Influence of Rotary Swaging on Creep Behaviour of P92 Steel

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Abstract. This work investigates the influence of rotary swaging on creep behaviour and microstructure heterogeneity of P92 martensitic steel. Its as-received coarse-grained state was processed by rotary swaging at room temperature. The bar of 30 mm in diameter was reduced at room temperature down to 15 mm by rotary swaging. The value of equivalent strain imposed during rotary swaging was about 1.4. Microstructures were investigated by scanning electron microscope Tescan Lyra 3 equipped with NordlysNano EBSD detector operating at an accelerating voltage of 20 kV with specimen tilted at 70°. It was found that the processing by rotary swaging led to the formation of heterogeneous microstructure containing a mixture of fine nearly equiaxed grains and large grains, which were significantly, elongated parallel to the swaging direction. Constant load tensile creep tests in tension and compression were conducted at 923 K under different applied stresses. In order to evaluate the influence of microstructural heterogeneity on the creep behaviour, the creep tests were performed both in longitudinal and also transverse directions to the rotary swaging. The results of tensile creep tests revealed that specimens manufactured in the direction parallel to the rotary swaging exhibited significantly higher creep life and strain to fracture in comparison with the results measured in the transverse direction. The creep results showed that creep tests performed at medium stresses exhibit a pronounced minimum creep rate. The microstructure results suggest that creep behaviour depends on the microstructure coarsening during creep testing and the orientation of the grain boundaries with respect to the applied stress.

1. Introduction

The rotary swaging (RS) is one of the severe plastic deformation (SPD) methods [1-3] that enables to impose relatively large strain into material. Thus, the standard coarse-grained microstructure is refined by subsequent transformation of subgrain- into grain- boundaries [4,5]. Usually the microstructure processed by SPD at room temperature is homogeneous and completely ultrafine-grained (UFG) when the equivalent strain higher than 10 is imposed in the material. However, the strain imposed during rotary swaging is not so high compared to other SPD methods [6], such as high-pressure torsion (HPT) or equal-channel angular pressing (ECAP).

It was observed that SPD-processed microstructures with the imposed strain about one are heterogeneous and contain predominantly low-angle grain boundaries (LAGBs). These materials may exhibit higher creep life than its coarse-grained undeformed state [7,8]. However, the creep properties of SPD-processed materials are usually tested only in one section, usually parallel to the direction of deformation [7]. In other sections, the creep behaviour may be different, especially at low imposed
strains when the microstructure is heterogeneous. In the present time there are no works investigating creep behavior of RS-processed materials in different sections. The aim of this work is comparison of creep behavior of RS-processed P92 steel in the different sections with respect to the direction of rotary swaging.

2. Experimental material and procedures
The experimental material used in present work was advanced tungsten modified 9%Cr P92 steel. The chemical composition is shown in Table 1. The as-received coarse-grained state was normalized at 1323 K/60 min/air and tempered at 1013 K/140 min/air. This state is referred to herein as the received state.

| Element | C  | Cr  | Mo  | W   | Si  | Mn  | V   | Nb  | P   | N   | Al  | S   |
|---------|----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| wt.%    | 0.11 | 8.58 | 0.33 | 1.67 | 0.37 | 0.48 | 0.23 | 0.06 | 0.013 | 0.037 | 0.017 | 0.005 |

The von Mises equivalent strain imposed by rotary swaging [9,10] was estimated by $\varepsilon_{eq} = \ln(D_0/D_n)^2$, where $D_0$ is the initial diameter ~ 30 mm and $D_n$ is the final diameter ~ 15 mm after application of RS at room temperature. The imposed equivalent strain was about 1.4.

Figure 1. Scheme of arrangement of tensile specimens on the RS-processed rod; a) longitudinal and b) transverse direction
The tensile (Figure 1) and compression specimens were manufactured in longitudinal and transverse sections with respect to the rotary swaging direction. Constant load tensile creep tests were conducted at 923 K in protective argon atmosphere. For tensile creep tests were used flat specimens with the gauge length of 10 mm and the cross section of 3 x 1 mm² cut in the longitudinal section. The flat tensile specimens cut from transverse section had the gauge length of 4 mm and the cross section of 3 x 1 mm². The tensile creep test at constant load were run up to the final fracture. Uniaxial compression tests at constant stress were conducted on cylindrical specimens with initial height of 10 mm and diameter of 5 mm. The compression creep tests were interrupted at creep strain about 0.2 – 0.25. Microstructure investigations were performed using a scanning electron microscope (SEM, Tescan Lyra 3 equipped with NordlysNano EBSD detector operating at accelerating voltage of 20 kV with specimen tilted at 70°). EBSD was used to determine the misorientations \( \theta \) between neighbouring grains. Misorientation \( \theta = 15° \) was taken to distinguish HAGBs from LAGBs.

3. Results and discussions

3.1. Creep behaviour

Figure 2 shows the comparison of tensile creep test of specimens manufactured in the longitudinal and transverse directions. The results demonstrate that transverse creep specimens exhibit faster minimum creep rate \( \dot{\varepsilon}_{\text{min}} \), shorter creep life and lower ductility in comparison with longitudinal creep specimens. The longitudinal creep specimens also exhibit increase of fracture strain \( \varepsilon_f \) with a decrease of applied stress from 150 MPa to 80 MPa. However, the opposite results can be found in the creep of transversal tensile specimens.

![Figure 2. Tensile creep curves for longitudinal and transverse RS-processed specimens; a) strain rate vs. strain; b) strain rate vs. time.](image)

One can see that specimen parallel to RD tested at 80 MPa exhibits a pronounced softening after reaching of minimum creep rate. The softening slows down in the strain region of about 0.1 – 0.3. In this region the largest creep strain occurs.

The pronounced softening after reaching of \( \dot{\varepsilon}_{\text{min}} \) is also seen in the strain rate – strain dependence (Figure 2) for transverse specimen tested at 80 MPa. However, after the pronounced softening, there is no slowing down of the creep rate but fracture processes occur.
The differences in creep behaviour of tensile specimens manufactured in longitudinal and transverse sections can be caused by microstructure heterogeneity that leads to faster fracture processes in transverse specimens.

Figure 3 shows the compression creep tests at constant stress of longitudinal and transverse specimens. The compression tests were interrupted at creep strain about 0.2 – 0.25.

Although the compression creep tests are not influenced by fracture, the minimum creep rates $\dot{\varepsilon}_{\text{min}}$ of longitudinal specimens are about order of magnitude slower compared to transverse compression specimens.

In Figure 3 is also shown the stress change test of longitudinal specimen. The stress was increased from 100 to 150 MPa at creep strain about 0.05. The creep results demonstrate that $\dot{\varepsilon}_{\text{min}}$ of longitudinal specimen after stress increase from 100 to 150 MPa at strain of about 0.05 is faster than $\dot{\varepsilon}_{\text{min}}$ of longitudinal specimen tested only at 150 MPa. The faster $\dot{\varepsilon}_{\text{min}}$ after strain about 0.05 can be explained by microstructure changes occurring during creep testing, such as microstructure coarsening [6,8].

![Figure 3. Strain rate vs. strain curves for longitudinal and transverse specimens tested in compression at 923 K.](image)

3.2. Microstructure after creep

Figure 4 shows the microstructure processed by RS and after subsequent creep testing at 100 MPa in the longitudinal and transverse sections. In Figure 4a are shown the directions of applied stress in the case of longitudinal ($\sigma_L$) and transverse ($\sigma_T$) specimens. One can see that the creep testing led to the significant coarsening of microstructure. The tensile longitudinal specimen reached fracture strain about 0.34 (Figure 2a). However, the microstructure observed in transverse section after RS processing and after subsequent creep showed no significant grain coarsening during creep. The tensile specimen tested at 100 MPa reached fracture strain only about 0.15. For this reason, the higher fracture strain in longitudinal specimen compared to transverse specimen can be caused by significant coarsening of microstructure during creep testing.

Figure 4d shows the cumulative frequency of boundaries measured in longitudinal and transverse section of P92 steel processed by RS and after subsequent creep. One can see that creep testing led to the decrease of LAGBs in the microstructure. The decrease of LAGBs, formed by RS processing at room temperature, during creep testing leads to the predominance of HAGBs in the microstructure. It is seen that decrease of LAGBs after creep is larger in the coarsened microstructure of longitudinal
section than in microstructure of transverse section. For this reason, the contribution of grain boundary mediated processes to the total creep strain is higher [11-13] in the longitudinal compared to the transverse specimen.

Figure 4. Microstructure of P92 steel; a) processed by RS (longitudinal section), b) after subsequent creep at 100 MPa (longitudinal section), c) after subsequent creep at 100 MPa (transverse section) and d) cumulative frequency of boundaries measured in longitudinal and transverse section.

Figure 5 shows the cavities near fracture surface in longitudinal RS-processed specimen tested at 150 MPa. It is seen that cavities are predominantly formed along boundaries parallel to the stress axis. Although the cavities often nucleate on boundaries which are transverse to the stress axis. In the specimen manufactured in the longitudinal section cavities are distributed along boundaries parallel to the stress axis. It can be suggested that the nucleation of cavities is associated with the high local stress concentration at large carbides distributed along grain boundary [11,14].
Figure 5. Cavitation near the fracture surface in longitudinal tensile specimen tested at 923 K and 150 MPa; a) long cavities approximately parallel to the stress axis, b) formation of cavities along boundaries. Stress axis is horizontal.

Figure 6 shows the cavities near fracture surface in the transverse RS-processed specimen tested at 150 MPa. The results demonstrate that near fracture surface can be seen many small cavities. These small cavities are formed at grain boundary triple points (Figure 6b) and some of them look like wedge-type cracks. One can see that cavities propagate and interlinked together along transverse grain boundaries. The previous studies suggested that the cavities at triple points can be caused by grain boundary sliding insufficiently accommodated by high local plastic deformation [15-17]. However, the fracture strain of transverse specimen significantly decreases with decreasing value of applied stress.

Figure 6. Cavitation in transverse tensile specimen tested at 923 K and 150 MPa; a) cavities near the fracture surface, b) formation of wedge cracks at triple points, c) interlinkage of small cavities along boundaries approximately perpendicular to the stress axis. Stress axis is horizontal.
It can be caused by the fact that boundaries of elongated grains formed during RS-processing (Figure 4a) are, in the case of transverse specimen, predominantly perpendicular to the stress axis. Many works [11,14] reported that cavities nucleate and propagate particularly along boundaries transverse to the stress axis. For this reason the formation and grow of cavities can be easier in the transverse specimen which led to the premature fracture at lower creep strain compared to longitudinal specimen.

4. Conclusions
Creep tensile and compression tests were conducted on an advanced creep-resistant 9% Cr P92 steel to evaluate the creep behaviour of RS-processed steel in longitudinal and transverse section. The creep tests were performed at 923 K and with an applied stress range from 80 to 150 MPa.

The main results are as follows:

1. The RS-processed P92 steel exhibits shorter creep life and lower ductility in transverse section compared to the longitudinal creep specimens.

2. The higher creep strain in longitudinal tensile specimens led to the significant coarsening of microstructure.

3. The formation of cavities and resulting ductility of RS-processed P92 steel depends on the orientation of the grain boundaries with respect to the applied stress.

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