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CORRELATION OF SUBSTRUCTURE WITH
MECHANICAL PROPERTIES OF PLASTICALLY
DEFORMED AISI 304 AND 316 STAINLESS STEEL

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PROGRESS REPORT

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The optical micrograph on the cover of this report is from a 304 stainless steel tensile specimen tested at 650°C, $\dot{\epsilon} = 8.33 \times 10^{-5} \text{ sec}^{-1}$. The magnification is 200X. Intergranular microcracks and the deformation of grains adjacent to the cracks are clearly visible. An interesting feature of this micrograph is the formation of microcracks at the twin boundaries within the grains located in the upper left-hand portion of the picture. The particular test conditions experienced by this specimen were apparently favorable for not only grain boundary microcrack formation but for twin boundary crack formation as well.

TABLE OF CONTENTS

	Page
LIST OF TABLES	iii
LIST OF FIGURES	iv
SUMMARY	vi
OBJECTIVE	1
INTRODUCTION	1
EXPERIMENTAL PROGRAM - PHASE II	2
Tensile Properties (316 SS)	2
Hot Microhardness (316 SS)	9
Cavitation (304 SS)	11
Literature Review	12
Experimental Results	17
PLANS FOR FUTURE WORK	21
REFERENCES	22
DISTRIBUTION LIST	45

LIST OF TABLES

Table

Page

- I Tensile and Substructure Properties
 of 316 Stainless Steel, Heat No.
 332990

25

LIST OF FIGURES

<u>Figure</u>		<u>Page</u>
1	Type 316 Stainless Steel Tensile Test Specimens, Heat No. 332990, As-received.	26
2	Substructure of Shoulder of 316 Stainless Steel Tensile Specimen 7189.	27
3	Representative Deformation Substructure of Tensile Tested 316 Stainless Steel, Heat No. 332990 (21°C to 593°C).	28
4	Representative Deformation Substructure of Tensile Tested 316 Stainless Steel, Heat No. 332990 (650°C to 816°C).	29
5	Representative Dislocation Configurations in Tensile Tested 316 Stainless Steel, Heat No. 332990 (650°C to 816°C).	30
6	Relation between Mobile Dislocation Density and Maximum True Tensile Stress.	32
7	Relation between Average Subgrain or Dislocation Cell Intercept and Maximum True Tensile Stress.	33
8	Relation between Mobile Dislocation Density and Average Subgrain Intercept.	34
9	Average Subgrain or Dislocation Cell Intercept as a Function of Tensile Test Temperature.	35
10	Relation between Strain Hardening Exponent and Average Subgrain or Dislocation Cell Intercept.	36
11	Relation between Strength Coefficient and Average Subgrain or Dislocation Cell Intercept.	37
12	Hot Microhardness of 316 Stainless Steel Tensile Test Specimen (7189) Shoulder as a Function of Temperature.	38

13	Microcrack Length Distribution as a Function of Crack Length for 304 Stainless Steel at 650°C.	39
14	Measured Microcrack Distribution as a Function of Angle from the Specimen Axis for 304 Stainless Steel at 650°C.	40
15	Representative Micrographs of Crack Formation for 304 Stainless Steel Tensile Tested at 650°C, $\dot{\epsilon} = 8.33 \times 10^{-4} \text{ sec}^{-1}$.	41
16	Representative Micrographs of Crack Formation for 304 Stainless Steel Tensile Tested at 650°C, $\dot{\epsilon} = 8.33 \times 10^{-5} \text{ sec}^{-1}$.	42
17	Representative Micrographs of Crack Formation for 304 Stainless Steel Tensile Tested at 650°C, $\dot{\epsilon} = 8.33 \times 10^{-6} \text{ sec}^{-1}$.	43
18	Example of Wedge-type Microcrack in 304 Stainless Steel at 650°C.	44

SUMMARY

A program to evaluate the substructure of austenitic stainless steels (316 and 304) deformed at elevated temperatures in the creep, tensile, and fatigue modes is being conducted. Transmission electron microscopy is the primary tool employed in this research. Complementary work using optical microscopy and immersion density is presently underway to characterize the grain boundary microcracks and bulk density change observed subsequent to deformation in these materials.

The substructure resulting from tensile deformation is found to contain dislocation cells and evidence of planar slip for deformation below $0.5 T_m$ and subgrains for deformation above $0.5 T_m$. The substructure of the specimen tested at $\sim 0.5 T_m$ exhibits evidence of subgrain formation in regions near grain boundaries. Electron micrographs representative of the substructure are presented.

The average subgrain and dislocation cell intercepts, d , and mobile dislocation density, ρ_m , are related to the maximum true tensile stress, σ , by $d \propto \sigma^{-1}$ and $\rho_m^{1/2} \propto \sigma$, respectively.

The hot-microhardness of the shoulder portion of the tensile specimen tested at 21°C (70°F) shows a definite

recovery of prior, unrecovered deformation. This result is in agreement with the substructure of the shoulder which indicates the presence of unrecovered deformation.

A detailed literature review on intergranular crack formation during the elevated temperature plastic deformation of polycrystalline metals was completed.

A condensed version of this review is presented.

The microcrack length distribution of 304 stainless steel tensile tested at 650°C and at three strain rates indicates that the majority of cracks are one grain boundary facet in length. The measured crack percentage angular distribution as a function of angle from the tensile axis shifts toward higher angles with decreasing strain rate. Micrographs which illustrate the observed cracks are presented.

OBJECTIVE

The objective of this program is to evaluate (a) the time, temperature, and stress dependent properties of austenitic stainless steel and (b) the relationship of these properties to the substructure as observed principally with a 200 kV transmission electron microscope. Special emphasis has been given to operating conditions typical of LMFBR applications, including the knowledge that the radiation environment can influence the substructure of these metals leading to significant changes in the mechanical property behavior. Related substructure studies on other alloy systems will also be considered.

INTRODUCTION

Detailed transmission electron microscopy has been completed on 316 stainless steel tensile specimens from a second reference heat tested at temperatures from ambient to 816°C. These results are in substantial agreement with the results from the Phase I portion of this research program. The creep-rupture specimens from the second reference heat are now being investigated.

Hot microhardness testing has been continued. Results are reported for 316 stainless steel from the second reference heat where the shoulder portion of one

tensile specimen served as the hot microhardness specimen.

Optical metallography was used to continue the investigation of microcrack formation in tensile and creep-rupture tested 304 stainless steel. Additional results for the tensile specimens are now presented.

EXPERIMENTAL PROGRAM - PHASE II [D. J. Michel and J. Moteff]

Tensile Properties (316 SS)

Preliminary substructure results for 316 stainless steel tensile specimens fabricated and tested to failure by WADCO (Reference Heat No. 332990) have been presented in a previous report.⁽¹⁾ These tensile specimens, as well as creep-rupture specimens from the same heat, constitute a study complimentary to the Phase I portion of this program (316 stainless steel from PNL Reference Heat No. 65808).^(2, 3, 4) The tensile specimens are of particular interest because they represent a coherent group of data for the temperature range from 21°C (70°F) to 816°C, and thus serve to extend the range of experimental conditions for which substructure data are available.

Figure 1 shows the tensile specimens as-received and prior to the removal of thin transverse slices from the gage section of each specimen for transmission electron microscopy. Previously discussed experimental procedures for electrochemical thinning, microscope

examination, and the extraction of substructure information from the micrographs were used in this work. (2) Longitudinal sections were also removed from the shoulder portion of specimen 7189 tested at 21°C (70°F). The shoulder is normally considered to have experienced no deformation due to testing and should therefore be representative of the pre-test substructure. Figure 2 shows the typical substructure of this specimen shoulder. The high dislocation density and {111} slip traces are the result of deformation received by this material at some point in its fabrication or production history, possibly the final swaging operation prior to tensile specimen machining, which has not been completely recovered by the subsequent heat treatment. The pre-test substructure of these specimens is clearly not characteristic of a well annealed material. These observations may also account for the differences in the reported room temperature ultimate tensile strength and yield strength of this material (reference 1, page 12) when compared to that for the previously studied PNL material (reference 2, page 25), even though the reported grain size, total elongation, and reduction in area values for both heats are comparable.

A summary of the tensile test results and substructure data for the WADCO specimens is given in Table I. These substructure results are consistent with those

previously reported for 316 stainless steel subsequent to tensile and creep deformation.⁽²⁾

The substructure characteristic of the specimens tested at temperatures in the range from 21°C (70°F) and 593°C (1100°F) is shown in Figure 3. Dislocation cells, whose walls are composed of massive dislocation tangles, were observed at all temperatures. The average cell intercepts increase gradually with increasing test temperature (Table I). Also evident were slip traces characteristic of $\{111\} \langle 110 \rangle$ slip, Figure 3a, whose tendency decreased with increasing temperature. The absence of profuse intragranular carbide precipitation on dislocations, which acts to hinder cell formation at similar temperatures during creep testing,⁽²⁾ is considered to be responsible for promoting the formation of the dislocation cells observed in these specimens.

Figure 4 illustrates the substructure of specimens tested between 650°C (1200°F) and 816°C (1500°F). The apparent early stages of subgrain formation by the rearrangement of the dislocation cell walls into sub-boundaries parallel to $\{111\}$ traces was evident in the specimen tested at 650°C, especially near existing grain boundaries (Figure 4a). At these locations, the critical dislocation density has apparently been achieved near active $\{111\}$ slip planes and at the proper temperature to promote the reduction of

the high dislocation cell wall strain energy by cross slip and climb processes resulting in the formation of low strain energy sub-boundaries and subgrains somewhat larger than the previous cells.

The precipitation of large carbides along grain boundaries was found to be the greatest for the specimen tested at 650°C. The presence of these carbides and numerous surface fissures led to preferential electrolyte attack at these sites during electropolishing of the specimens tested at and above 650°C (and at 593°C).

Distinct subgrains were formed at test temperatures of 704°C (1300°F) and above (Figures 4b through d). The subgrain boundaries at these temperatures were found to consist of the previously illustrated hexagonal dislocation networks formed by $b = a/2 \langle 110 \rangle$ type screw dislocations lying on $\{111\}$ planes.⁽²⁾

Examples of the dislocation configurations within the subgrains formed between 650° and 816°C are shown in Figure 5. The predominant dislocation Burgers Vector was found to be of the type $a/2 \langle 110 \rangle$ with both edge and screw dislocations present. The formation of loops is evident at positions designated A in Figures 5a and c. The loops range in diameter from approximately 400 Å at 650°C to approximately 1200 Å at 816°C. The formation of the dislocation configuration termed the "Double-Triple Node"⁽²⁾ was also observed

in these specimens, one example of which is shown at the position designated B in Figure 5d. The hexagonal network characteristic of sub-boundaries is evident near the lower edge of Figure 5c and at the lower left corner of Figures 5b (out of contrast) and d.

It is interesting to note the region somewhat "denuded" of dislocations immediately adjacent to the sub-boundary in Figure 5c. This denuding may be the result of the attraction of the dislocations formerly in this region to the sub-boundary during its formation by cross slip and climb processes.

No clear evidence of stacking faults was found for these specimens even though some indications of the separation of partials were observed.

The density of the dislocations not associated with the sub-boundaries (termed the mobile dislocation density), ρ_m , could be estimated with reasonable confidence for only those specimens tested at temperatures of 704°C (1300°F) and above. The foil areas used for these estimates were those where the maximum number of dislocation images were visible, obtained by tilting the foil to achieve the proper reflecting conditions. The average foil thickness was found to be 2500 Å using an extinction contour method.⁽⁵⁾ The resulting values of ρ_m were found to be related to the maximum true stress, σ , according to the relationship

$$\rho_m^{1/2} \propto \sigma \quad (1)$$

shown in Figure 6. For comparison purposes, equivalent results obtained from the 316 stainless steel creep specimens (PNL Reference Heat No. 65808)⁽²⁾ are also plotted in this figure.

Figure 7 shows that the accepted relationship between average subgrain (or dislocation cell) intercept, d , and the maximum true stress, σ ,

$$d \propto \sigma^{-1} \quad (2)$$

describes the present results. The previously reported creep results⁽²⁾ have been used again for comparison.

The quantities ρ_m and d for the WADCO tensile specimens are also related by a function of the form

$$\rho_m \propto d^{-2} \quad (3)$$

as shown in Figure 8.

Figure 9 is a plot of the average subgrain and dislocation cell intercepts versus tensile test temperature. The gradual increase in cell intercept with increasing temperature between 21°C and 593°C in contrast to the rapid increase in subgrain intercept above 650°C (i.e. above $\sim 0.5 T_m$) is clearly evident. These observations combined with the foregoing results suggest that the strain hardening capability of the material remains relatively constant

below 650°C but decreases in some proportion to subgrain intercept above 650°C. The tensile test results given in Table I agree with this suggestion. If a plot of strain hardening exponents versus the subgrain or dislocation cell intercept values (Table I) is now constructed, Figure 10, the result is generally consistent with that recently proposed primarily on the basis of fatigue results for the PNL material (316 stainless steel Reference Heat No. 65808).⁽⁶⁾ A comparison plot of strength coefficient against the subgrain or cell intercept values in Table I, Figure 11, is also similar to that proposed from the PNL material tensile specimen data. Examination of Figure 10 indicates that the plot not only approaches a constant strain hardening exponent at low temperatures (small subgrain intercepts), as previously suggested,⁽⁶⁾ but may also approach a constant subgrain intercept at higher temperatures ($\lim_{T \rightarrow T_m} d \rightarrow d_{\max}$). A similar behavior for the strength coefficient is suggested by Figure 11. The differences in the pre-test substructure of the two stainless steels, as discussed earlier in this report, may contribute to the degree of uncertainty concerning the curvature of Figures 10 and 11. This uncertainty is believed to be relatively small, however, due to the reasonable agreement between the results from both materials in Figures 6, 7, and 8.

The deviations in the linearity of Figures 10 and 11, suggested by the WADCO tensile specimen results, imply that the Hollomon-type equation,⁽⁶⁾ proposed for the PNL material tensile results on the basis of the fatigue and tensile data for this material plotted in both figures,

$$\sigma_t = 60d^{-\frac{4}{3}}(\epsilon_t)^{\frac{0.0002}{d}} \quad (4)$$

may require modification to adequately describe the deformation behavior of 316 stainless steel at both high and low temperatures. However, the prediction that a small subgrain intercept (or size) results in both a greater strength coefficient and a greater strain hardening exponent⁽⁶⁾ is clearly substantiated by the WADCO tensile specimen results.

The average misorientation between adjacent subgrains, as determined from diffraction pattern Kikuchi line shifts across sub-boundaries, could be measured for only those specimens tested at 760°C and 816°C. Evidence for misorientation was observed for the specimens tested at lower temperatures but could not be measured due to the small subgrain or dislocation cell size.

Hot Microhardness (316 SS)

In the previous section of this report, the substructure of the shoulder portion of the WADCO tensile specimen tested to failure at 21°C was interpreted to indicate incomplete

recovery of pre-test deformation. To examine this condition further, the hot microhardness of a second shoulder portion of this specimen was determined between room temperature and 800°C.

Figure 12 shows a plot of the hot microhardness as a function of temperature for this specimen. The data are from one ascending and descending temperature test (two indentations averaged for each temperature). The continuous curve represents the hot microhardness of 316 stainless steel sheet specimens (PNL reference heat material) from one ascending and descending temperature test.^(7, 8) The sheet specimens received 20% cold work followed by an anneal at 980°C for one hour in H₂ prior to the hot microhardness test. (It has been demonstrated in a previous report⁽¹⁾ that the hot microhardness of the PNL material and the ultimate tensile stress of the WADCO specimens exhibit a direct-proportion relationship.)

Examination of Figure 12 indicates that the WADCO material has undergone further recovery during the hot microhardness test. This result reinforces the substructure interpretation previously given for this specimen shoulder, assumed to have received no deformation during tensile testing.

An estimate of the amount of deformation recovered by the WADCO specimen may be reached by comparing the difference in the H_v values at room temperature from the ascending and

descending temperature test results (Figure 12) to similar results for the PNL material where the specimen had received 20% cold work prior to testing. The descending temperature hot microhardness test results for this specimen are identical to the continuous curve in Figure 12. The ΔH_v for the PNL material was found to be ~ 106 whereas the value for the WADCO specimen is ~ 27 . This suggests that $\sim 5\%$ deformation was recovered by the WADCO specimen during the hot microhardness test.

Based on the above observations, an examination of the substructure of the WADCO specimen is planned to ascertain the nature of the recovery as the result of hot microhardness testing and to evaluate the deformation due to the microhardness indentations.

Cavitation (304 SS)

A subprogram is presently underway to quantitatively study the phenomenon of grain boundary microcracking which occurs during the elevated temperature plastic deformation of stainless steels. This work includes optical metallographic studies of microcracks formed during slow tensile and creep-rupture deformation of type 304 stainless steel and immersion density studies of bulk density changes which result from the microcrack formation. A condensed review of the detailed literature survey performed prior to the initiation of this subprogram is presented in the following

section.

Literature Review - Several excellent reviews are available concerning intergranular cavitation at elevated temperatures in polycrystalline metals and alloys. (9, 10, 11) These reviews show that failure is transgranular for metals deformed at high strain rates and low temperatures but intergranular at low strain rates and high temperatures. A noticeable loss of ductility is characteristic of the transition from transgranular to intergranular failure which has been attributed to the formation and propagation of grain boundary cavities. The variables which affect the mode of cracking include stress, strain, grain size, and composition, in addition to strain rate and temperature. Impurity levels appear to enhance grain boundary crack formation by suppressing boundary migration, phase changes, and recrystallization. Pure metals are observed to usually fail in a transgranular manner.

Two types of cavities are usually observed along grain boundaries during creep and slow tensile deformation at elevated temperatures. Generally, wedge-type cracks are formed at relatively lower temperatures and higher stresses while a more rounded cavity forms at relatively higher temperatures and lower stresses.

The nucleation mechanisms of both types of cavitation

are poorly understood. Stress concentration due to grain boundary sliding⁽¹²⁾ appears to be responsible for the formation of cracks near grain boundary triple junctions (triple point cracks) or along boundary edges (secondary cracks). A more rounded cavity shape is believed to occur at grain boundary edges, jogs, and inclusions (nonwetting particles).

The growth of these cavities then occurs by grain boundary sliding, dislocation motion, vacancy migration, or a combination of these processes. From the available evidence, it appears that a process involving vacancy condensation may be more dominant than mechanical processes in the growth of round-type cavities.

Failure of the material then occurs by the growth, linking, or propagation of the cavities. The failure process is, surprisingly, poorly understood in spite of its importance to the mechanical behavior of materials at elevated temperatures.

For the particular case of wedge-type cracks, there exists ample experimental evidence that the cracks are often initiated and terminated at triple junctions.⁽¹³⁻¹⁷⁾ In this case, the crack lengths are conveniently expressed in terms of the number of grain boundary facets which the cracks occupy. Lindborg⁽¹³⁾ has adopted this method to express the crack length distribution and crack

number density in a 20 Cr - 35 Ni austenitic stainless steel tested in slow tension at 700°C. He found that the cracks grew in a step-wise manner with rapid propagation between triple junctions, but were arrested at triple junctions. Most cracks were found to have a length of one boundary facet (65%) while the number of cracks occupying more than one facet decreased as crack length increased. The crack length distribution from the onset of testing to fracture was found to remain relatively constant regardless of strain rate and test time resulting in an average crack length of approximately 1.5 to 1.6 boundary facets. The number of cracks increased exponentially with strain or time resulting in an average growth rate which is constant as a function of strain. The percentage of cracked boundaries near the fracture zone, ~ 6%, was almost independent of strain rate and agrees with the value predicted from a statistical model for intercrystalline wedge crack linking.⁽¹⁸⁾

An extension of this study⁽¹⁴⁾ indicated that failure occurs by crack growth for large grain sizes and by linking for small grain sizes for the range of grain sizes from 20 to 275 μm . A companion study⁽¹⁶⁾ on the same 20 Cr - 35 Ni austenitic stainless steel at 700°C suggests that the Griffith-Orowan mechanism is controlling during the final fracture process. Similar results were obtained by

Taplin^(17, 19) for α -brass and copper.

Williams⁽²⁰⁾ has reported that the growth of triple point cracks occurs in two stages in an Al-20% Zn alloy. The crack first grows in a slow, stable manner until a critical wedge height is reached where upon the growth accelerates in an unstable manner until the next triple junction is reached. The critical wedge height was found to be an inverse function of the applied stress. A similar two stage crack growth process has been recently proposed by Evans⁽²¹⁾ based on energy balance arguments.

The effect of strain rate, grain size, and stress have been investigated with respect to the angular incidence of cavities with respect to the stress axis.^(20, 22-30) At higher strain rates, the maximum frequency of distribution is at approximately 45° to the stress axis while at lower strain rates the maximum is near 90° . This suggests that mechanical shearing processes are operative at higher strain rates as opposed to vacancy condensation mechanisms at lower strain rates. It is also reported that the angular distribution is independent of strain rate.⁽²³⁾ Grain boundary sliding is reported to be more significant for small grain sizes from the fact that the most cavities occur at 45° to the stress axis as compared to a maximum frequency near 90° for large grain sizes.^(22, 23) The angle between the stress axis and crack direction is also

reported⁽²⁰⁾ to enhance crack growth rate. At angles near 45° the growth rate is at least 10 times faster than when compared to the rate at angles near 90°.

Measurements of the bulk density changes which result from the formation of cracks during the deformation of metals at elevated temperatures have been reported.^(26, 31-37)

This technique has proved to be a very accurate method of detecting the total amount of cracking present, but cannot differentiate between crack nucleation and growth processes. The materials investigated by this technique include copper,^(31, 34) aluminum,⁽³⁵⁾ nickel and nickel alloys,^(32, 37) magnesium and magnesium alloys,^(36, 37) and steels.^(33, 35) The fractional density change has been found to be proportional to the period of creep raised to the third power for a 20% Cr - 25% Ni stainless steel⁽³³⁾ and magnesium.⁽³⁶⁾ For the steel, it was found that the exponent is not dependent on grain size and irradiation exposure. Other investigators have reported that the fractional density change is linearly related to time⁽³⁵⁾ or to strain throughout the latter stages of creep.⁽³²⁾ The growth of helium bubbles in irradiated steels has been reported to enhance the rate of crack propagation which leads to a loss of ductility.⁽³³⁾

Studies of crack morphology by shadow graphic and fractographic electron microscopy techniques have been

reported. (28, 38-45) These techniques have the advantage of being able to reveal the true cavity shape and to provide information supplementary to that observable by optical microscopy.

Experimental Results - Initial microcrack results were presented in a previous report⁽¹⁾ for type 304 stainless steel specimens (ORNL Reference Heat No. 9T2796) tensile tested to failure at 650°C. The crack number density as a function of distance toward the fracture zone was found to be strongly dependent on strain rate (see Figure 10 of reference 1).

The crack length distribution as a function of crack length (number of grain boundary facets) for these specimens is shown in Figure 13. It is evident that the predominant crack length is one grain boundary facet. The fractions of single and double cracks, v_1/v and v_2/v_1 where v is the total number of cracks, are of the same order of magnitude as results for a material with approximately the same grain size⁽¹⁴⁾ but do not remain constant as a function of strain rate. The present results support the observation⁽¹⁶⁾ that triple points are a significant barrier to intergranular crack propagation since most cracks were observed to either originate or terminate in the vicinity of triple junctions.

The relative order of the rates of crack number density

increase toward the fracture zone for the three strain rates (Figure 10 of reference 1) may be understood by noting that the intermediate strain rate specimen ($\dot{\epsilon} = 8.33 \times 10^{-5} \text{ sec}^{-1}$) exhibits the largest fraction of cracks whose length is two grain boundary facets, v_2/v (Figure 13). This fraction is smallest for the lowest strain rate specimen ($\dot{\epsilon} = 8.33 \times 10^{-6} \text{ sec}^{-1}$) and intermediate for the highest strain rate specimen ($\dot{\epsilon} = 8.33 \times 10^{-4} \text{ sec}^{-1}$). The tensile test results for elongation and reduction of area (see Table I of reference 1) also follow this same order, only inversely.

Figure 14 shows the measured crack percentage distribution as a function of angle from the specimen tensile axis for the 304 stainless steel specimens. The observed shift toward a higher angle with decreasing strain rate is as expected. A smaller, second maximum at about 50° , in addition to the primary maximum fraction at 80° , is clearly evident for the intermediate strain rate specimen.

Figures 15, 16, and 17 show the appearance of the microcracks in the 304 stainless steel tensile specimens as a function of distance toward the fracture surface. The magnification is 80X. The specimen tested at the highest strain rate ($\dot{\epsilon} = 8.33 \times 10^{-4} \text{ sec}^{-1}$) contains few cracks at a location well removed from the fracture surface (Figure 15a). The relative amount of cracking is seen to

increase for the locations nearer to the fracture surface (Figures 15b and c). A somewhat rounded crack appearance is also evident for this specimen.

Figure 16 shows that the cracks are rather narrow and appear to be predominantly wedge-type (sharply pointed) for the specimen tested the intermediate strain rate ($\dot{\epsilon} = 8.33 \times 10^{-5} \text{ sec}^{-1}$). Note that most cracks occupy all or a portion of one grain boundary facet. The total number of cracks has clearly increased when compared to Figure 15.

The specimen tested at the lowest strain rate ($\dot{\epsilon} = 8.33 \times 10^{-6} \text{ sec}^{-1}$) exhibits the highest crack density at all points except near the fracture zone. This is evident in Figure 17c where the cracks are significantly larger in terms of width, but not in length, than those in Figure 16c. The relative number of the large cracks is, however, less in Figure 17c than in Figure 16c. Figures 17a and b appear to be quite similar to Figures 16a and b, as expected, except that more small cracks are evident in Figures 16a and b.

Deformation markings due to slip may also be within the grains adjacent to cracks in all three figures as well as for the cover micrograph of this report. The presence of extensive twinning is clearly illustrated by these figures.

A typical example of a well-developed wedge-type crack in the 304 stainless steel specimens is shown in Figure 18a at 200X and illustrated schematically in Figure 18b. This

crack is located approximately 2.2 mm from the fracture surface of the specimen tested at a strain rate of $8.33 \times 10^{-6} \text{ sec}^{-1}$. The orientation of the central portion of the crack is approximately 80° from the stress axis. This crack extends almost three grain boundary facets in length. Evidence of the barrier to crack extension provided by the grain boundary triple points may be seen at those locations where the crack has changed direction to follow the grain boundaries during growth. The portion of the crack near the top of Figure 18a appears to have been arrested by the triple point present, while the clearly wedge shaped portion toward the bottom of this figure may still be capable of further growth. The deformation of the grains adjacent to crack is also evident. For comparison, Figure 18c shows at 200X a wedge-type crack, one grain boundary facet in length, developed approximately 5.5 mm from the fracture zone in this same specimen.

PLANS FOR FUTURE WORK

Transmission electron microscopy work on creep-rupture specimens from the second heat of 316 stainless steel will be continued.

Microcrack studies of 304 stainless steel tested in the tensile and creep-rupture modes will continue. Data reduction and analysis of the tensile specimen results will be completed.

An investigation of the substructure which results from hot microhardness testing is planned in order to develop a quantitative understanding of the correlation between hot microhardness and hot tensile or creep data for 316 stainless steel.

A subprogram to study the substructure of 316 stainless steel tubular creep specimens will be developed.

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TABLE I
TENSILE AND SUBSTRUCTURE PROPERTIES OF 316 STAINLESS STEEL, HEAT NO. 332990

Specimen Number	Temperature T(°C)	Strain Rate $\dot{\epsilon}$ (sec ⁻¹)	True Stress(ksi)		True Strain at Fracture	Strain Hardening Factor m	Strength ^a Coefficient k(ksi)	Average Subgrain or Dislocation Cell Intercept d(10 ⁻³ mm)	Mobile Dislocation Density ρ_m (cm ⁻²)	Average Subgrain Misorientation (degrees)
			Ultimate	Maximum						
H07189	21	2.7x10 ⁻⁵	80.41	123.85	0.472	0.391	170.91	0.19(15) ^{b,c}	n.m. ^d	n.m.
H07190	149	2.7x10 ⁻⁵	70.49	95.82	0.347	0.286	133.21	0.28(19)	n.m.	n.m.
H07191	204	2.7x10 ⁻⁵	71.07	96.55	0.346	0.286	134.18	0.29(19)	n.m.	n.m.
H07192	260	2.7x10 ⁻⁵	74.59	101.65	0.348	0.293	141.25	0.30(10)	n.m.	n.m.
H07143	316	2.7x10 ⁻⁵	69.10	97.00	0.372	0.336	138.44	0.32(16)	n.m.	n.m.
H07193	371	2.7x10 ⁻⁵	71.56	103.49	0.400	0.346	144.81	0.33(32)	n.m.	n.m.
H07134	427	2.7x10 ⁻⁵	72.11	100.58	0.362	0.320	141.06	0.37(18)	n.m.	n.m.
H07135	538	2.7x10 ⁻⁵	67.24	94.72	0.368	0.335	134.34	0.39(13)	n.m.	n.m.
H07186	593	2.7x10 ⁻⁵	54.56	74.40	0.355	0.265	100.85	0.40 (6)	n.m.	n.m.
H07138	650	2.7x10 ⁻⁵	44.96	57.45	0.322	0.182	73.30	0.72(12)	n.m.	n.m.
H07188	704	2.7x10 ⁻⁵	30.16	37.36	0.466	0.120	43.70	1.16(16)	2.75x10 ⁹ (9)	n.m.
H07141	760	2.7x10 ⁻⁵	24.67	29.06	0.483	0.059	30.85	1.55(17)	1.91x10 ⁹ (15)	0.63(6)
H07194	816	2.7x10 ⁻⁵	20.41	22.94	0.507	0.023	22.73	1.75(16)	1.41x10 ⁹ (9)	0.69(4)

a. Conversion factor for ksi to kg/mm²: 1 ksi = 0.702 kg/mm².

b. The values tabulated for temperatures of 21°C(70°F) to 593°C(1100°F) are for dislocation cells.

c. The number in parenthesis represents the number of plates used for that determination.

d. n.m. - not measured

A.

B.

FIGURE 1

TYPE 316 STAINLESS STEEL, HEAT No. 332990

TENSILE TEST SPECIMENS (AS-RECEIVED)

TEST TEMPERATURES (LEFT TO RIGHT):

A. 21°C, 149°C - 427°C

B. 538°C - 816°C

(MAGNIFICATION 1.25X)

FIGURE 2

SUBSTRUCTURE OF SHOULDER PORTION OF
SPECIMEN 7189. $\langle 111 \rangle$ SLIP TRACES ARE
CLEARLY EVIDENT.

FIGURE 3

A. SPECIMEN 7189
21°C (70°F)

B. SPECIMEN 7190
149°C (300°F)

C. SPECIMEN 7193
371°C (700°F)

D. SPECIMEN 7135
538°C (1000°F)

REPRESENTATIVE DEFORMATION SUBSTRUCTURE
OF TENSILE TESTED 316SS, HEAT No. 332990
(21°C TO 593°C).

FIGURE 4

A. SPECIMEN 7138
650°C (1200°F)

B. SPECIMEN 7188
704°C (1300°F)

C. SPECIMEN 7141
760°C (1400°F)

D. SPECIMEN 7194
816°C (1500°F)

REPRESENTATIVE DEFORMATION SUBSTRUCTURE
OF TENSILE TESTED 316SS, HEAT No. 332990
(650°C TO 816°C).

FIGURE 5

A. SPECIMEN 7138

B. SPECIMEN 7188

REPRESENTATIVE DISLOCATION CONFIGURATIONS
IN TENSILE TESTED 316SS, HEAT No. 332990
(650°C TO 816°C).

FIGURE 5 (CON'T.)

C. SPECIMEN 7141

D. SPECIMEN 7194

FIGURE 6

MOBILE DISLOCATION DENSITY RESULTING FROM TENSILE AND CREEP DEFORMATION VERSUS MAXIMUM TRUE TENSILE STRESS AND INITIAL APPLIED CREEP STRESS.

FIGURE 7

RELATION BETWEEN AVERAGE SUBGRAIN OR DISLOCATION CELL INTERCEPT AND MAXIMUM TRUE TENSILE STRESS OR INITIAL APPLIED CREEP STRESS.

FIGURE 8

RELATION BETWEEN MOBILE DISLOCATION DENSITY AND
AVERAGE SUBGRAIN INTERCEPT.

FIGURE 9

AVERAGE SUBGRAIN OR DISLOCATION CELL
INTERCEPT AS A FUNCTION OF TENSILE TEST
TEMPERATURE.

FIGURE 10

RELATION BETWEEN STRAIN HARDENING EXPONENT AND AVERAGE SUBGRAIN OR DISLOCATION CELL INTERCEPT.

FIGURE 11

RELATION BETWEEN STRENGTH COEFFICIENT AND AVERAGE SUBGRAIN OR DISLOCATION CELL INTERCEPT.

FIGURE 12

HOT MICROHARDNESS OF 316 SS TENSILE TEST SPECIMEN (7189) SHOULDER AS A FUNCTION OF TEMPERATURE.

FIGURE 13

MICROCRACK LENGTH DISTRIBUTION AS A
FUNCTION OF CRACK LENGTH FOR 304SS
AT 650°C.

FIGURE 14

MEASURED MICROCRACK DISTRIBUTION AS
A FUNCTION OF ANGLE FROM THE SPECIMEN
AXIS FOR 304SS AT 650°C.

FIGURE 15

REPRESENTATIVE MICROGRAPHS OF CRACK
FORMATION FOR 304SS TENSILE TESTED
AT 650°C, $\dot{\epsilon} = 8.33 \times 10^{-4} \text{ SEC}^{-1}$.

A. $\sim 15\text{MM}$ FROM FRACTURE ZONE. B. $\sim 9\text{MM}$ FROM FRACTURE ZONE. C. $\sim 2\text{MM}$ FROM FRACTURE ZONE.

FIGURE 16

REPRESENTATIVE MICROGRAPHS OF CRACK
FORMATION FOR 304SS TENSILE TESTED
AT 650°C, $\dot{\epsilon} = 8.33 \times 10^{-5} \text{ sec}^{-1}$.

A. ~15MM FROM FRACTURE ZONE. B. ~ 9MM FROM FRACTURE ZONE. C. ~ 2MM FROM FRACTURE ZONE.

FIGURE 17

REPRESENTATIVE MICROGRAPHS OF CRACK
FORMATION FOR 304SS TENSILE TESTED
AT 650°C, $\dot{\epsilon} = 8.33 \times 10^{-6} \text{ SEC}^{-1}$.

A. $\sim 15\text{MM}$ FROM FRACTURE ZONE. B. $\sim 9\text{MM}$ FROM FRACTURE ZONE. C. $\sim 2\text{MM}$ FROM FRACTURE ZONE.

FIGURE 18

(A)

(B)

(C)

EXAMPLE OF WEDGE-TYPE MICROCRACK
IN 304SS AT 650°C.

(A) WEDGE-TYPE MICROCRACK ALMOST THREE
GRAIN BOUNDARY FACETS IN LENGTH;
 $\dot{\epsilon} = 8.33 \times 10^{-6} \text{ sec}^{-1}$; CRACK LOCATED
 $\sim 2.2 \text{ mm}$ FROM FRACTURE ZONE (200X).

(B) SCHEMATIC ILLUSTRATION OF (A)

(C) WEDGE-TYPE MICROCRACK IN SAME
SPECIMEN AS (A) LOCATED $\sim 5.5 \text{ mm}$
FROM FRACTURE ZONE (200X).

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