CREEP AND DIMENSIONAL STABILITY OF HIGH PURITY NIOBIUM ELECTRON BEAM WELDS

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Abstract

A study was conducted to characterize the microstructure of electron beam welds in high purity niobium and its effect on creep behaviour at room temperature. The parent material was 2mm sheet with a 50 μm grain size. The weld fusion zone had ~ 1mm grains, implying that these grains all intersected the free surface. The parent material showed no room temperature creep deformation below the yield stress, but room temperature creep of weld specimens caused up to 10% strain in the weld region at ~75% of the yield strength, over 1-2 months. Creep deformation was not smooth or continuous; the strain saturated at some value, and then after an incubation time, the strain increased and saturated again several times over 1-2 months. The magnitude of the strain for several specimens was similar but the creep deformation behavior was highly dependent on the actual microstructure and loading history. An initial prestrain with unloading shut down the creep deformation mechanism at the prior stress due to a dislocation locking effect. The local stresses in the weld fusion zone arose from anisotropic elastic interactions due to different crystal orientations that caused local regions to exceed the yield strength.

INTRODUCTION

Creep deformation at room temperature is very important for niobium superconducting RF cavities since it may affect the dimensional stability of cavities. Inelastic deformation of bcc metals is strongly rate and temperature dependent. The rate-controlling mechanism of plastic deformation in bcc metals is associated with the Peierls-Nabarro potential barriers to dislocation motion [1-2]. On the other hand, the deformation of the bcc materials is strongly affected by the interaction between the dislocations and solute atoms. Therefore, the rate-controlling mechanism is typically the thermal activation of mobile dislocations past interstitial solute atoms.

In general, the room temperature creep for metals like niobium is a consequence of time-dependent dislocation glide [3]. The creep strain rate, $\dot{\varepsilon}$, depends on the density of moving dislocation (ρ) and their average velocity (ν), as

$$\dot{\varepsilon} = \rho b v \qquad (1)$$

where b is the Burgers vector. Typically the creep strain rate is the highest at the beginning of the creep since this corresponds to the highest density and velocity of *mobile* dislocations [4]. The changes in dislocation density and arrangement take place as creep deformation proceeds, so that mobile dislocations are gradually trapped, causing the dislocation velocity to be reduced as a result of work hardening. Eventually, the creep rate is reduced to zero as the mobile dislocations are exhausted [5].

It has been observed that the extent of creep deformation is normally dependent on the magnitude of creep stress, loading history [4, 6]. In addition, the microstructure of materials might be another factor that may influence creep deformation. Since the weld region of niobium has very different microstructure than the parent materials, the creep deformation behaviour is expected to be different.

The objective of this paper was to study the influence of loading stress and microstructure on the room temperature creep deformation of niobium. Room temperature creep tests on as-received and electron-beam machining (EBM) weld niobium specimens are presented and discussed. The microstructures corresponding to as-received and EBM weld specimens are examined by optical microscopy and Orientation Imaging Microscopy.

EXPERIMENTS PROCEDURES

The material used in this work was commercial high purity niobium (RRR=150) sheet from Tokyo Denkai LTD. Two 2 mm thick niobium sheets were butt welded together using an electron beam weld procedure. Tensile and creep samples were electrical discharge machining (EDM) machined from the welded plate according to ASTM standard E8. Samples with the same dimension size were also made from the as-received 2 mm niobium sheet to compare with the weld specimens.

Tensile tests were conducted according to guidelines in ASTM standard test methods. The weld tensile specimens were EDM machined so the tensile axis was transverse to the weld direction, and the weld direction was parallel to the rolling direction (RD). The EB weld region is about 5.6 mm wide on the side under the beam, and width on the other side was about 2 mm. Thus the weld occupied about 20% of the specimen's 31 mm gage length. Creep samples were cut from a welded part in the same way as tensile specimens. The samples were machined into the standard "dog bone" shape, with a gage length of 25.4 mm, and gage width of 6 mm. The tensile tests were performed on a series 4200 Instron testing machine, model number 43K2. The creep tests were preformed on an ATS series 2300 lever arm creep testing system at room temperature. Strain was measured using two transducers that had better than 5 μm displacement resolution. The creep test was performed at room temperature starting at low stresses, and increasing them in a stepwise manner and held until creep deformation saturated.

The microstructure was observed in a light optical microscope under polarized light. Local crystal orientation data were collected using a Camscan 44FE with an hkl Technologies electron backscatter diffraction pattern (EBSP) mapping system (version 4.2), microtexture evaluation was determined by means of automatic indexing of Kikuchi patterns. A finite element model of the heat affected zone microstructure was built based upon a measured EBSP map, and the collected crystal orientation data were used to determine the anisotropic elastic constants for each grain orientation. The model was computationally deformed elastically to achieve a strain similar to the initial creep strain just after loading.

RESULTS

Microstructure

Fig 1 shows that weld specimens have very large grain size (~1mm) in the weld zone; the grain size decreased with increasing distance from the weld. 80 mm away from weld zone, the grain size was the same as the parent material, 50μm.

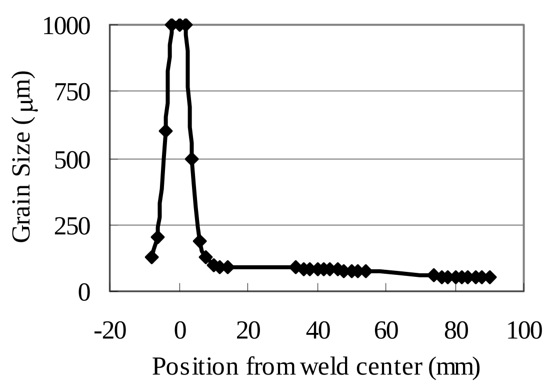

Fig.1. Grain size distribution in weld specimens

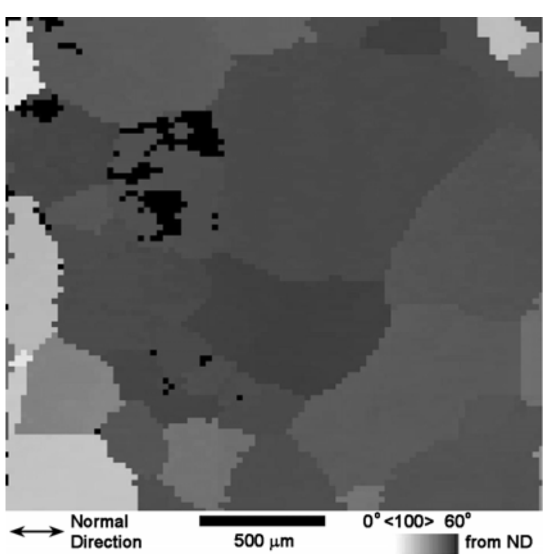

Fig.2. Crystal orientation map of full thickness of sheet shows that grains on surface have orientations with <100> close to the normal direction. The edge of the fusion zone is 2 mm above the map.

In the heat affected zone in Figure 2, the grain size is larger than the parent material, but smaller than in the fusion part of the weld. The crystal orientations present in the heat affected zone are similar to the parent material.

Mechanical properties

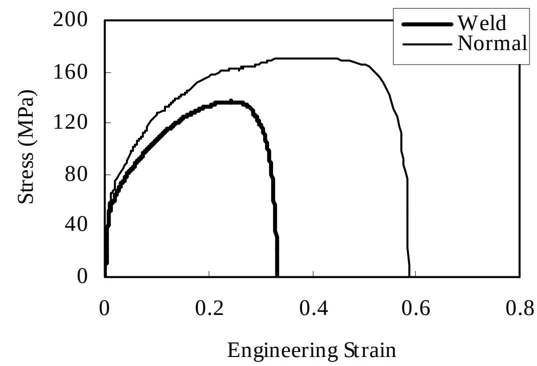

Fig.3. Stress-strain curves of as-received and weld niobium tested at a strain rate of 3×10^{-3} s^{-1}.

The yield strength of as-received specimens was slighty higher than butt-weld specimens (58.4MPa vs 51MPa), but the ultimate tensile strength (UTS) of as-received specimens was ~30MPa higher than the butt-weld specimens, and elongation was ~25% higher.

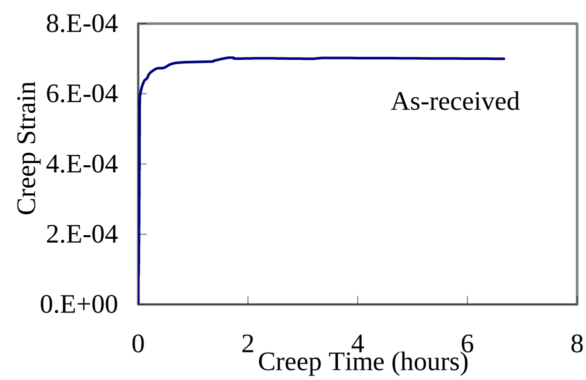

Fig. 4. Creep curve of as-received niobium at stress level 55MPa

There was no significant creep deformation observed in as-received niobium specimens near the yield stress level (55MPa) at room temperature as shown in Fig. 4. Fig. 5 shows the results of two creep experiments with histories of two weld samples plotted. In a preliminary test, the sample was initially deformed with 25 MPa, where no measurable strain occurred in 1 hour, unloaded and reloaded to 45 MPa, where a strain of 0.18% occurred in 2 hours. Thus, the second specimen was initially deformed with a stress in between, starting at 36 MPa, and given an increasing stress history with time, which included two full unloading events. The feedback system for the creep frame was able to maintain the load within 1

MPa. The third specimen was deformed with a similar stress history, but never unloaded. The strains measured were much larger. Dimensional changes due to creep were very small and failure did not occur in either case. The creep deformation became saturated after some time. Creep deformation usually restarted after a stress increase, but sometimes it increased part way through a constant stress hold time. The creep rate of the weld samples was ~10^{-8} /s and the creep rate generally increased with stress.

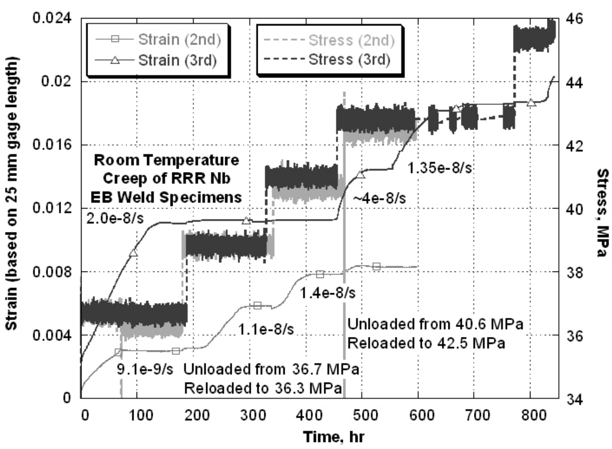

Fig. 5. Creep curves of weld niobium specimens

DISCUSSION

No significant room temperature deformation was observed in as-received niobium specimens, perhaps due to the small grain size. The mobile dislocations rapidly pile up against a grain boundary, which is a hardening process that decreased the mobile dislocation density very fast. Due to high purity and large grain size in the weld specimen, the mobile dislocation densities must have been much higher than in the as-received niobium specimen. This made it much easier for the weld specimen to creep at room temperature. Originally fairly uniform but low dislocation densities are converted into very inhomogeneous defect structures, so there are some areas which have a high density of dislocation tangles and networks, whereas there are others which are relatively dislocation free, due to heterogeneous stress states in big grains [7]. Since all dislocations have stress fields associated with them, this implies that the state of internal stress in the specimen is not uniform. Dislocation motion is driven by the difference between the applied stress and these internal stresses, so the inhomogeneous nature of the stress field can exhibit itself in some strange ways [8].

It is well known that the exhaustion of creep deformation can be the result of immobilization of the mobile dislocations. These dislocations are either locked by interstitial solute atoms and grain boundaries or tangled with the dislocation network in the niobium. To re-activate the immobilized dislocations after saturation, a higher stress is needed for continuing deformation. Once the stress reaches a critical level, a large number of trapped dislocations may become mobile at the same time, or a new source may be started, that allows creep to resume. However, since the loading stress oscillated by 0.5MPa, it is possible that restart of creep deformation was caused by a slight over stress that caused a re-organization of a locked source or activated a new source.

The anisotropic elastic FEM deformation simulation using the crystal orientations and geometry in Figure 2 showed that the local stress distribution depends strongly on the orientation and misorientation of neighbouring grains. High stress concentrations appeared near the grain boundaries with a large misorientation angle between grains. The highest von-mises stress reached in the weld was 41.6 MPa when a nominal stress of 20 MPa tensile stress was applied as a boundary condition. The stress distribution is highly dependent on the misorientation and grain size of niobium which could lead to big differences in creep deformation behaviour in different weld specimens.

CONCLUSIONS

There is significant room temperature creep deformation observed in welded niobium while there is no significant creep deformation observed in as-received niobium, which could be attributed to the grain size effect. The creep deformation is highly dependent on the microstructure and loading history. The misorientation of grains will increase stress concentration on grain boundaries and cause larger creep deformation. The creep may be shut down by reloading following the exhaustion of creep deformation at a stress equal to previous one due to dislocation-locking effect.

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