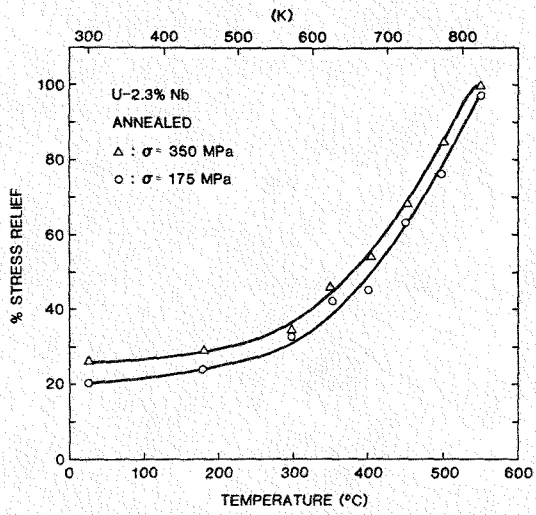

Figure 2. The effects of temperature and stress magnitude on annealed uranium - 2.3% niobium stress relieved for six hours (yield stress = 550 MPa).

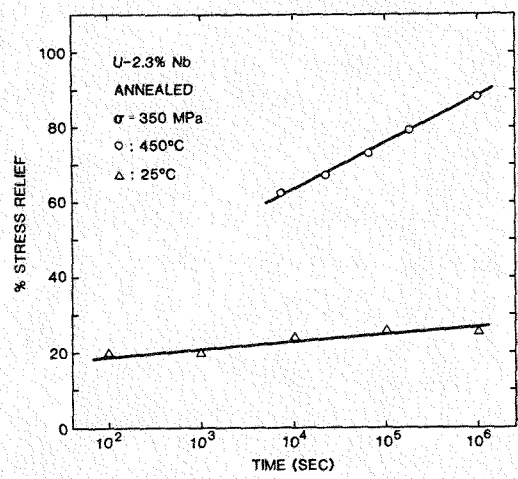

Figure 3. The effects of time on the high and low temperature stress relieving processes in annealed uranium - 2.3 niobium.

lower than expected for a diffusion controlled mechanism in alpha uranium. For comparison the activation energy reported for creep of alpha uranium at temperatures in excess of 400°C is 60 ± 9 kcal/mole°K (13), and that for self diffusion in alpha uranium at temperatures between 580°C and 650°C is 41 ± 15 kcal/mole°K**. It appears likely that low temperature stress relief occurs due to athermal plasticity i.e., to very small amounts of plastic deformation which occur due to dislocation glide (rather than dislocation climb) at stresses well below the macroscopic yield stress. This hypothesis is consistent with the fact that many metals, including uranium, exhibit drastically reduced activation energies for plastic deformation at temperatures below 25% of their melting point (13,16,17). The apparent activation energy reported for athermal plasticity in alpha uranium at 25°C, in fact, is 9 ± 4 kcal/mole°K (13), consistent with the 6 kcal/mole°K estimate obtained from the stress relieving data. Additionally, this hypothesis is supported by the fact that the "elastic" portion of the stress-strain curve of pure uranium departs from linearity at stresses as low as 4 MPa (approximately 2% of the macroscopic yield stress)(18).

relief is very small compared to those reported for high temperature creep and self diffusion in alpha uranium. Similar caution should be exercised in interpreting the other crude estimates of activation energy determined in this report.

**This confidence interval was calculated by the author from the data of Adda and Kirianenko (14) using the method described by Natrella (15).

The high temperature stress relieving process, on the other hand, is markedly dependent on temperature and time, as is typical of creep controlled stress relief in other metals. A crude estimate based on the data in Figures 2 and 3 indicates that the activation energy for this process is on the order of 60 kcal/mole°K. This is in good agreement with the 60 + 9 kcal activation energy reported for diffusional creep in alpha-uranium (13). This agreement is consistent with the hypothesis that the high temperature stress relieving process is controlled by classical diffusional creep of the primary alpha-phase, but additional data would be required to confirm this hypothesis unequivocally.

Quenched Uranium - 2.3 % Niobium

Water quenching of uranium - 2.3 % niobium from 800°C results in a supersaturated martensitic variant of alpha-uranium termed alpha-prime. This phase is metastable at room temperature, but undergoes changes at elevated temperatures which result in age hardening and overaging, as shown in Figure 4. Age hardening occurs between 200°C and 360°C due to very fine scale microstructural changes in the alpha-prime, perhaps the formation of coherent precipitates (4). Overaging occurs above 360°C due to cellular decompositon of the alpha-prime to the equilibrium phases found in the annealed material (4).

The effects of temperature, stress magnitude, and time on stress relief in quenched uranium - 2.3 % niobium are shown in Figures 5 and 6. Two temperature regimes are again observed.

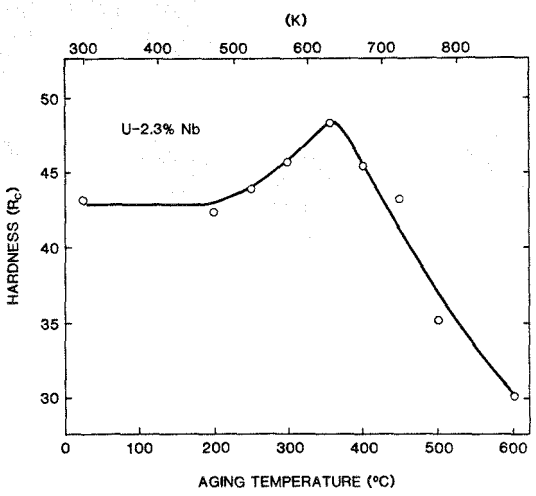

Figure 4. The age hardening behavior of previously quenched uranium - 2.3% niobium.

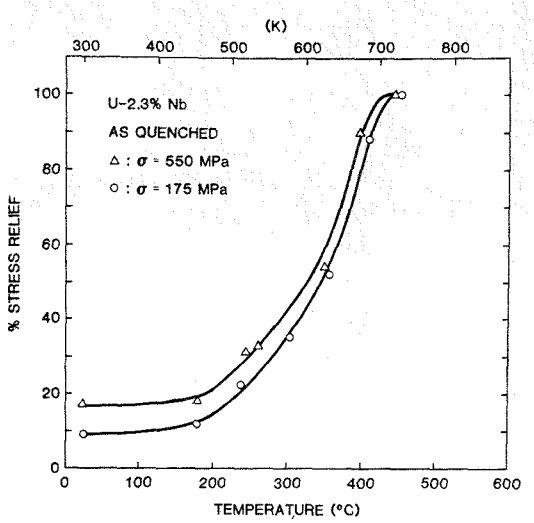

Figure 5. The effects of temperature and stress magnitude on quenched uranium - 2.3% niobium stress relieved for six hours (yield stress = 730 MPa).

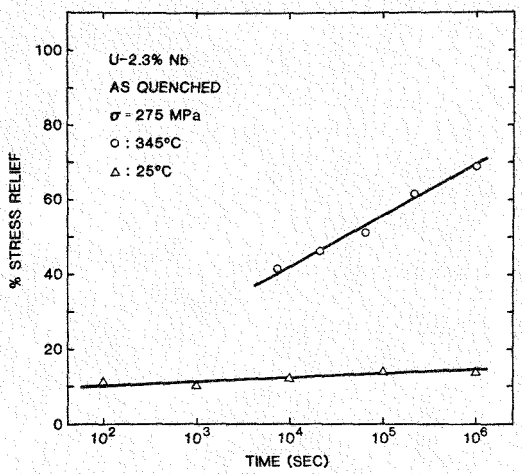

Figure 6. The effects of time on the high and low temperature stress relieving processes in quenched uranium - 2.3% niobium.

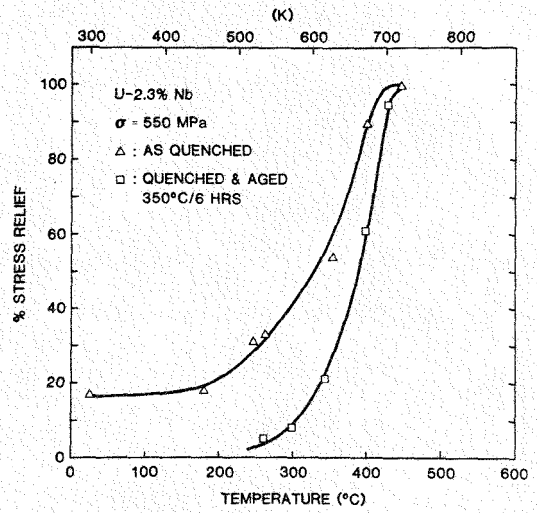

Figure 7. Comparison of the behaviors of quenched and quenched plus aged uranium - 2.3% niobium stress relieved for six hours (yield stresses = 730 MPa, 1200 MPa, respectively.

As was the case with the annealed material, the low temperature process exhibits very weak temperature and time dependence. The apparent activation energy for this process is on the order of 10 kcal/mole°K. These weak temperature and time dependencies plus the very low activation energy again suggest that an athermal microplasticity mechanism is responsible for low temperature stress relief. The primary difference between the low temperature behavior of the annealed and quenched materials is that a smaller amount of stress relief occurs in the quenched material at a fixed percentage of the macroscopic yield stress. This implies that the niobium atoms in solid solution in the quenched material restrict microplastic deformation.

The high temperature process again appears to be controlled by a classical diffusional creep mechanism. Stress relief in this regime is strongly dependent on temperature and time, and the activation energy is on the order of 50 kcal/mole°K. The high temperature process begins to dominate at 200°C and results in 100% stress relief at 450°C. Comparison of Figures 2 and 5 reveals that this temperature range is approximately 100°C lower that that over which the high temperature process operates in annealed material. It is believed that the acceleration of stress relieving in the quenched material is a result of phase transformations which are occurring in the metastable alpha-prime phase during stress relieving at temperatures in excess of 200°C. Previous investigators have shown that precipitation of very fine carbides during tempering accelerates stress relief in martensitic steels and have interpreted their results in terms of plastic deformation fields associated with the growing pre-precipitates (19). Comparison of Figures 4 and 5 indicates that both rapid stress relief and age hardening begin at 200°C, suggesting that transformation enhancement is also responsible for the accelerated stress relieving observed in quenched uranium-2.3% niobium.

To test this hypothesis, a series of quenched specimens were aged at 350°C for six hours prior to insertion in the four-point-bending fixture. This pre-aging produces a material in which no additional transformation will occur at temperatures significantly below that at which pre-aging had been done. The stress relieving behaviors of the pre-aged and as-quenched materials are compared in Figure 7. The fact that the onset of rapid stress relieving is pushed back to approximately 300°C by the pre-aging treatment confirms that the accelerated stress relief occurring in the as-quenched materials is caused by the instability of the alpha-prime phase. In addition, Figure 7 shows that pre-aging essentially eliminates low temperature stress relief. This adds additional support to the hypothesis that low temperature stress relief occurs due to microplasticity, as short range dislocation glide would be expected to be significantly restricted by the types of fine scale microstructural changes that result in age hardening. Similarly, it has been shown that pre-aging markedly reduces the extent of low temperature stress relief in copper-beryllium alloys (20), suggesting that this is a general effect in age hardenable alloys.

Uranium - 2 % Molybdenum

Uranium - 2 % molybdenum is similar to uranium - 2.3 % niobium in both physical metallurgy and stress relieving behavior. Furnace cooling from 800°C results in a two phase microstucture of virtually pure alpha uranium plus the intermetallic compound U_2Mo. This microstructure is stable over the entire range of stress relieving temperatures investigated. Water quenching from 800°C results in a supersaturated alpha-prime martensite which is metastable at room temperature. This martensite age hardens at temperatures between 200°C and 400°C and overages due to cellular decomposition at temperatures above 400°C (21).

The effects of temperature on stress relieving of annealed and as-quenched uranium - 2 % molybdenum are shown in Figure 8. Low temperature and high temperature regions are again seen, and stress relieving of the as-quenched material occurs at about 100°C lower temperatures than the annealed material. Figure 9 compares the behaviors of as-quenched and quenched plus pre-aged material, confirming again that instability of the alpha-prime phase is responsible for the accelerated high temperature stress relieving in the as-quenched material and that pre-aging eliminates low temperature stress relieving.

Uranium - 0.75 % Titanium

The physical metallurgy of uranium - 0.75 % titanium is similar to those of the alloys discussed previously. Furnace cooling from 800°C results in a two phase microstructure of alpha uranium plus U_2Ti, but the material is never used in this condition because of poor mechanical properties, hence the stress relieving behavior of annealed material was not measured. Water quenching from 800°C results in a supersaturated alpha-prime martensite which can be age hardened between 300°C and 460°C and which overages due to cellular decomposition above 460°C (21).

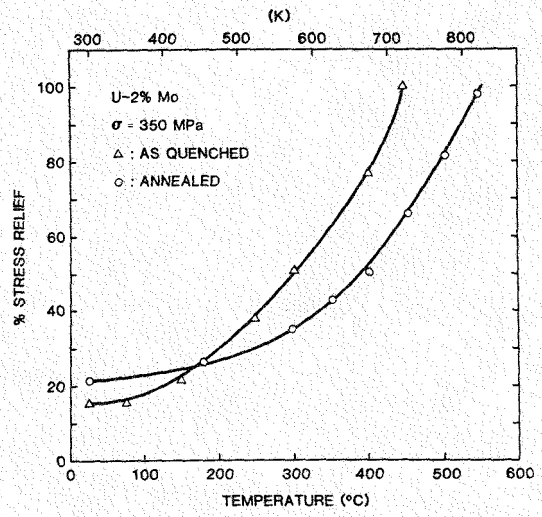

Figure 8. The effect of temperature on annealed and quenched uranium - 2% molybdenum stress relieved for six hours (yield stresses = 480 MPa, 715 MPa, respectively).

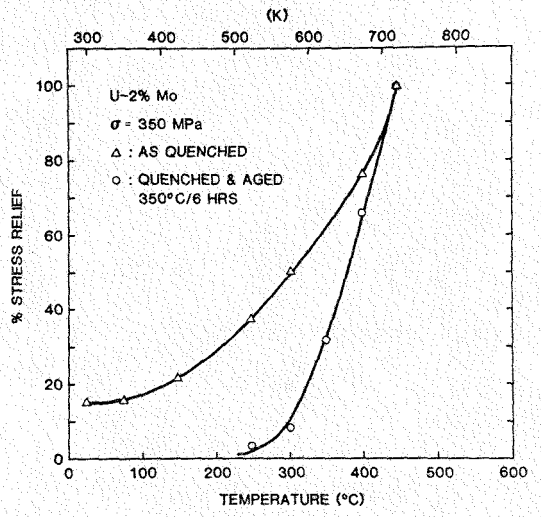

Figure 9. Comparison of the behaviors of quenched and quenched plus aged uranium - 2% molybdenum stress relieved for six hours (yield stresses = 715 MPa, 1200 MPa, respectively).

The stress relieving characteristics of the as-quenched and quenched plus pre-aged material are shown in Figure 10. A low temperature process which is apparently controlled by athermal microplasticity dominates at temperatures below 300°C in the as-quenched material. A high temperature process which is apparently creep controlled dominates above the 300°C temperature at which age hardening begins. Pre-aging again eliminates the low temperature process and

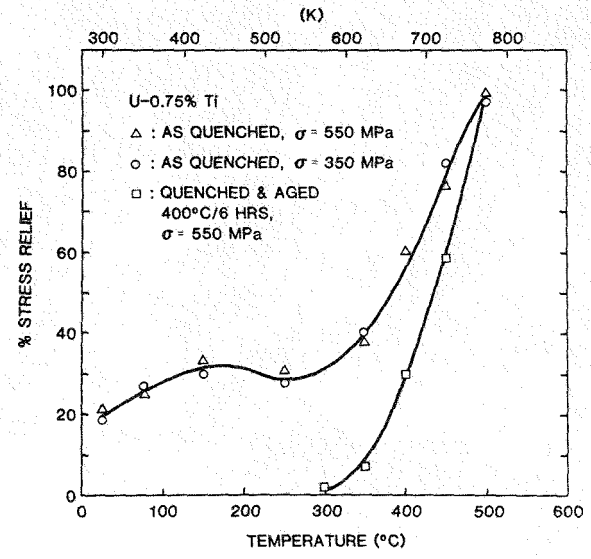

Figure 10. The behavior of quenched and quenched plus aged uranium - 0.75% titanium stress relieved for six hours (yield stresses = 700 MPa, 1150 MPa, respectively).

retards the high temperature process, reconfirming that low temperature stress relief occurs due to microplasticity and that instability of the alpha prime martensite accelerates high temperature creep controlled stress relief.

Effect of Stress Magnitude

It is difficult to draw fundamental conclusions about the effect of stress magnitude on thermal stress relieving from this work because each specimen contained a spectrum of stresses ranging from the outer fiber stress in tension on one surface, to zero stress at the neutral axis, to the outer fiber stress in compression on the opposite surface. This stress distribution, however, is very similar to that which exists in quenched or deformation hardened components, hence these tests closely simulate conditions in actual engineering structures. While it would be invalid to draw precise quantitative conclusions regarding the effect of stress magnitude on stress relieving kinetics, the results clearly indicate that stress magnitude is of secondary importance, especially in the high temperature stress relieving range of primary practical interest. This is fortunate because it will permit stress relieving heat treatments to be designed without first performing the difficult task of determining the distributions and magnitudes of residual stresses in engineering components.

Implications Regarding Residual Stress Induced Stress Corrosion Cracking of Uranium - 0.75 % Titanium

As was discussed previously, perhaps the greatest danger associated with residual stresses in uranium alloys is the possibility of their causing stress corrosion cracking during the service life of a component. The threshold stress intensities for stress corrosion cracking of uranium - 0.75 % titanium in an aqueous solution containing 50 ppm chloride have been shown by Magnani to decrease with increasing age hardening (5). This might seem to imply that age hardening increases the susceptibility of engineering components to residual stress induced stress corrosion cracking. This conclusion is not necessarily valid, however, as the residual stress relieving that accompanies age hardening also reduces the driving force for stress corrosion cracking. The net consequences of these competing effects can be evaluated by combining the data of Magnani and those of Figure 10 in a simple fracture mechanics analysis.

It is well known that a pre-existing flaw in an engineering structure will not propagate as a stress corrosion crack provided the stress intensity, K, associated with the flaw is less than the critical threshold intensity, K_{ISCC}, for the material and environment in question (22). The stress intensity, K, is analytically related to the flaw size, the stress magnitude, and the geometry of the structure. Once the value of K_{ISCC} has been determined on a test specimen of simple geometry, then, calculations can be made on structures of engineering concern to determine what combinations of stress magnitude and flaw size will render the structure susceptible to stress corrosion cracking. For obvious reasons, it is desirable to establish inspection procedures that will insure that the stress magnitudes and flaw sizes are sufficiently small to keep the

stress intensities in the structure below K_{ISCC}. The larger the tolerable flaw, the easier it is to inspect the parts and identify those which should be rejected. Hence, critical flaw size provides a logical figure of merit for evaluating resistance to residual stress induced stress corrosion cracking.

For a large number of common engineering geometries the stress intensity, K, is proportional to the stress, σ, and the square root of the flaw size, a. The constant of proportionality varies with geometry, but for the simple case of a shallow edge crack in a wide plate the relationship is:

$$K \approx 2.0\ \sigma\sqrt{a}\ ^{*} \qquad (5)$$

Equation 5 can be rearranged to give an expression for the critical flaw size as a function of the stress magnitude and the threshold stress intensity for stress corrosion cracking:

$$a_c \approx .25\left(\frac{K_{ISCC}}{\sigma}\right)^2 \qquad (6)$$

The following values for K_{ISCC} were determined by Magnani (5): as-quenched, 27.1 MPa $m^{1/2}$; aged 357°C for 18 hours, 20.9 MPa $m^{1/2}$; aged 400°C for 42 hours, 13.2 MPa $m^{1/2}$. Using the 50 kcal/mole°K activation energy for age hardening of this material (21) it was found that the 357°C/18 hour and 400°C/42 hour aging treatments are equivalent to six hour ages at 375°C and 437°C, respectively. In addition, based on the fact that no changes in tensile properties occur during aging treatments up to 300°C/6 hours, it was assumed that K_{ISCC} = 27.1 MPA $m^{1/2}$ for all aging temperatures below 300°C. This information is given in Table 2. Values of residual stress after stress relieving, σ, were determined for six hour stress relieving treatments at 25°C, 175°C, 250°C, 375°C, and 437°C using the information in Figure 10 and assumed initial residual stress levels of 200, 400, and 600 MPa. This information is also given in Table 2. The values of K_{ISCC} and σ appropriate to each temperature were then substituted into Equation 6 to determine the effect of aging on critical flaw size. The results are shown in Table 2 and Figure 11. It can be seen that critical flaw size generally increases with increasing aging temperature. This is contrary to what one would casually expect based on the fact that K_{ISCC} decreases with increasing aging temperature. The increase in critical flaw size, however, results from the fact that the beneficial reduction in residual stress magnitude accomplished by aging outweighs the detrimental reduction in K_{ISCC}. The slight inflection in critical flaw size between 180°C and 280°C corresponds to the unusual but apparently reproducible inflection in stress relieving over that temperature range. For the part part, however, the susceptibility of uranium 0.75% titanium to residual stress induced stress corcosion cracking is reduced by age hardening.

*The constant of proportionality increases from the limiting value of 2.0 as the crack length becomes appreciable with respect to the width of the plate (22).

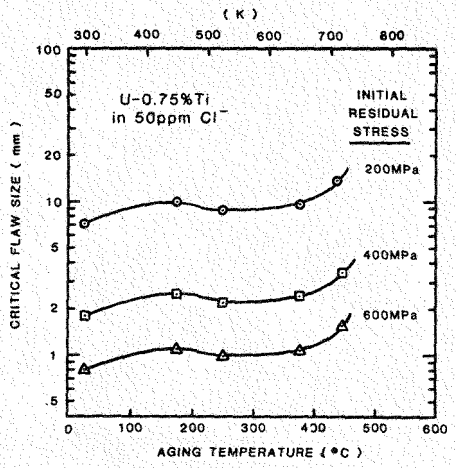

Figure 11. The effects of aging on critical flaw size for a small flaw in the edge of a wide plate of uranium - 0.75% titanium tested in 50 ppm chloride solution.

Table 2

Determination of Critical Flaw Sizes for Stress Corrosion Cracking of U-0.75% Ti in 50 ppm Cl^-

Aging/Stress Relieving Temp* (°C)	K_{ISCC} (Mpa $m^{1/2}$)	Stress Relief (%)	Initial Stress (MPa)	Stress After Stress Relief (MPa)	Critical Flaw Size (mm)
25	27.1	20	200	160	7.17
			400	320	1.79
			600	480	0.797
175	27.1	32	200	136	9.93
			400	272	2.48
			600	408	1.10
250	27.1	28	200	144	8.85
			400	288	2.21
			600	432	0.984
375	20.9	47	200	106	9.72
			400	212	2.43
			600	318	1.08
437	13.2	72	200	56	13.9
			400	112	3.47
			600	168	1.54

*Six hours at temperature

Conclusions

1. Thermal stress relieving in dilute uranium alloys occurs by two processes, one which dominates at low temperatures ($T < 300°C$), and one which dominates at high temperature ($300°C < T < 550°C$).

2. The low temperature process is very weakly dependent on temperature and time, exhibits very low apparent activation energies, and can be completely suppressed by prior age hardening. All of these features suggest that this process is controlled by an athermal microplasticity mechanism. This is also consistent with previous observations of extensive microplasticity below the macroscopic yield stress of uranium and dilute uranium alloys.

3. The high temperature process is much more strongly dependent on temperature and time, and exhibits activation energies which are consistent with those for diffusion controlled creep of alpha uranium. These features suggest that this process is controlled by classical diffusional creep.

4. The high temperature process is accelerated when stress relieving and age hardening occur simultaneously.

5. Stress magnitude has a relatively minor influence on stress relieving kinetics.

6. Calculations indicate that age hardening of uranium - 0.75% titanium reduces susceptibility to residual stress induced stress corrosion cracking in 50 ppm chloride solution because the beneficial effect of stress relieving outweighs the detrimental effect of reduced values of K_{ISCC}.

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