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RESIDUAL STRESSES IN URANIUM AND URANIUM ALLOYS*

K. H. Eckelmeyer

Sandia National Laboratories, Albuquerque, NM 87185, USA

ABSTRACT

Residual stresses are generated in uranium alloys primarily by metal working operations and heat treatments. These stresses can be large enough to seriously degrade material manufacturability and integrity. they have been observed to cause centerline bursting in bars, stress corrosion cracking in components of various geometries subjected to a variety of environments, and distortion during machining. The magnitudes and consequences of these stresses can be minimized by mechanical or thermal stress relieving. Residual stresses in uranium alloy components have been measured mostly by destructive relaxation methods, but recently a non-destructive neutron diffraction method has been developed.

INTRODUCTION

Uranium is used in a variety of applications because of its very high density and/or its unique nuclear properties. Uranium and its alloys exhibit typical metallic ductility, can be fabricated by most standard hot and cold working techniques, and can be heat treated to hardnesses ranging from approximately R_B 92 to R_C 55.

Solid elemental uranium exhibits three polymorphic forms: γ (gamma) phase (BCC) above 775°C, β (beta) phase (tetragonal) between 775°C and 667°C, and α (alpha) phase (orthorhombic) below 667°C. Hot working (rolling, forging, extruding) is readily accomplished in the γ (800-840°C) or high α (600-640°C) temperature ranges, and cold or warm working (rolling, swaging, drawing) can be done from room temperature to about 400°C. Recrystallization of cold worked material can be performed in the high α region (500-640°C). The material can be machined by most normal cutting and grinding techniques, but special tools, cutting conditions, and safety precautions are recommended.

Uranium is frequently alloyed to improve its corrosion resistance and/or modify its mechanical properties. These alloys can be fabricated hot, warm, or cold similar to unalloyed uranium. As shown in Fig. 1, the high temperature γ phase

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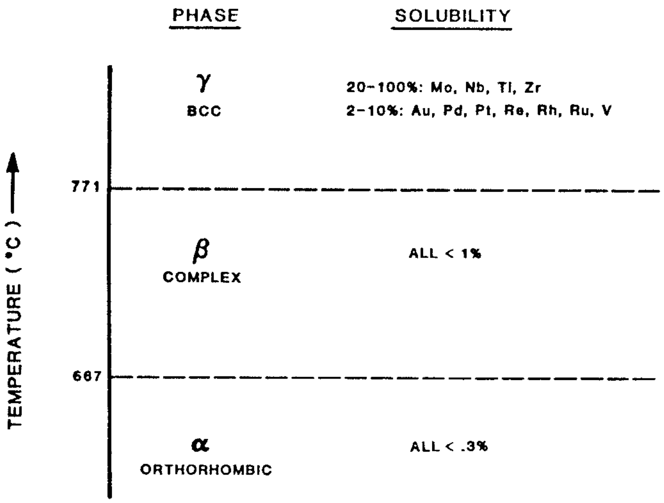

Fig. 1 Polymorphism and solubilities of alloying elements in uranium.

can dissolve substantial amounts of a number of alloying elements, but these elements are much less soluble in the intermediate and low temperature β and α phases. Uranium alloys are generally heat treated at about 800°C to get all of the alloying additions into solid solution in the γ phase, and then cooled at various rates to room temperature. Slow cooling permits the γ phase to decompose diffusionally to two phase microstructures consisting of nearly pure α-uranium and alloy rich second phases (frequently intermetallic compounds). These annealed microstructures are analogous to pearlite in steels. Rapid quenching suppresses these diffusional decomposition modes and results in microstructures containing a variety of metastable variants of α or γ-uranium. These quenched microstructures are analogous to martensite and retained austenite in steels. The metastable phases produced by quenching are supersaturated, hence they are amenable to subsequent lower temperature heat treatment. Since they are substitutional solid solutions they are relatively soft (R_B 92 to R_C 35). As a result, subsequent lower temperature heat treatment results in increased hardness and strength. Age hardening occurs at temperatures below 350°C to 450°C due to fine scale microstructural changes that can only be observed by transmission electron microscopy and/or other very high resolution techniques. Overaging occurs at higher temperatures by decomposition of the metastable structures. This decomposition commonly occurs by cellular or discontinuous precipitation, and can be readily observed by optical metallography. Additional information on uranium alloy metallurgy is available in several reviews [1-6].

Residual stresses are frequently introduced into uranium and uranium alloys by metal fabrication processes (rolling, welding, etc.) and heat treatment (especially quenching). Relatively few studies have been conducted in which these stresses have been explicitly measured by standard techniques for residual stress determination. Instead, their presence has been largely inferred from their consequences, which include centerline bursting, delayed cracking, and distortion during machining. The generation, relief, and consequences of residual stresses are discussed in greater detail in the following sections.

GENERATION OF RESIDUAL STRESSES

Residual stresses can be generated in uranium and uranium alloys primarily by metal fabrication processes and heat treatment, and to a lesser extent by machining processes and environmental interactions such as oxidation.

Metal Fabrication Processes

Residual stresses are generated when metal fabrication processes are carried out below the temperature at which spontaneous stress relieving occurs. Primary fabrication of uranium and uranium alloys is done by rolling, extrusion, or forging in either the γ (800-950°C) or high-α (600-640°C) temperature ranges [7]. Recrystallization occurs spontaneously at these temperatures [1], hence primary metal working processes should not be a major source of residual stress. Indeed, U-0.75% Ti* bars rolled at 600°C [8] and U-2% Mo bars forged at 500°C [9] were shown to contain little or no residual stress. These processes, however, can produce preferred crystallographic orientations (textures) [1], and since the thermal expansion of uranium is strongly anisotropic, substantial residual stresses can be generated during air cooling from the working temperature due to anisotropic thermal contraction. Rapid cooling from the working temperature can also produce substantial residual stresses, as will be discussed in more detail in the section on heat treatment. Hot formed uranium shapes are sometimes sprayed with water in order to minimize the extent of oxidation which occurs rapidly at high temperature. This is believed to have been responsible for the large residual stresses observed in U-2% Mo bars forged at 620°C by one manufacturer [9].

Secondary fabrication processes, such as rolling, swaging, deep drawing, shear forming, and straightening are sometimes carried out at temperatures as low as 25°C. When these processes are carried out below the temperature at which spontaneous recovery and/or recrystallization occur they introduce residual stresses or substantially alter existing residual stress patterns. The effect of cold straightening on residual stress distribution in U-0.75% Ti, for example, has been computer modelled [10]. The results indicate that portions of the surface which had exhibited ∿425 MPa compressive residual stress in the as-quenched condition will exhibit ∿ 340 MPa tensile residual stress following straightening. Substantial tensile surface stresses have also been measured in 50% cold rolled bars of U-0.75% Ti [11]. Dimensional instability and delayed cracking have also been observed in material fabricated at 25°C [7], thus confirming that low temperature fabrication results in substantial residual stresses. No detailed studies have been performed, however, in which the magnitudes or distributions of these stresses have been systematically measured and related to processing conditions. Studies of recovery and recrystallization in uranium suggest that fabrication induced residual stresses are likely to occur at deformation temperatures below 400 to 500°C [1].

Welding results in substantial residual stresses which sometimes result in delayed cracking in various uranium alloys [12]. Detailed measurements of welding induced residual stresses, however, have not been made.

Heat Treatment

The quenching step associated with solution heat treatment of many uranium alloys frequently results in the generation of very high residual stresses. Quenching of solid cylinders results in radial heat extraction. As the outside of the cylinder cools it shrinks and becomes rigid, while the interior, which

*All compositions are given in weight per cent.

remains hot and soft, plastically deforms to accommodate the shrinking exterior. When the interior subsequently cools it too attempts to shrink, but is constrained by the rigid outer shell. As the interior cools, then, a state of triaxial tension develops, reaching a maximum along the centerline of the cylinder, as shown in Fig. 2 [10,13]. Volume changes associated with the phase transformations which occur on cooling could substantially contribute to residual stress development, but the magnitude of this effect has not yuet been studied in detail. Quenching of plate results in a similar sequence, but since heat extraction is planar, the resulting stress state is biaxial. Quenching of hollow cylinders is similar to plate, but since heat is extracted more effectively from the outer surface than the inner surface the resulting stress distribution is asymmetric through the cylinder wall, as shown in Fig. 3. In nearly all cases the stresses are primarily compressive on the surface and tensile in the interior. Machining of quenched parts into engineering components, however, can remove compressively stressed material, thus causing residual stress redistribution and sometimes resulting in products with tensile surface stresses in some areas [10,11].

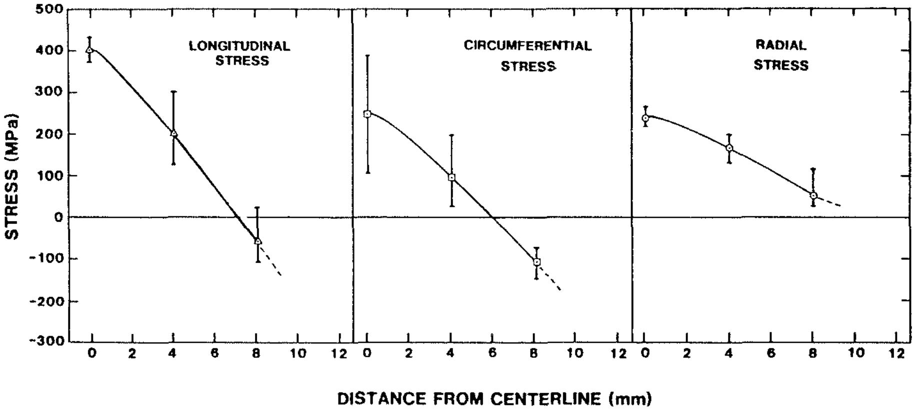

Fig. 2 Residual stresses determined by neutron diffraction in 25.4 mm diameter cylinder of U-0.75% Ti which had been quenched from 850°C, rotary straightened and aged at 380°C for six hours. The data bars denote the actual range of measurements made at four 90° intervals around the bar [8].

Machining and Grinding

Machining and grinding can produce residual stresses, particularly when excessive cutting speeds and feeds are used [14]. In most cases, however, these effects are smaller and/or more closely confined to the surface region than those produced by fabrication or heat treatment. Machining induced cracking has been observed [7], but is not a frequent problem. X-ray diffraction, however, reveals evidence of substantial surface deformation associated even with fine grinding, and in some cases, deformation induced phase changes [15]. In soft materials such as unalloyed uranium and U-6% Nb, plastic deformation has been observed to penetrate quite deeply into the material [14,16], thus producing and/or altering residual stresses, causing distortion, and making machining to close tolerances difficult.

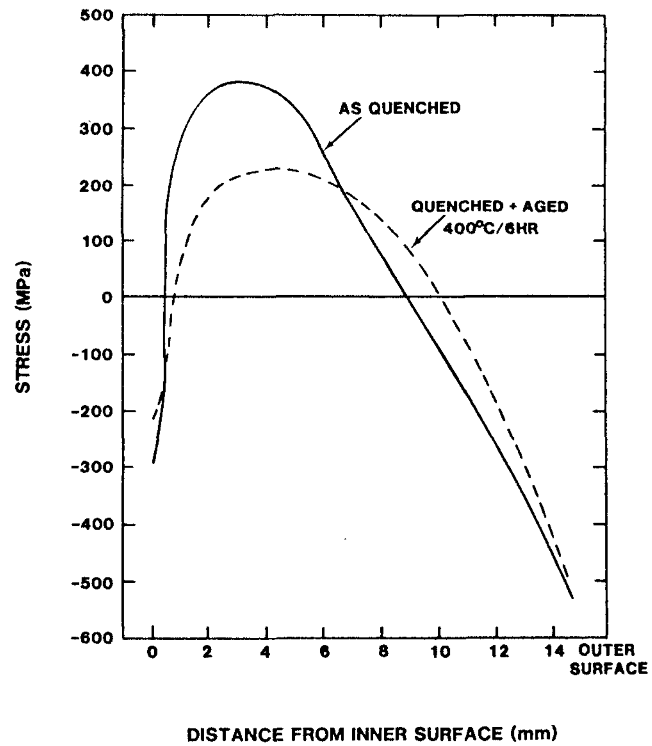

Fig. 3 Residual stresses determined by the Sachs boring method in 206 mm od x 168 mm id cylinder of U-0.75% Ti which had been water quenched from 800°C. Approximately 1.6 mm was removed from the inside and outside surfaces after heat treatment to provide a scale free surface to which strain gauges could be bonded. Longitudinal and circumferential stresses were found to be identical to within the accuracy of the measuring technique [13].

Environmental Interactions

The volume changes associated with the formation of thin layers of oxide on uranium have been shown to produce stresses in the underlying metal [17], but this is not generally of much consequence except to the dimensional stability of very thin parts. The volume changes associated with the formation of corrosion products in cracks, however, can cause crack wedging and production of tensile stresses ahead of the crack tip, thus providing a situation in which stress corrosion cracking can occur in a self-propagating manner [18], potentially even through regions where the macroscopic residual stresses are compressive.

MEASUREMENT AND PREDICTION OF RESIDUAL STRESSES

Relaxation Methods

Until very recently relaxation methods were the only successfully used techniques for measuring residual stresses in uranium alloys. The Sachs boring method

[19,20] was applied to uranium as early as 1961 to measure machining stresses [14] and stresses introduced by fuel element cladding processes [21]. More recently this methoud has been used to measure the stresses generated by quenching in solid and hollow cylinders of U-0.75% Ti [10,13]. A small amount of material was machined from the cylinders after quenching to provide a scale free surface to which four pairs of strain gauges (one longitudinal and one circumferential) could be bonded at 90° intervals. In some cases the parts were aluminum ion plated to facilitate strain gauge bonding. After strain gauging the cylinders were incrementally bored by electrical discharge machining. The longitudinal and circumferential stress distributions through the thickness of ther part were calculated from the strain gauge responses associated with each boring increment. The hollow cylinder results, which have already been discussed, are shown in Fig. 3.

While this is a well established technique and appears to be applicable to uranium alloys, it suffers from several drawbacks. First, it is destructive, time consuming, and costly. Second, it assumes that the stress distribution is symmetric about the cylinder axis and gives only average stresses as a function of radial distance from this axis, rather than stresses at specific spacial locations. Finally, stresses are calculated based on average values of elastic properties assuming linear elasticity and isotropic behavior, whereas uranium and its alloys are known to be substantially anisotropic, especially in wrought products containing substantial preferred crystallographic orientation (texture), and to depart from linear elastic behavior at stresses well below those required for macroscopic plastic deformation. Hence, the stress magnitudes determined by this technique should be taken as approximate, rather than precise values. This final shortcoming is also applicable to most other techniques which have been used for residual stress measurement in uranium alloys.

A hole drilling method [22] has also been used to measure residual stresses in components which had been machined from ∿ 26 mm diameter rods [11]. Following heat treating and machining, strain gauge rosettes were bonded to the components at several locations of interest, and the responses of these gauges were monitored as ∿ 1.6 mm diameter holes were drilled into the cylinders using an airjet abrasive technique. Near surface stress distributions were calculated from these responses. It was found that U-2% Mo components which had been aged at 500°C contained small compressive stresses at the surface. While this technique permits determination of stresses at specific locations, it is also destructive and time consuming, and it can only accurately determine the distribution of stresses near the surface, rather than through the entire thickness of the part.

Diffraction Methods

While x-ray diffraction is commonly used to measure surface residual stresses in a variety of metals, x-ray stress measurement in uranium and its alloys is severely complicated by the fact that the back reflection region of x-ray diffraction patterns produced with the standard x-ray tubes contain a large number of overlapping peaks. Some early successes were reported in this area [23,24], but later experimenters were unsuccessful in their efforts to use x-ray diffraction methods for residual stress determination in uranium alloys [10,25]. Consideration is currently being given to developing a technique based on use of long wavelength x-rays such as silver or palladium Lα (λ = 4.15 or 4.37 A, respectively) [26]. This would overcome the problem of overlapping diffraction peaks by moving well defined high d-spacing peaks such as $\{111\}_\alpha$ into the back reflection region. The relatively low energy of this radiation, however, would severely limit its penetration into the sample (more than 95% of the diffracted radiation would originate from within 1 μm of the surface), thus making this technique useful only for surface stress determination, and making the results extremely sensitive to surface preparation and condition.

A neutron diffraction technique has recently been demonstrated for residual stress measurement in uranium alloys [8]. Residual stress measurement by neutron diffraction is conceptually identical to the x-ray diffraction technique in that stress magnitudes are determined from small detected shifts in crystal plane spacings. Neutrons have the great advantage, however, of penetrating orders of magnitude deeper into the sample than x-rays (> 1 cm in uranium). Hence, neutron diffraction offers the possibility of non-destructively measuring residual stresses through the entire thickness of an engineering component. In the referenced study, for example, shifts in the spacings of the $\{111\}_\alpha$ planes were used to measure the residual stresses at the centerline and several mid-radius positions in a 25.4 mm diameter cylinder of U-0.75% Ti. Spacial resolution was obtained by selecting diffraction conditions resulting in a diffracting angle of approximately 90°, and by using relatively narrow slits on the orthogonal primary and diffracted beams to define the ΔX and ΔY dimensions of the areas being characterized. Volumes as small as 1.5 mm x 1.5 mm x 30 mm or 2.5 mm x 2.5 mm x 5 mm were characterized with this technqiue. The results of this study are shown in Fig. 2. The data spread is not indicative of error or uncertainty in the data, but represents the actual range of measurements taken at 90° intervals around the bar. The substantial differences between quadrants apparently result from the rotary straightening operation that was performed on the bar after heat treatment. This is consistent with the results of a computer modeling study which predicts that a small amount of non-uniform plastic deformation, such as occurs during straightening of a slightly bent bar, will substantially alter the residual stress distribution in the bar and result in a non-axially symmetric set of residual stresses [10].

The neutron diffraction method has substantial advantages in that it is a non-destructive technique capable of measuring internal stress distributions with moderate spacial resolution. It is potentially capable of including mechanical property anisotropy in its formulation, although this was not done in the referenced study due to lack of the appropriate data. By far its greatest disadvantage, however, is that it requires a reactor source of neutrons, a resource not generally available at most research or production facilities.

Prediction of Residual Stresses

Only recently have attempts been made to predict residual stresses in uranium alloys. One group has developed a model for predicting how existing residual stresses in quenched bars will be redistributed by cold straightening operations [10]. While this model's predictions have not been experimentally tested, it represents a relatively straightforward approach and seems likely to give reasonably accurate predictions.

The more general problem of predicting residual stresses induced by quenching is much more complex. While prediction of quenching stresses is difficult for any metal, several features make uranium alloys perhaps the most difficult of all metals in which to model this phenomenon. First, uranium alloys undergo a variety of equilibrium and non-equilibrium phase transformations under different cooling conditions. Many of these transformations occur in very short times, hence a modest amount of imprecision is associated with the determination of their kinetics. In addition, a range of heats of transformation are associated with these reactions. Since these heats of transformation must be input into the thermal model, even the simplest task of predicting cooling rates and the resulting microstructures becomes a complex iterative process. Second, dimensional changes are associated with these phase transformations and must be taken into account. Third, the transformation products have a range of static and time dependent mechanical properties which must be measured and input into the residual stress model. Many of these microconstituents are metastable and exist only for very short times in specific temperature ranges, thus making their properties virtually impossible to measure. Finally, the thermal expansion behaviors and mechanical properties of uranium alloys are strongly anisotropic, and most uranium alloy components exhibit preferred crystallographic orientations. Hence quantitative descriptions of crystallographic texture and property anisotropy must also be input into the model, and the common simplifying assumption of material isotropy cannot be made to facilitate solution of the problem. Despite these and other difficulties, a model has very recently been developed for predicting microstructures and residual stresses in quenched U-0.75% Ti cylinders [27]. This model will now need to be experimentally tested and perhaps modified. If successful, however, it will represent an impressive solution to a very complex problem, and will provide a real breakthrough in the understanding of residual stress effects in uranium and its alloys.

RELIEVING OF RESIDUAL STRESSES

Mechanical Methods

Residual stresses in uranium alloy components are sometimes mechanically relieved by stress leveling. Cylinders of U-0.75% Ti containing surface compressive stresses and interior tensile stresses as large as 350 MPa have been stress relieved by approximately 3% compressive deformation (upsetting) at 25°C [13]. Following this stress leveling treatment no residual stresses were detected by the Sachs boring technique. Additional investigation revealed that distortion during subsequent machining was substantially reduced, but not eliminated [28]. This implies either that moderate undetected residual stresses remained after the stress leveling treatment, or that subsequent residual stresses were introduced by the machining operation. While stress leveling is a valuable technique for reducing residual stresses, it can only be applied to parts of simple geometry, such as cylinders.

Thermal Methods

Components of any geometry can be heated to temperatures where small amounts of microplasticity occur, thus relieving residual stresses. Sharpening of the x-ray diffraction peaks in previously cold worked unalloyed uranium indicates that such recovery processes begin in the vicinity of 400°C [1]. A detailed study has been conducted on stress relaxation in annealed unalloyed uranium in the 150°C to 500°C range [29]. The primary results of this study are shown in Fig. 4. Relatively little stress relieving occurs below approximately 350°C, and a temperature of approximately 500°C is required to substantially stress relieve engineering components in a practical time. Stress relaxation was found to occur in two stages: an initial rapid stage having an activation energy substantially lower than that for self diffusion in α-uranium, followed by a steady state stage having an activation energy consistent with that for self diffusion in α-uranium. It was also shown that stress relaxation occurs more rapidly in quenched material and much more rapidly in material that had been cold worked. This was attributed to stress induced migration of excess vacancies in the quenched and cold worked materials.

A study of thermal stress relieving has also been conducted on a variety of uranium alloys [30]. Thin specimens were deflected in a four point bending fixture to produce predetermined outer fiber stresses. The fixture and sample were then vacuum heat treated at various temperatures and times. Following heat treatment the sample was removed from the fixture and the amount of stress relief was determined from the amount of springback (no springback implying 100% stress relief, and complete springback implying 0% stress relief).

The results of experiments conducted on annealed U-2.3% Nb are shown in Figs. 5 and 6. In the annealed condition the alloy consisted of a mixture of α-uranium and a niobium enriched γ phase, and had a 0.2% offset yield strength of 550 MPa.

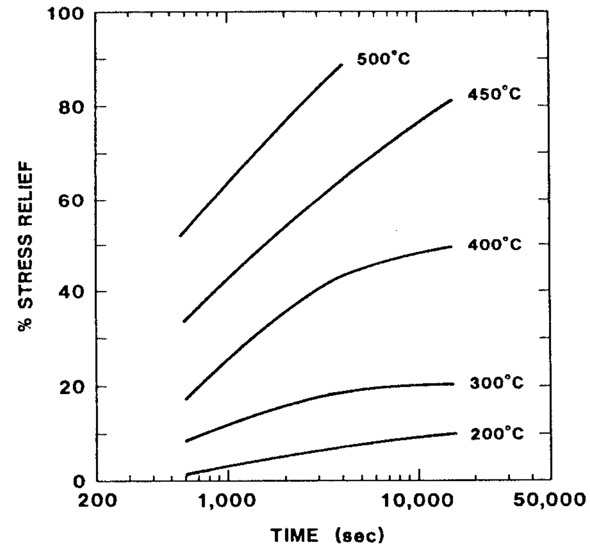

Fig. 4 The effects of temperature and time on stress relaxation in annealed unalloyed uranium (29).

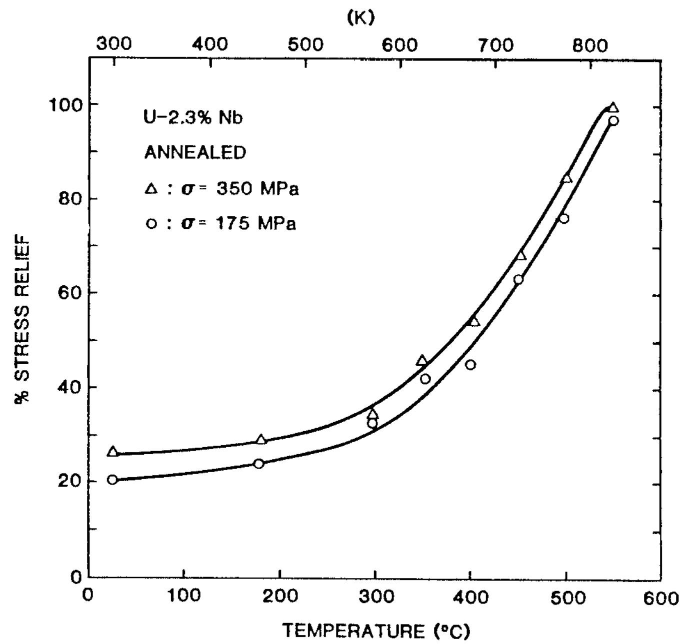

Fig. 5 The effects of temperature and stress magnitude on annealed uranium - 2.3% niobium stress relieved for six hours (yield stress = 550 MPa).

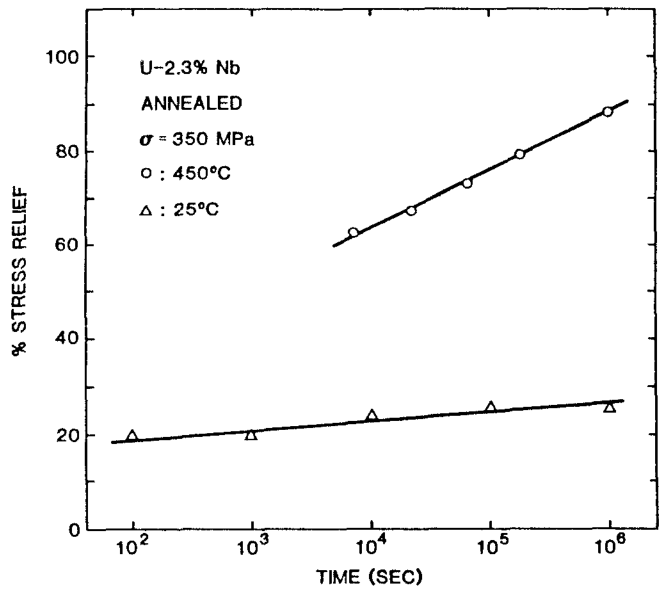

Fig. 6 The effects of time on the high and low temperature stress relieving processes in annealed uranium - 2.3% niobium.

Very little change occurred in microstructure or properties in the 25°C to 600°C range of stress relieving temperatures studied. This material exhibited two regions of stress relieving, one below about 300°C, and one above about 300°C. The percentage of stress relieved was relatively insensitive to the magnitude of the applied stress; and the stresses were completely relieved at about 550°C. The low temperature mechanism was found to be weakly dependent on both temperature and time. The activation energy for this process was shown to be consistent with that for athermal microplasticity in α-uranium. The high temperature mechanism was much more strongly dependent on temperature and time, and its activation energy was shown to be consistent with that for diffusional creep in α-uranium.

The results of similar experiments conducted on quenched U-2.3% Nb are shown in Figs. 7 and 8. Water quenching this alloy from 800°C produced a supersaturated metastable variant of the α phase termed α_b', which had a 0.2 offset yield strength of 730 MPa. Material in this condition underwent substantial changes in microstructure and mechanical properties in the stress relieving temperature range. Substantial age hardening occurred between 200°C and 350°C, and overaging occurred above 350°C due to cellular decomposition of the α_b'. The stress relieving behavior of this microstructurally unstable quenched material was similar to that of the microstructurally stable annealed material, except that both the onset of the high temperature mechanism and the completion of stress relieving occurred about 100°C lower. It was hypothesized that this acceleration of stress relieving was associated with microstructural instabilities in the quenched material, as has been observed in steels [31]. This hypothesis was tested by aging a series of quenched specimens at 350°C prior to stress relieving. This pre-aging produced a material whose microstructure was stable up to nearly the temperature of the pre-aging treatment. The results of stress relieving experiments conducted on these pre-aged samples are compared with those on as-quenched samples in Fig. 9. The fact that the onset of the high

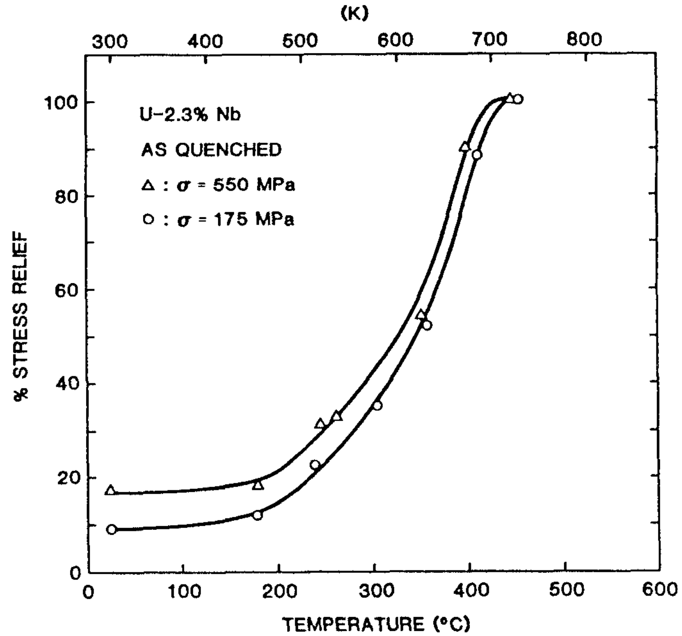

Fig. 7 The effects of temperature and stress magnitude on quenched uranium - 2.3% niobium stress relieved for six hours (yield stress = 730 MPa).

temperature mechanism was significantly retarded by pre-aging confirms that microstructural instabilities accelerate diffusional stress relieving in uranium alloys. In addition, these results show that low temperature stress relieving by athermal microplasticity is absent in age hardened material.

Similar stress relieving behaviors were found for U-2.0% Mo and U-0.75% Ti [30]. The U-0.75% Ti results are shown in Fig. 10. It was concluded that thermal stress relieving of microstructurally stable dilute uranium alloys occurs in the 300°C to 550°C range (a limited amount of athermal stress relieving also occurs spontaneously at room temperature), and that microstructural instabilities significantly accelerate the stress relieving process.

The results of these relatively fundamental studies are semi-quantitatively consistent with limited data available on engineering components. Figure 10 indicates that approximately 30 to 35% of the stresses remaining after athermal relief at room temperature should be relieved by aging at 380°C for six hours in U-0.75% Ti. Sachs boring measurements on hollow cylinders of this material in the as-quenched and quenched and aged conditions revealed that inner surface compressive stresses were reduced 25%, and maximum internal tensile stresses were reduced 40% by this aging treatment [13], as shown in Fig. 3.

CONSEQUENCES OF RESIDUAL STRESSES

Residual stresses can have substantial deleterious effects on the mechanical integrities of uranium alloy parts, and on the ease with which parts can be machined to close tolerances. Several specific examples of the consequences of

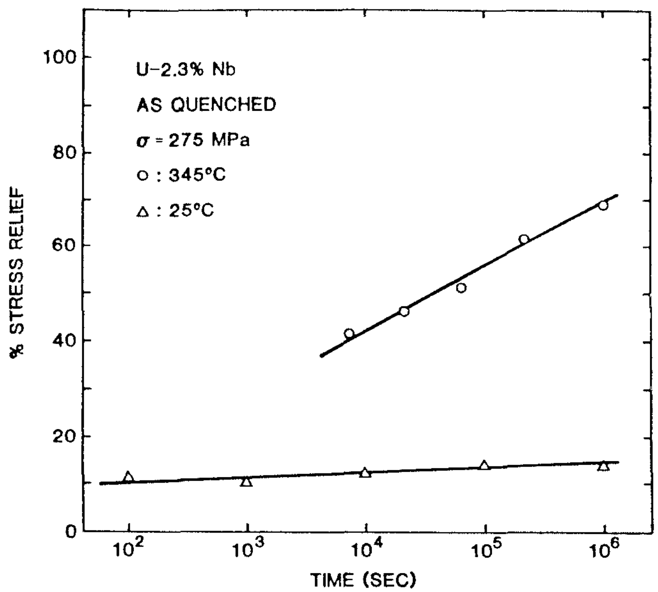

Fig. 8 The effects of time on the high and low temperature stress relieving processes in quenched uranium - 2.3% niobium.

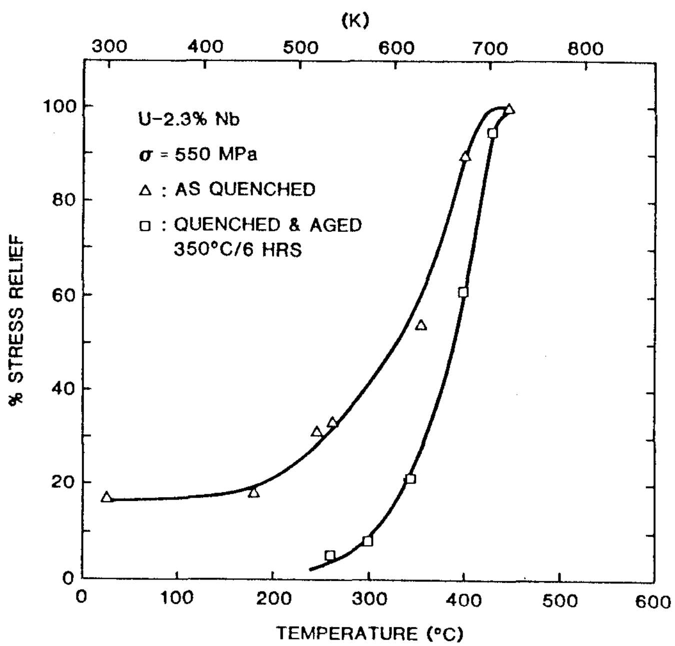

Fig. 9 Comparison of the behaviors of quenched and quenched plus aged uranium - 2.3% niobium stress relieved for six hours (yield stresses = 730 MPa, 1200 MPa, respectively).

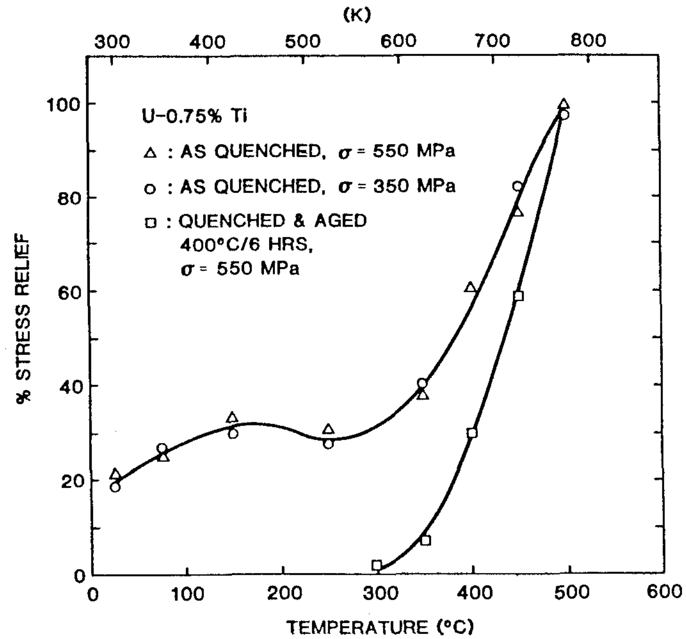

Fig. 10 The behavior of quenched and quenched plus aged uranium - 0.75% titanium stress relieved for six hours (yield stresses = 700 MPa, 1180 MPa, respectively).

residual stresses are presented in the following paragraphs.

Centerline Bursting

Centerline bursting commonly occurs when U-0.75% Ti bars greater than approximai 2.0 cm in diameter are plunge quenched from 800°C into water [10,13]. Radial heat extraction produces very high triaxial stresses along the centerline, as shown in Fig. 11, and since triaxial stress states typically promote fracture rather than plastic deformation these stresses often result in internal fractur and the formation of ∿0.5 mm voids. Quenching of plates or hollow cylinders, on the other hand, occurs by planar heat extraction and results in a less severe condition of biaxial tension which can more easily be accommodated by small amounts of plastic deformation. Hence void formation does not occur in plates. Centerline bursting in bars can be avoided by lowering the bar end first into the quench bath at a controlled rate [32] (typically ∿ 8 mm/sec). This results in semi-longitudinal, as opposed to radial, heat extraction, thus substantially decreasing the degree of residual stress triaxiality, and prevent void formation. While the formation of centerline voids has been reported and studied mostly in U-0.75% Ti, there is no reason to believe that its occurrence is peculiar to this alloy. It will probably occur in most uranium alloys giver a sufficiently severe state of triaxial quenching stresses.

Stress Corrosion Cracking

Uranium alloys are quite susceptible to environmentally enhanced cracking, and residual stresses have frequently been sufficient to cause this type of cracki

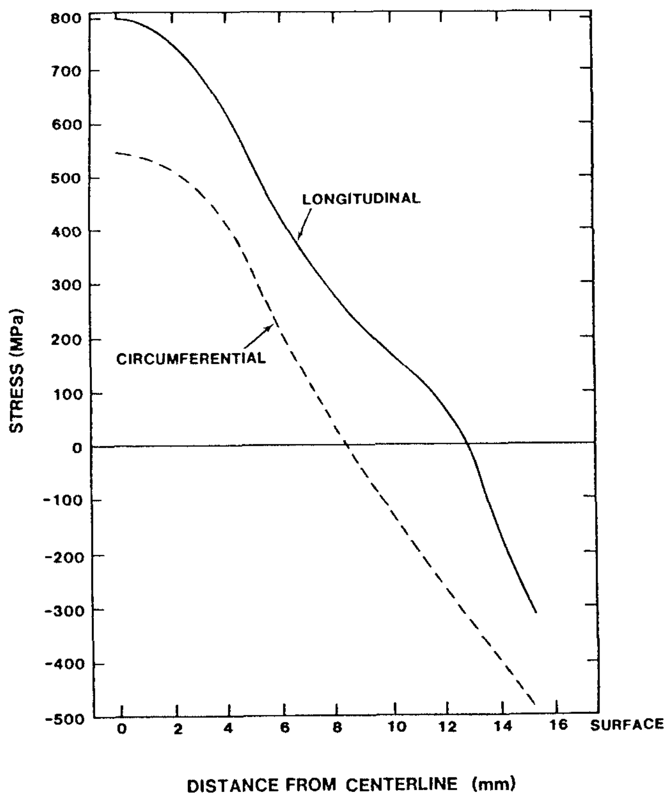

Fig. 11 Residual stresses determined by the Sachs boring method in a 35 mm diameter solid cylinder of U-0.75% Ti which had been plunge quenched into water from 800°C [10]. Note that maximum logitudinal tensile stress of 800 MPa is higher than the uniaxial yield strength of 650 MPa estimated for a quenched bar of this diameter [38].

to occur. An extensive body of literature is available in this area as well as two review articles [33,34], hence only a brief overview will be given here. Environments containing chloride ions appear to be the most devastating, but even environments as apparently benign as oxygen or high humidity air can cause substantial difficulties. Very dilute alloys, such as U-0.75% Ti and U-2.3% Nb, tend to be sensitive to moisture, apparently because of a surface reaction between water and uranium which produces embrittling hydrogen. More concentrate alloys, such as U-10% Mo and U-7.5% Nb-2.5%Zr, tend to be sensitive to oxygen. Threshold stress intensities for cracking range from $\sim$ 5 to $\sim$ 40 MPa ($m^{1/2}$), depending on material composition, heat treatment, and environmental severity. Under extreme conditions, however, the formation of high volume corrosion products in cracks can produce a wedging effect, resulting in substantially increased tensile stresses at the crack tip, and causing self-perpetuating stress corrosion crack growth, even in areas where macro-stresses are absent or compressive. This can result in complete failure of virtually unstressed engineering components.

As is the case with other alloy systems, stress corrosion susceptibility in

uranium alloys increases substantially with increasing material strength [35]. Interestingly, however, the age hardening process both decreases the threshold for stress corrosion cracking and decreases the magnitudes of the residual stresses which frequently provide the driving force for crack growth. Combining data on the effect of aging on K_{ISCC} [35] with data from Fig. 10 on the effect of aging on stress relief enables calculation of the effect of aging on the critical flaw size for stress corrosion cracking [30]. The results are shown in Fig. 12. It can be seen that the critical flaw size generally increases with aging temperature, indicating that relief of residual stresses on aging overrides the decrease in K_{ISCC}.

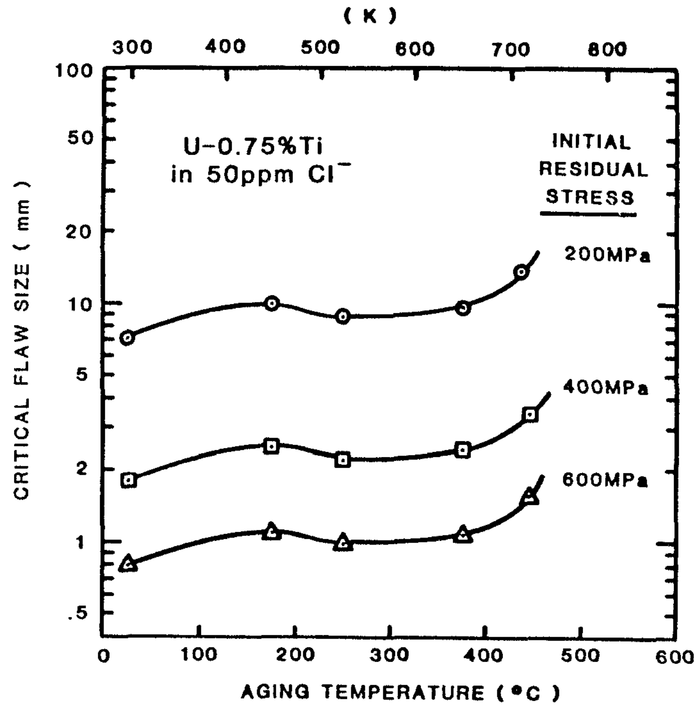

Fig. 12 The effects of aging on critical flaw size for a small flaw in the edge of a wide plate of uranium -0.75% titanium tested in 50 ppm chloride solution.

Machining Distortion

One very common consequence of residual stresses in uranium alloys is machining distortion. When residual stresses are introduced into parts by heat treating or forming operations, geometric distortions occur in response to the tensile and compressive stresses in order to minimize elastic strain energy. A plate which is quenched from one side, for example, develops compressive stresses on the quenched surface and tensile stresses on the slowly cooled surface, and bends in response to these stresses. If material is removed from such a part in a way that disturbs the balance of stresses, then additional distortion will occur as the stresses redistribute themselves to again minimize the elastic strain energy in the machined part. If the simple illustrative bent plate is clamped to a milling table in a way that prevents it from further distorting, and a substantial amount of compressively stressed material is removed from one side in order to produce a flat surface, the part will bend in a reverse sense when removed from the clamping fixture because the balance between the tensile and compressive stresses had been disturbed. This phenomenon makes it difficult to produce parts requiring very close dimensional tolerances from material which contains substantial residual stresses. Hence, stress relieving prior to machining is highly desirable when close tolerances need to be maintained.

Another problem which can occur is the introduction or modification of residual stresses during the machining operation due to small amounts of plastic deformation which occur under the tool in conjunction with the cutting process [14]. This is most common with soft alloys such as unalloyed uranium and as-quenched U-6% Nb, and results in additional dimensional tolerance problems similar to those just discussed. This problem can only be overcome by using careful machining practices designed to minimize deformation.

Dimensional instability due to shape memory effects is a related problem which occurs in some uranium alloys such as solution quenched U-6% Nb. Small amounts of plastic strain introduced into the material at room temperature by fabrication or machining processes can later be reversed or recovered by heating the material slightly [36]. This can result in substantial temperature induced dimensional instabilities in tightly toleranced parts. This shape memory phenomenon is associated with a martensitic phase transformation which occurs in the U-6% Nb and other uranium alloys slightly above room temperature [37]. It can be minimized by low temperature aging, which increases the martensitic transformation temperature without drastically altering the mechanical properties of the material [37].

SUMMARY

1. Residual stresses are generated in uranium and its alloys primarily by heat treatment and low temperature metal fabrication operations such as cold rolling, welding, and occasionally machining.

2. Residual stresses in uranium alloys have most frequently been measured by destructive relaxation methods. A neutron diffraction method has recently been demonstrated for non-destructive spacially resolved measurement of internal stresses in uranium alloy components.

3. Residual stresses in uranium alloys can be relieved by mechanical or thermal methods.

4. Residual stresses in uranium alloys can cause centerline bursting in quenched bars, stress corrosion cracking in components exposed to a variety of environments, and dimensional instabilities primarily during machining.

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