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Thermal Aging Behavior of ERNiCr-3 Alloy (Weld and Base Metal)

R. L. Klueh
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Printed in the United States of America. Available from
National Technical Information Service
U.S. Department of Commerce
5285 Port Royal Road, Springfield, Virginia 22161
NTIS price codes—Printed Copy: A03 Microfiche A01

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GCR Structural Materials Studies (FTP/A 01332)

THERMAL AGING BEHAVIOR OF ERNiCr-3 ALLOY (WELD AND BASE METAL)

R. L. Klueh and J. F. King

Date Published: August 1981

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THERMAL AGING BEHAVIOR OF ERNiCr-3 ALLOY (WELD AND BASE METAL)

R. L. Klueh and J. F. King

ABSTRACT

The nickel-base filler metal alloy ERNiCr-3, containing nominally 67% Ni, 20% Cr, 3% Fe, 3% Mn, and 2.5% Nb, is used widely to make welds for elevated-temperature service. To determine the effect of elevated temperature on tensile and creep-rupture properties of ERNiCr-3, weld metal specimens were thermally aged to 10,000 h at 510°C, to 15,000 h at 566°C, and to 1000 h at 677°C. Wrought ERNiCr-3 was also aged at 566 and 677°C.

The 0.2% yield strength of the ERNiCr-3 weld metal increased with thermal aging time at 510 and 566°C. The ultimate tensile strength also increased continuously with aging time at 566°C, whereas at 510°C it went through a maximum (the strength of the material aged 10,000 h was less than was that aged 5000 h). At 677°C we observed little change in the room-temperature yield strength or in the ultimate tensile strength with aging time. The ductility at 510 and 566°C decreased continuously with aging time but changed little at 677°C. The stress-rupture lives at the aging temperature for material aged at both 510 and 566°C displayed maxima. The total elongation and reduction of area during creep decreased with aging time.

Changes in strength and ductility were attributed to short-range order formation (which gives rise to an increase in strength) and to development of a precipitate that covered the grain boundaries (which can affect the ductility). Because short-range order decreases with increasing temperature, less short-range order is formed at 677°C; thus, we observed little effect of aging on properties at 677°C.

INTRODUCTION

The weld filler metal alloy designated by the American Welding Society as ERNiCr-3 is a high-temperature alloy with a nominal composition of 67% Ni, 20% Cr, 3% Mn, 3% Fe, 2.5% Nb, 0.3% Ti, and lesser amounts of several other elements. It is used conventionally to make

austenitic-ferritic dissimilar-alloy joints and to join high-nickel alloys. The weld filler alloy is commonly known as Inconel 82,* and the wrought alloy is known as Inconel 606.*

We previously reported on the elevated-temperature mechanical properties of this alloy.[1,2] In those studies we obtained evidence that short-range order forms in the alloy from 454 to 566°C (ref. 3). For the long-time creep tests between 566 and 732°C we found evidence of precipitation.[2] Because each of these processes can lead to changes in mechanical properties, we aged a series of weld metal specimens at 510 and 566°C to determine the effect of thermal aging on the room-temperature tensile and creep-rupture properties. A few specimens were also aged at 677°C for room-temperature tensile tests, as was wrought ERNiCr-3 aged at 566 and 677°C.

EXPERIMENTAL

For the weld metal studies, mechanical property specimens were machined from weld deposits of ERNiCr-3 made in 19-mm-thick (0.75-in.) plates of 2 1/4 Cr-1 Mo steel. The steel was machined with a 30° included-angle V-groove and spaced approximately 32-mm (1.25-in.) at the root; a type 304 stainless steel backing strip was used. The welding was performed by the automatic gas tungsten arc (GTA) process with hot-wire filler additions. Approximately 16 weld passes were required to fill the joint. The nominal welding conditions were 398 A and 15 V, with a travel speed of about 2.5 mm/s. Table 1 gives the chemical composition of the deposited weld metal, the AWS and ASTM composition specifications for ERNiCr-3, and the chemical composition of the wrought ERNiCr-3 (a rod). All welds were radiographed, and no defects were detected.

Mechanical property tests were made on buttonhead specimens with a 3.18-mm-diam (0.125-in.) by 28.6-mm-long (1.125-in.) gage section. The specimens from the weldments were longitudinal and were all weld metal. Before aging, all specimens were postweld heat treated 1 h at 732°C. Specimens were aged in air up to 15,000 h at 510, 566, and 677°C.

---

*Trade name of Huntington Alloy.

Table 1. Chemical composition of ERNiCr-3 filler metal and wrought alloy

Element	Chemical composition, wt %		
	Weld metal	Wrought alloy	ASTM B.304 AWS A5.14
Ni	Balance	Balance	67.0 min
Cu	0.2	0.12	0.5 max
Mn	3.2	2.95	2.5–3.5
Fe	1.2	0.91	3.0 max
Si	0.2	0.21	0.5 max
C	0.03	0.05	0.1 max
S		0.009	0.015 max
Ti	0.3	0.41	0.75 max
Nb	2.4	2.71	2.0–3.0
Cr	20.0	19.25	18–22

Room-temperature tensile tests were made on an Instron machine with a crosshead speed of 8.5 μm/s, which gave a nominal strain rate of 3×10^{-6}/s. Creep-rupture tests were made in air in constant-load lever-arm creep frames. The specimens were heated in a resistanace tube furnace in which the temperature was monitored and controlled by three Chromel versus Alumel thermocouples attached to the specimen gage length. Temperature was controlled to ±1°C with a variation along the gage length of less than ±2°C. Elongations were determined with mechanical extensometers attached to the specimen grips; the specimen elongation was read periodically on a dial gage.

RESULTS

Tensile and Creep Properties

Room-temperature tensile tests were made on the ERNiCr-3 weld metal after aging at 510, 566, and 677°C (Table 2). A limited number of tests were also made on ERNiCr-3 wrought alloy to determine if thermal aging has the same effect on the two product forms (Table 3).

Table 2. Room-temperature tensile properties of aged ERNiCr-3 weld metal (Inconel 82)

Aging time (h)	Strength, MPa			Elongation, %		Reduction of area (%)
	Yield	Ultimate	Fracture	Uniform	Total	
			Aged at 510°C			
0	326	615	499	36.9	43.3	55.5
200	339	618	546	36.5	40.7	55.2
1,000	346	634	511	40.4	47.3	53.2
2,500	395	656	536	34.5	37.1	50.8
5,000	404	688	533	30.0	33.6	51.0
10,000	416	666	600	26.9	28.5	30.2
			Aged at 566°C			
0	326	615	499	36.9	43.3	55.5
200	348	625	520	38.0	42.2	54.5
1,000	389	681	578	35.2	38.5	45.2
2,500	452	720	611	26.6	30.1	53.2
5,000	462	745	667	31.5	35.6	43.9
15,000	523	823	743	26.3	29.2	39.3
			Aged at 677°C			
0	326	615	499	36.9	43.3	55.5
200	346	625	480	35.6	41.8	56.2
1,000	341	656	578	36.6	41.7	52.0

Table 3. Room-temperature tensile properties of aged wrought ERNiCr-3 (Inconel 606)

Aging time (h)	Strength, MPa			Elongation, %		Reduction of area (%)
	Yield	Ultimate	Fracture	Uniform	Total	
			Aged at 566°C			
0	429	774	549	39.4	46.3	65.4
200	427	771	584	40.6	46.2	61.8
1000	472	806	627	37.1	43.6	57.1
2500	525	843	694	32.9	38.9	54.6
5000	578	910	742	31.0	36.5	52.2
			Aged at 677°C			
0	429	774	549	39.4	46.3	65.4
200	422	771	597	40.3	47.4	59.7
1000	411	764	653	41.3	48.0	54.4

The results clearly indicate that thermal aging increases the 0.2% yield strength (Fig. 1) and the ultimate tensile strength (Fig. 2) of the weld metal at 510 and 566°C and the wrought alloy at 566°C. Thermal aging at 677°C affected tensile behavior much less: essentially none on the wrought alloy and only very slightly on the ultimate tensile strength of the weld metal (Figs. 1 and 2). The tensile ductility (total elongation and reduction of area) decreased with thermal aging at 510 and 566°C, but changed little at 677°C (Fig. 3).

Creep-rupture tests on ERNiCr-3 weld metal specimens aged at 510 and 566°C (Table 4) were at 434 MPa and 510°C for specimens aged at 510°C and at 345 MPa at 566°C for specimens aged at 566°C (Figs. 4–6). At 510°C the rupture life decreased during the first 200 h of aging, increased for aging up to 5000 h, and decreased again. At 566°C the strength went through a maximum after 2500 h and leveled off [Fig. 4(*a*)]. The steady-state creep rate showed a similar behavior [Fig. 4(*b*)] (where the rupture life increased, the creep rate decreased and vice versa).

The total elongation and reduction of area values for the creep-rupture specimens at 510°C show an initial increase with aging time and a subsequent decrease (Fig. 5). At 566°C ductility generally decreases, although it increases slightly between 2500 and 15,000 h. However, even with the increase, the ductility does not reach the value levels of the as-received material. Qualitatively, the changes in ductility are similar to the changes in minimum creep rate with aging [Fig. 4(*b*)].

In previous creep studies on ERNiCr-3 weld metal,[2,4] we noted a peculiar behavior that we described as an "instant elongation" or a "strain burst." The phenomenon exhibited itself in the form of a jump in the creep curve. Strain bursts were also observed in this study. At 566°C, only the unaged specimen displayed a burst, which occurred after about 1 h of the test. All the tests at 510°C displayed one or more strain bursts at some time during the test. Our observations on strain bursts are summarized in Table 5; an example of a strain burst is shown in Fig. 6. For the tests after aging 0, 1000, 2500, and 5000 h, the bursts occurred between 25 and 50 h. In tests of the specimens aged 200 and 10,000 h, the bursts occurred early in the test. Two bursts occurred for the specimen

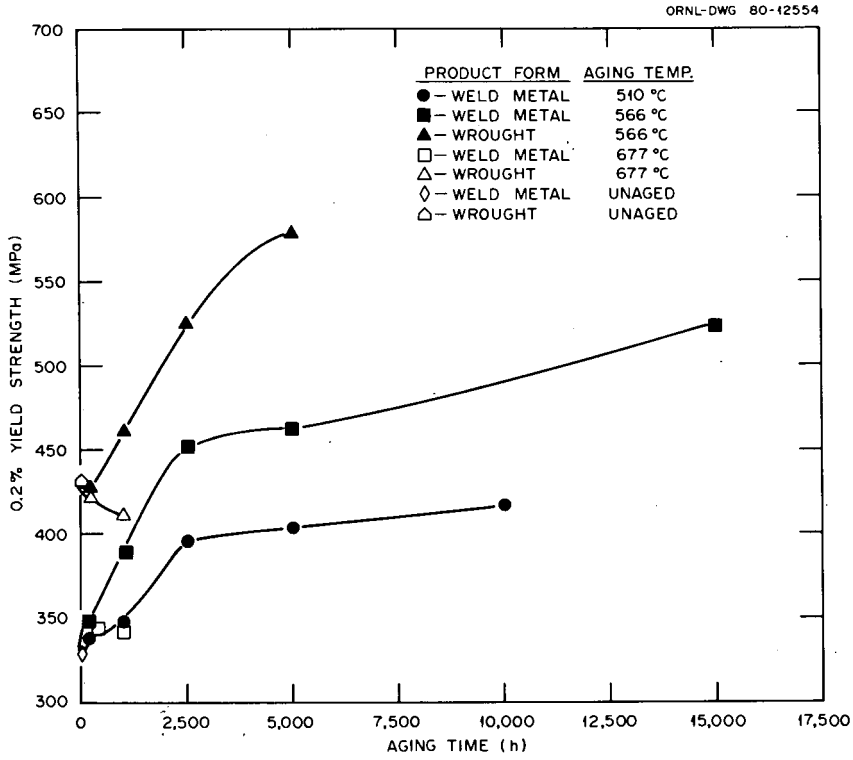

Fig. 1. Room-temperature 0.2% yield strength of ERNiCr-3 weld metal and wrought alloy as a function of aging time.

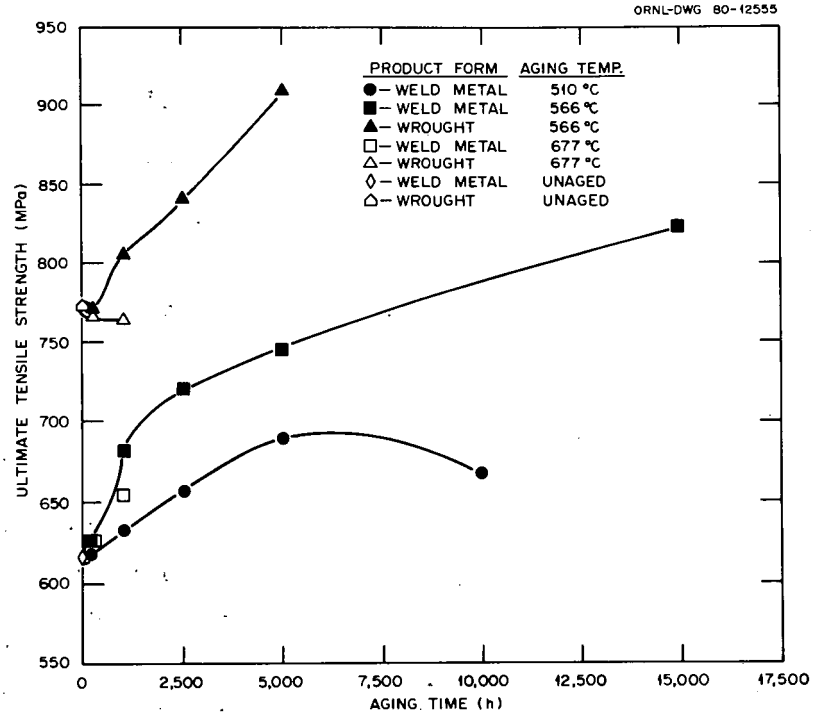

Fig. 2. Room-temperature ultimate tensile strength of ERNiCr-3 weld metal and wrought alloy as a function of aging time.

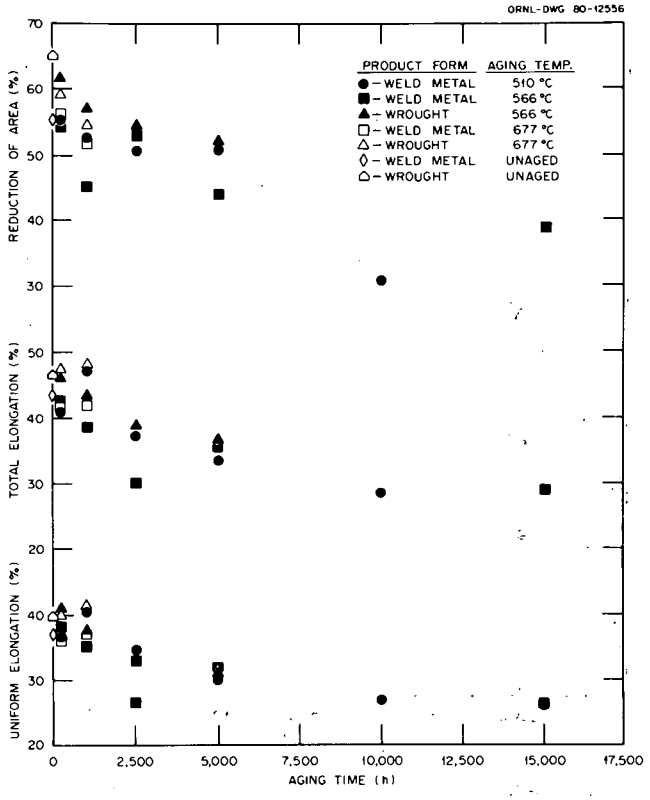

Fig. 3. Uniform elongation, total elongation, and reduction of area at room temperature as a function of aging time for ERNiCr-3 weld metal and wrought alloy as a function of aging time.

Table 4. Creep and rupture properties of aged ERNiCr-3 weld metal

Aging time (h)	Rupture life (h)	Elongation (%)	Reduction of area (%)	Minimum creep rate (%/h)	Loading strain (%)
Aged at 510°C, tested at 510°C and 434 MPa					
0	260.0	26.5	31.4	0.00347	18.0
200	72.9	37.2	40.0	0.0212	28.3
200	87.0	41.1	43.9	0.0222	28.7
1,000	123.9	38.0	40.9	0.0104	26.6
2,500	389.6	21.1	27.1	0.00210	16.5
5,000	1,294.8	19.5	23.2	0.000516	12.5
10,000	462.2	17.5	23.9	0.00186	10.8
Aged at 566°C, tested at 566°C and 345 MPa					
0	432.3	13.3	20.8	0.00107	8.0
200	851.9	10.7	16.4	0.00129	6.8
1,000	1,227.1	8.3	7.2	0.00122	4.7
2,500	3,052.3	3.6	2.9	0.000259	3.1
5,000	2,312.2	5.5	12.0	0.001031	0.8
15,000	1,907.6	3.2	7.2	0.000589	0.65

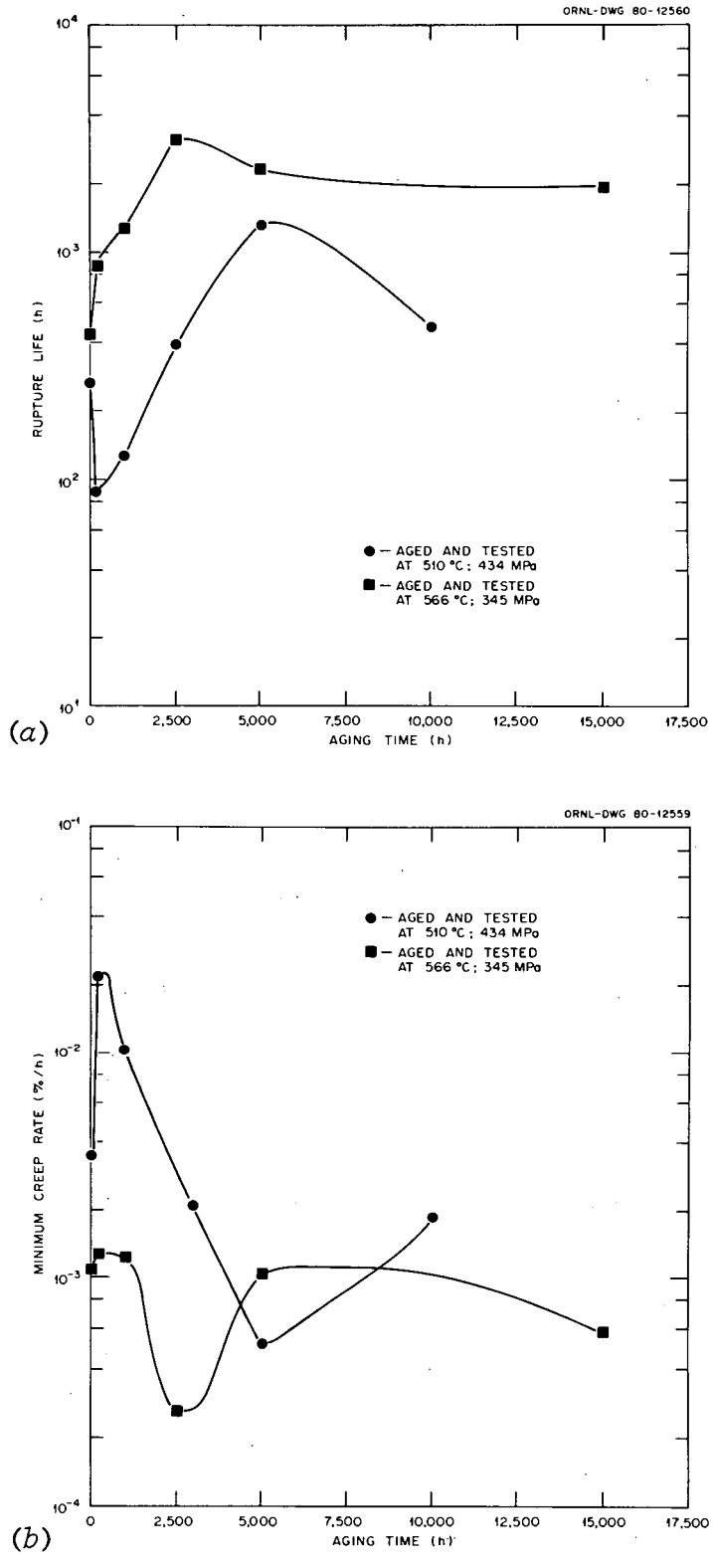

Fig. 4. Effect of aging time on (*a*) rupture life and (*b*) minimum creep rate for ERNiCr-3 weld metal aged at 510°C and tested at 434 MPa at 510°C and aged at 566°C and tested at 345 MPa at 566°C.

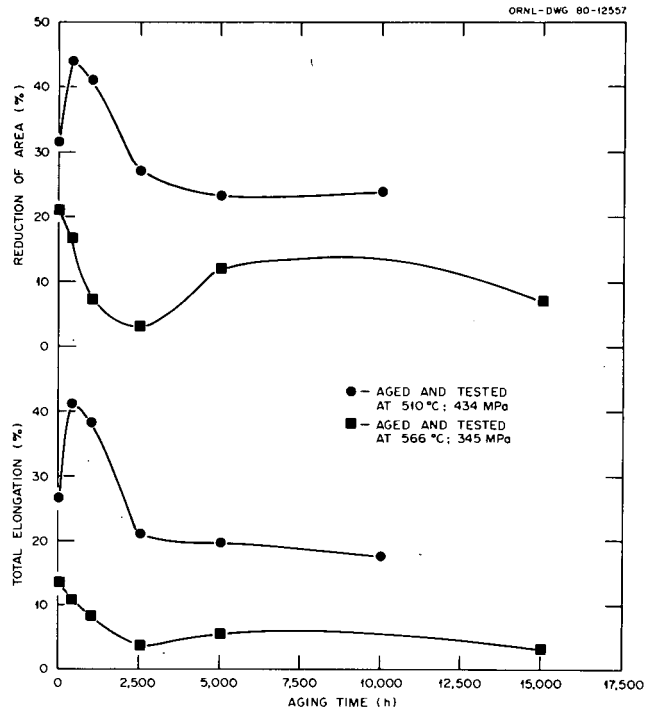

Fig. 5. Creep elongation and reduction of area as a function of aging time for ERNiCr-3 weld metal aged at 510°C and tested at 434 MPa at 510°C and aged at 566°C and tested at 345 MPa at 566°C.

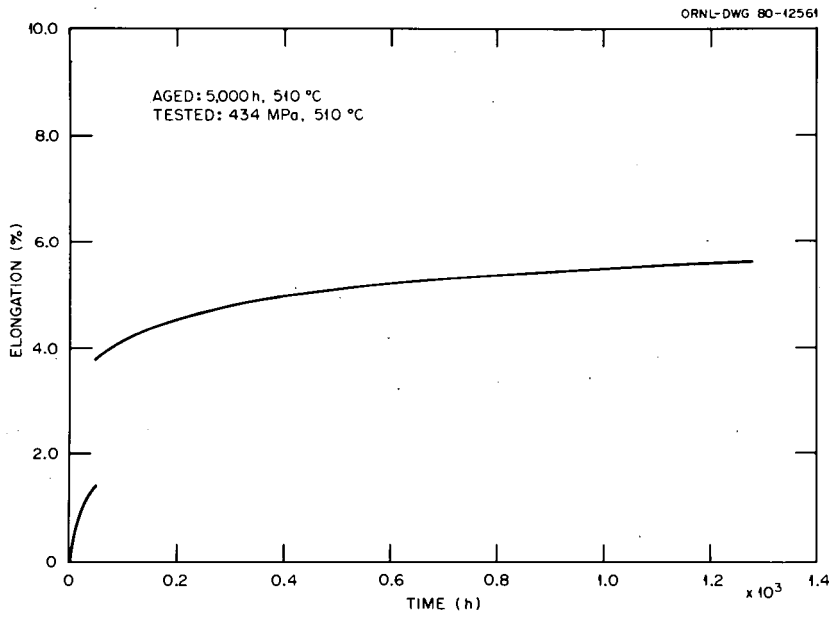

Fig. 6. Example of "instantaneous elongation" or a "strain burst" for the ERNiCr-3 weld metal aged 5000 h at 510°C and tested at 434 MPa at 510°C. Test ruptured at 1294.8 h.

Table 5. Observations on "strain bursts" in ERNiCr-3 weld metal

Aging time (h)	Approximate time of burst (h)	Magnitude of burst (%)
Aged and tested at 510°C		
0	35	2.6
200	<1,2	3,4
1,000	25	4.7
2,500	24	2.7
5,000	50	2.3
10,000	<1	2.3
Aged and tested at 566°C[a]		
0	1	1.5

[a]Only the unaged specimen tested at 566°C displayed a strain burst.

aged 200 h, one after less than 1 h and the second after about 2 h. The burst in the specimen aged 10,000 h occurred after less than 1 h of testing.

Microstructure

Pieces of the fractured tensile and creep specimens were polished and examined metallographically. The aged microstructure was studied by examining the buttonhead region of the tensile specimens. Microhardness measurements were made on the undeformed buttonhead regions to determine the effect of thermal aging on hardness.

Although consistent microhardness measurements are difficult to obtain, the results in this case were quite consistent. We found a continuous increase in the microhardness with aging time at 566°C. At 510°C the hardness fluctuated, although it generally increased with aging time (Fig. 7). The trends at 566°C especially are in general agreement with the tensile results (Figs. 1 and 2).

The microstructure of the as-welded ERNiCr-3 was a cellular-dendritic structure, which is fairly typical for a cast structure. After the 1-h

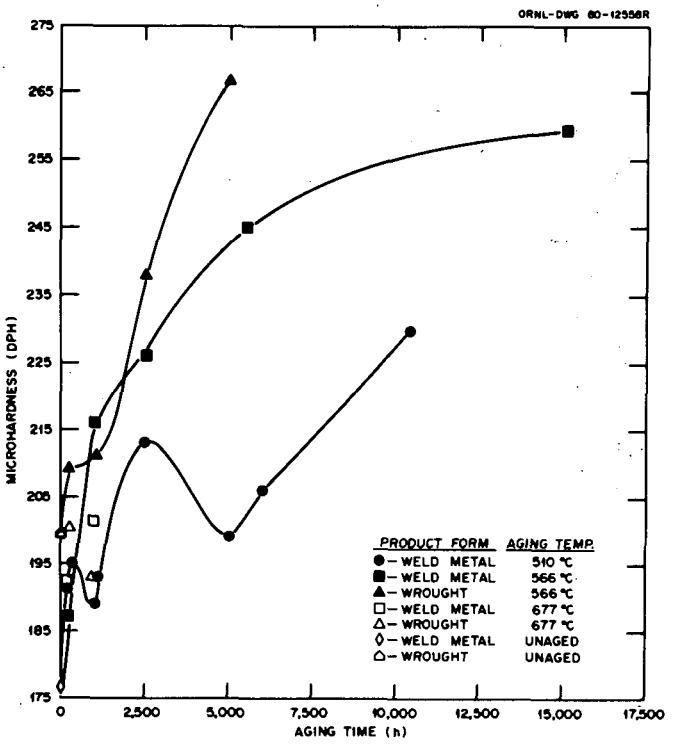

Fig. 7. Microhardness for ERNiCr-3 weld metal as a function of aging time at 510 and 566°C.

postweld heat treatment at 732°C, we observed an indication of grain structure (Fig. 8). However, the easily visible grains did not cover the entire cross section. Generally, the cellular-dendritic structure was present over much of the cross section, with no clear deliniation of grain boundaries.

After thermal aging we observed several interesting changes in microstructure. An inhomogenous grain structure developed when the weld metal was aged for 5000 h at 510°C [Fig. 9(*a*)] and 2500 h at 566°C [Fig. 9(*b*)]. After long aging times, this nonuniform grain structure was quite well developed. A very fine grain structure developed in the regions between adjacent passes (Fig. 10). At some distance from these interpass regions, the large elongated grains or the cellular-dendritic cast structure present in the unaged material were the only microstructural features that could be delineated.

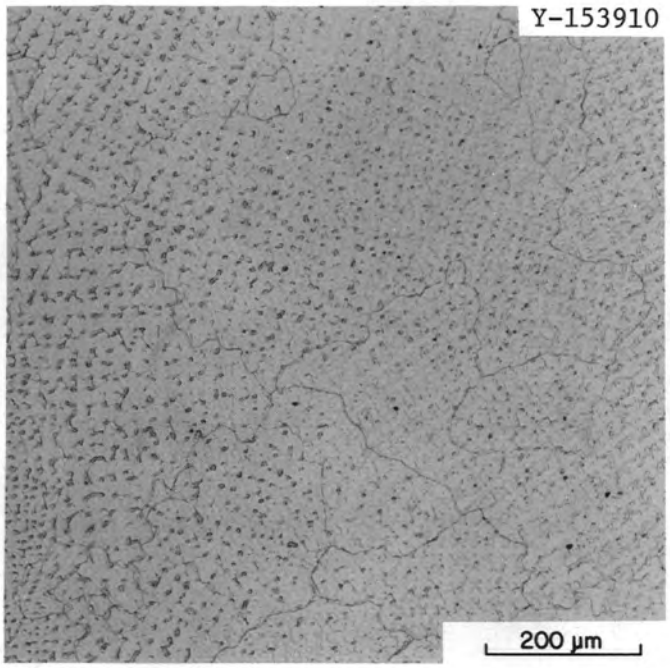

Fig. 8. Microstructure of ERNiCr-3 weld metal before aging.

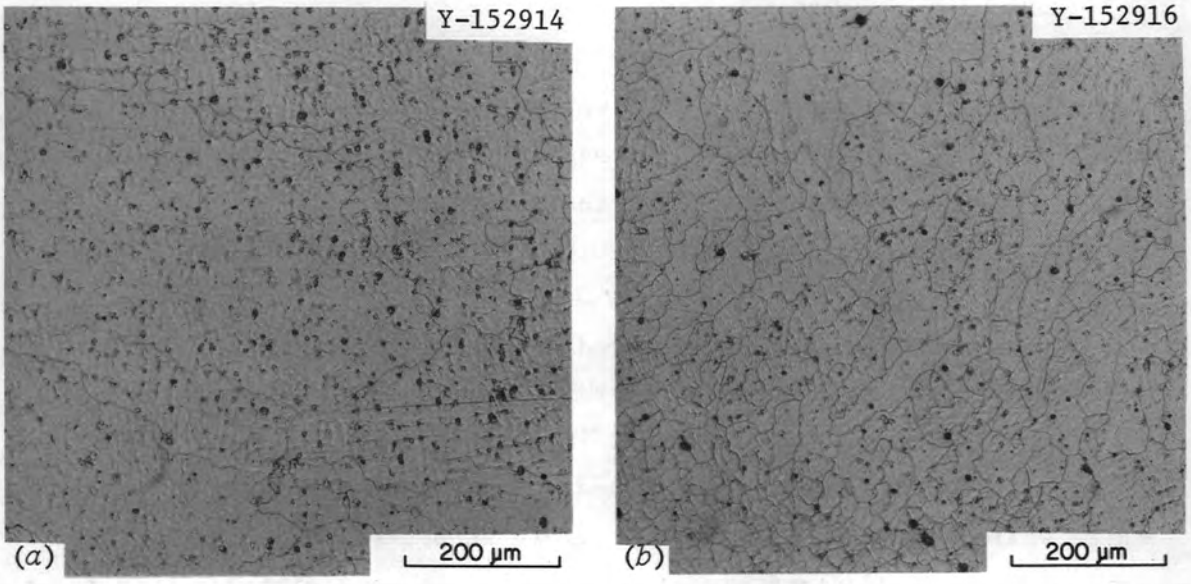

Fig. 9. Microstructure of ERNiCr-3 weld metal aged (*a*) 5000 h at 510°C and (*b*) 2500 h at 566°C.

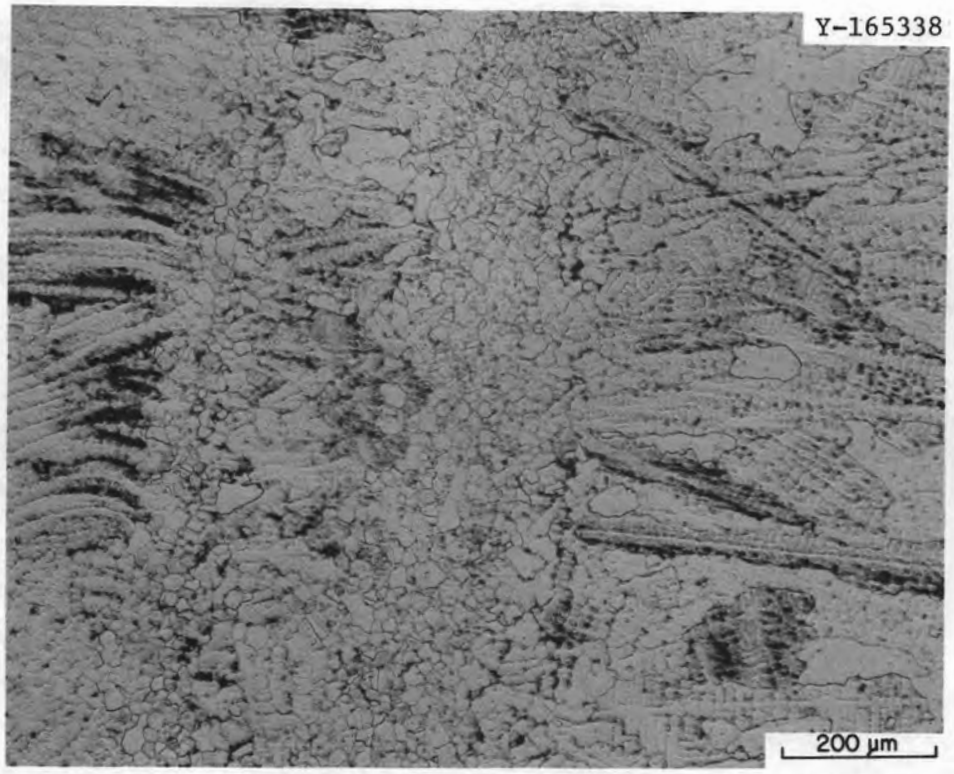

Fig. 10. Nonuniform microstructure developed in an ERNiCr-3 weld metal specimen aged 15,000 h at 566°C.

The grain structure seemed to develop in conjunction with the former cellular-dendritic structure. This was detected by examining the microstructure by differential interference contrast microscopy, the technique used to show the relationship of the cellular-dendritic structure and the grain boundaries in Fig. 11.

When the ERNiCr-3 weld metal was aged, an essentially continuous precipitate developed on the grain boundaries (Fig. 12), but there is little indication of a grain boundary precipitate after the postweld heat treatment (Fig. 13). One forms on aging, however, and the amount of precipitate present increases substantially with aging time at 510, 566, and 677°C. The rate of precipitate formation increases with aging temperature, as expected.

Development of the grain boundary precipitate within the weld structure appears to account for the observation on the development of the fine-grain structure with aging. That is, prior to precipitate formation

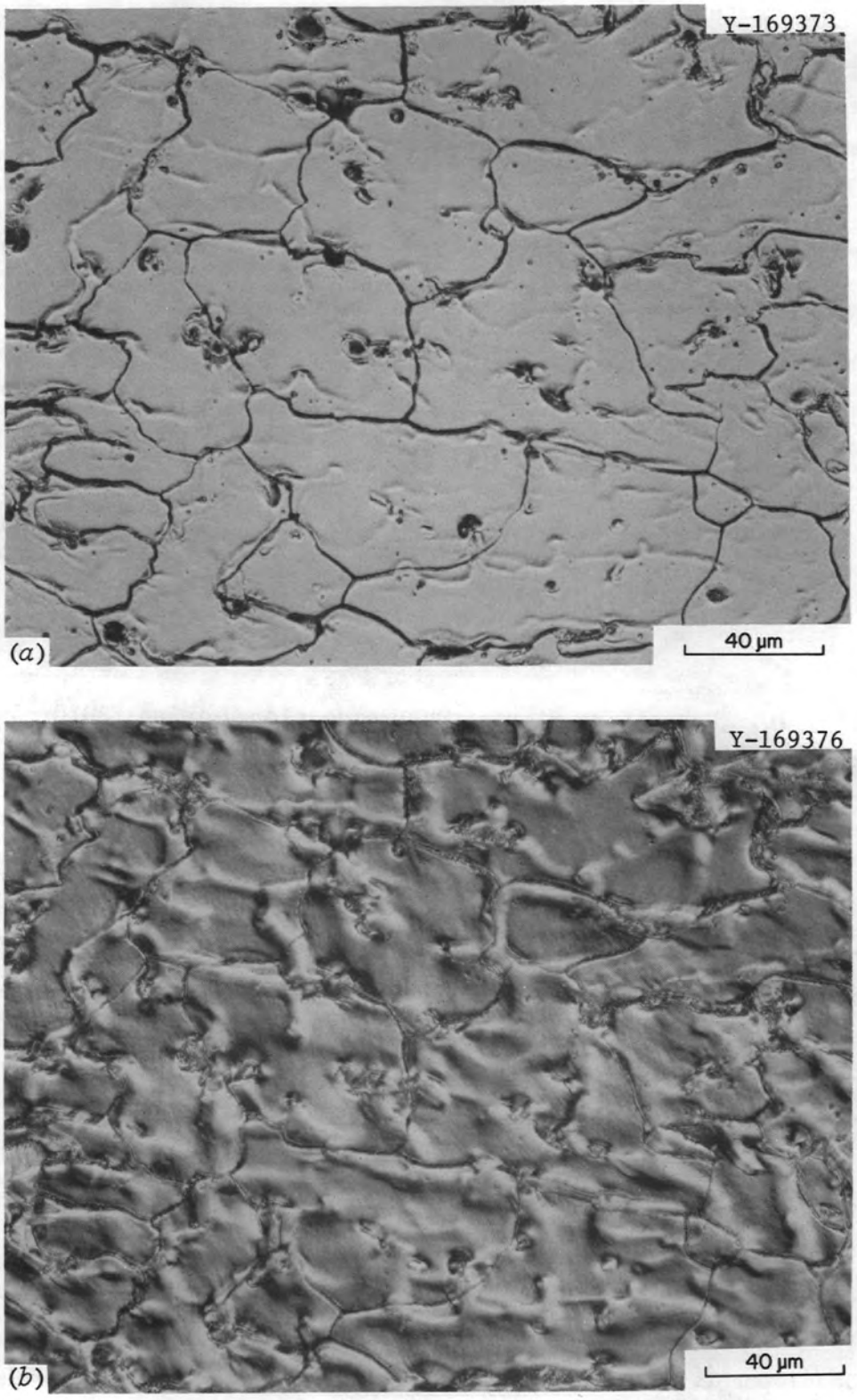

Fig. 11. Relationship between the cellular-dendritic and grain structures in a specimen aged 1000 h at 566°C. (*a*) Bright field and (*b*) differential interference contrast.

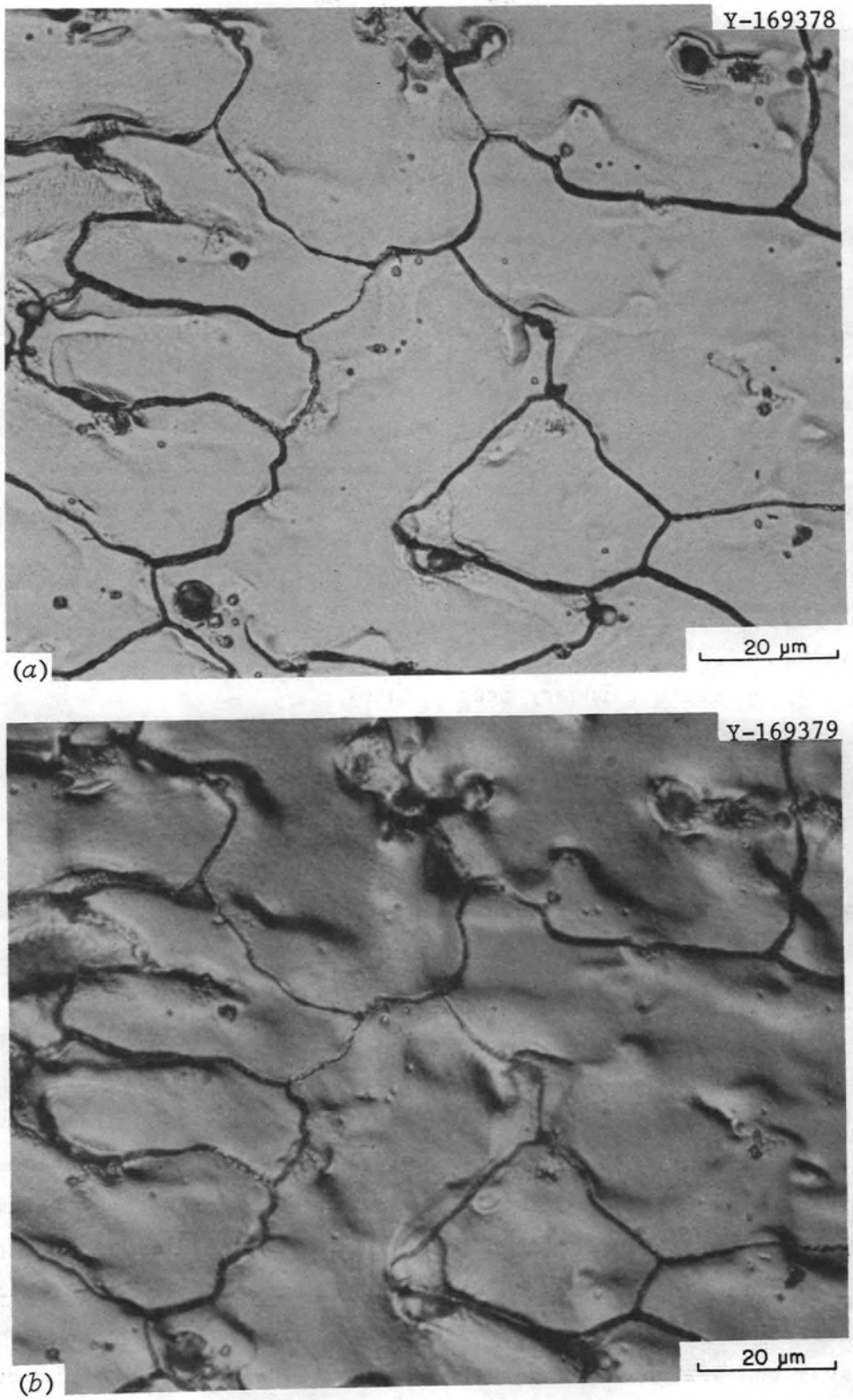

Fig. 12. Grain boundary precipitate that formed in ERNiCr-3 weld metal during thermal aging. Specimen was aged 1000 h at 566°C. (*a*) Bright field. (*b*) Differential interference contrast.

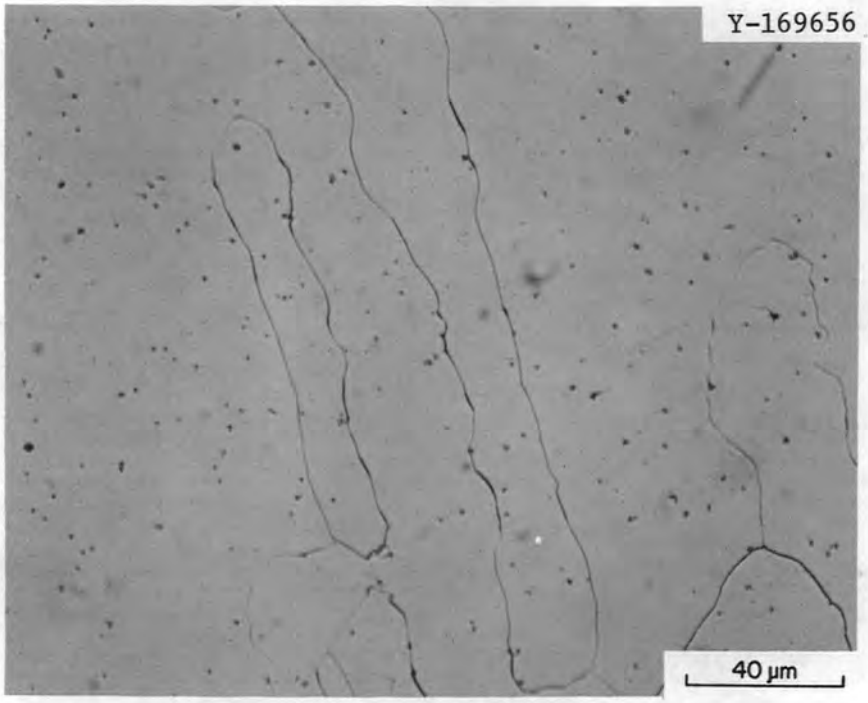

Fig. 13. Microstructure of ERNiCr-3 weld metal before aging showing essentially no grain boundary precipitate.

within the grains the grain structure may not be resolvable from the dendritic structure. This observation is supported to some extent by the microstructure shown in Fig. 14, where regions appear with only part of the grain structure visible. The other possibility is that the smaller grains somehow develop out of the cellular-dendritic weld structure during aging.

With time at the aging temperature, a precipitate also developed within the matrix (Fig. 15). This precipitate was undoubtedly associated with the prior cellular-dendritic substructure; that is, paralled rows of precipitate formed in regions where solute was rejected when the material solidified. Initially, a fine dark precipitate develops. With time, resolvable precipitate particles form from this fine diffuse precipitate (Fig. 15). The extent to which clearly defined precipitate particles were resolvable by optical microscopy depended on aging time and temperature. The higher the aging temperature (for a given aging time), the more quickly a precipitate could be resolved. After 1000 h at 677°C, some of the needlelike precipitates were observed that had previously been seen in specimens creep tested at 732°C (ref. 3).

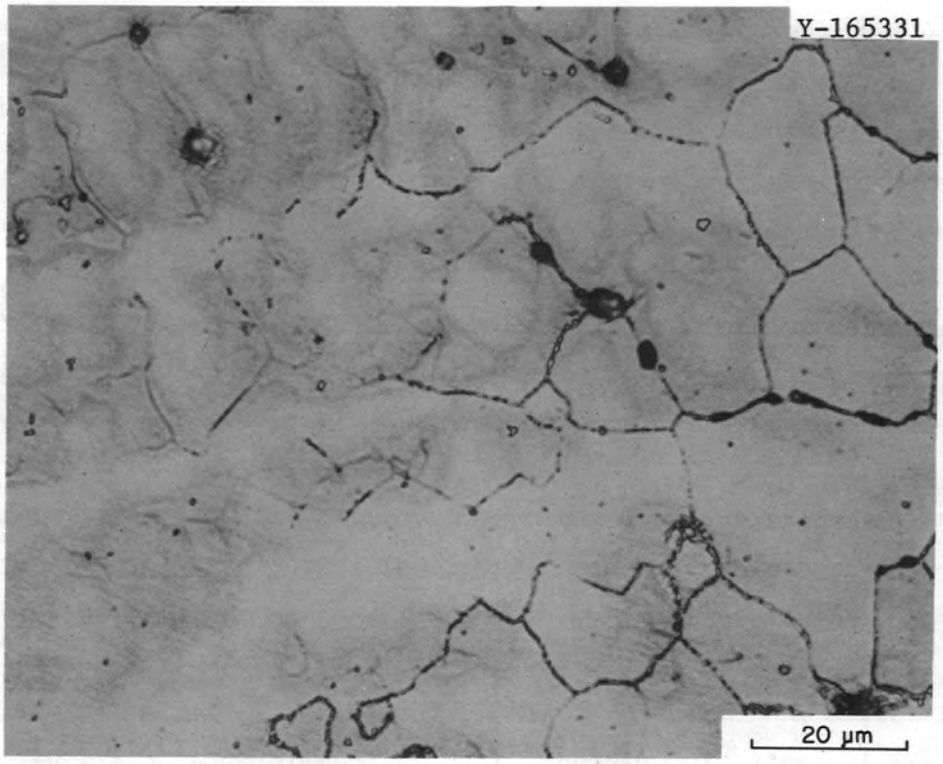

Fig. 14. Photomicrograph of ERNiCr-3 weld metal indicating that observation of grain structure depends on grain boundary precipitate.

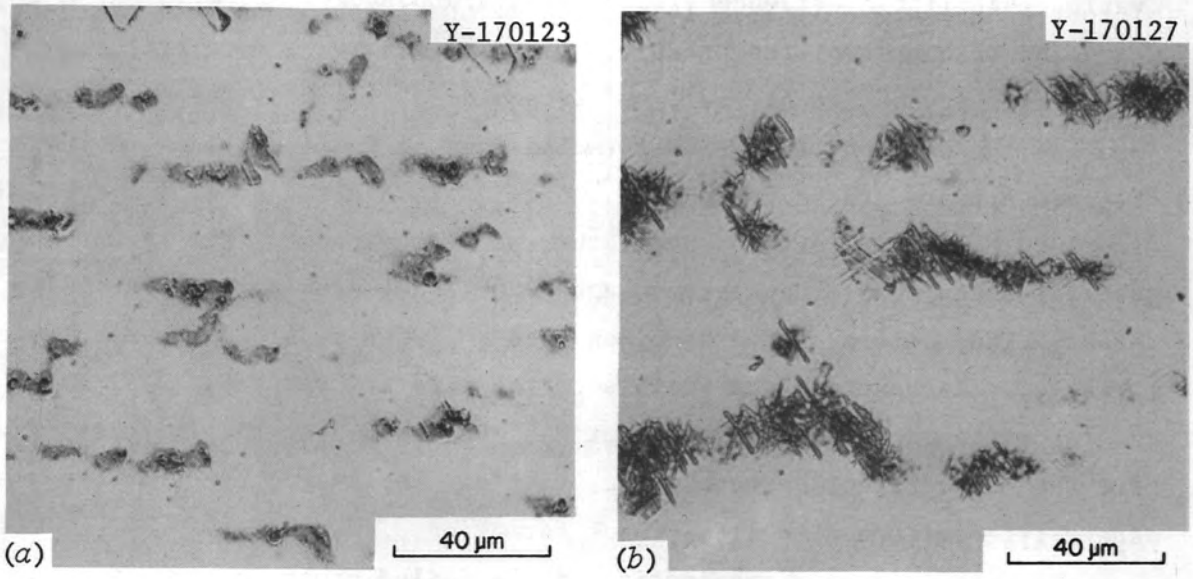

Fig. 15. Matrix precipitate in ERNiCr-3 weld metal aged (*a*) 200 h and (*b*) 1000 h at 677°C.

For the ERNiCr-3 weld metal aged for prolonged periods (10,000 h at 510°C), a lamellar precipitate formed on some grain boundaries (Fig. 16). This appears to be a discontinuous precipitate that forms in conjunction with grain boundary migration (Fig. 17). These discontinuous precipitates apparently formed when the grain boundary moved through a region where solute was rejected during the dendritic solidification [Fig. 17(*b*)]. We were unsuccessful in analyzing the lamellar precipitate by energy dispersive x-ray spectroscopy to determine if it was rich in any element relative to the matrix. The size of the precipitate was too small for such an analysis.

The microstructure of the unaged wrought ERNiCr-3 was typical of that for a well-annealed structure [Fig. 18(*a*)]. During aging at 566 and 677°C, a precipitate formed on the grain boundaries and twin boundaries [Fig. 18(*b*)]. No matrix precipitate could be resolved by optical microscopy.

Visual examination of the fractured tensile specimens indicated that they all necked and failed in a ductile shear-type mode. Similar observations were made for the failed wrought ERNiCr-3 tensile specimens. Optical metallography of the failed specimens verified the visual observation that little evidence existed of grain boundary crack formation.

The fractures of the creep-rupture specimens were more brittle, especially at 566°C. At 510°C on tests made at 434 MPa, the unaged specimen was the only one that showed indications of forming a pronounced neck. Failure appeared to be by ductile shear. Although the specimens aged 200 and 1000 h at 510°C did not neck, they also appeared to fail by the ductile-shear mode; the specimen aged 5000 h appeared to fail partially by shear. The failure of the specimen aged 10,000 h at 510°C was much more brittle; a large crack was visible on the external surface.

For the creep-rupture tests at 345 MPa and 566°C, the failures of all but the unaged specimen were quite brittle. The unaged specimen has a shear-type failure with little neck formation.

By metallographic examination of the failed creep specimens, we were able to verify our visual observations. The fractures of the specimens aged through 2500 h at 510°C displayed rather ductile failures: we found

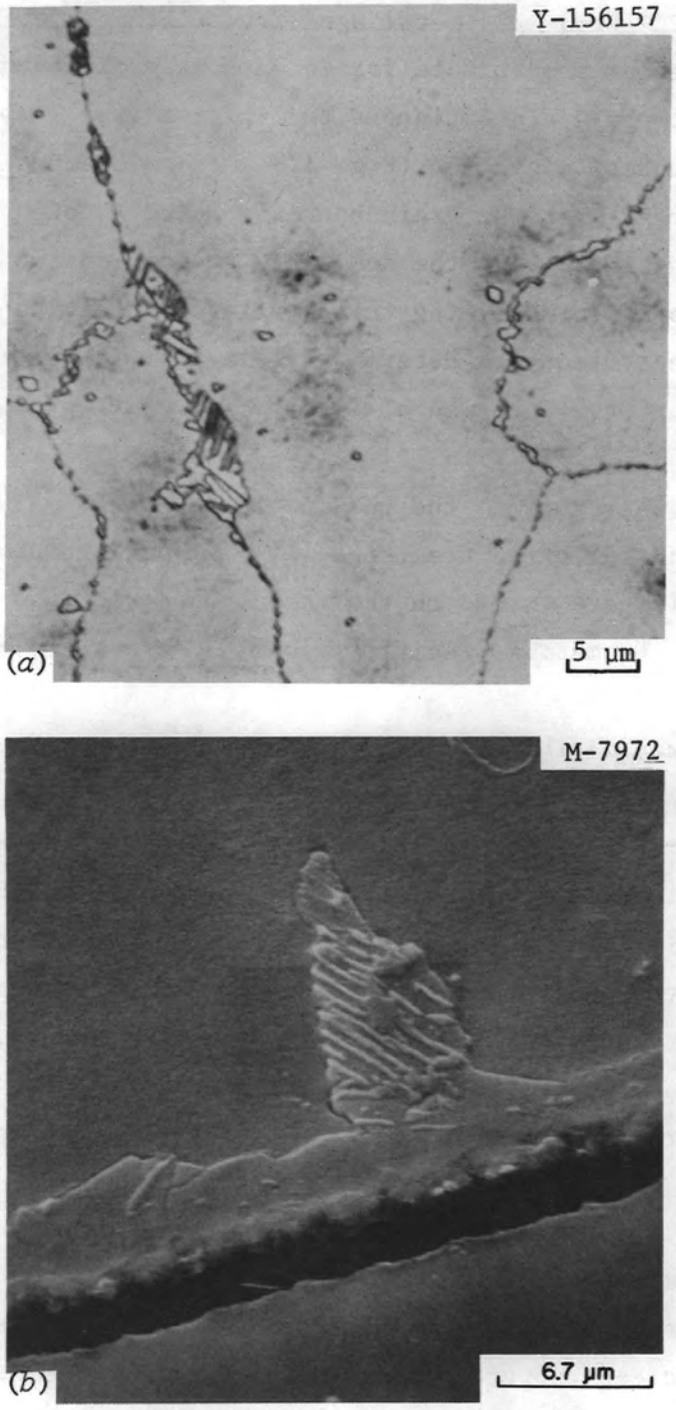

Fig. 16. Lamellar-appearing (discontinuous) precipitate formed in grain boundaries of specimen aged 15,000 h at 566°C. (*a*) Optical micrograph. (*b*) Scanning electron microscope photomicrograph.

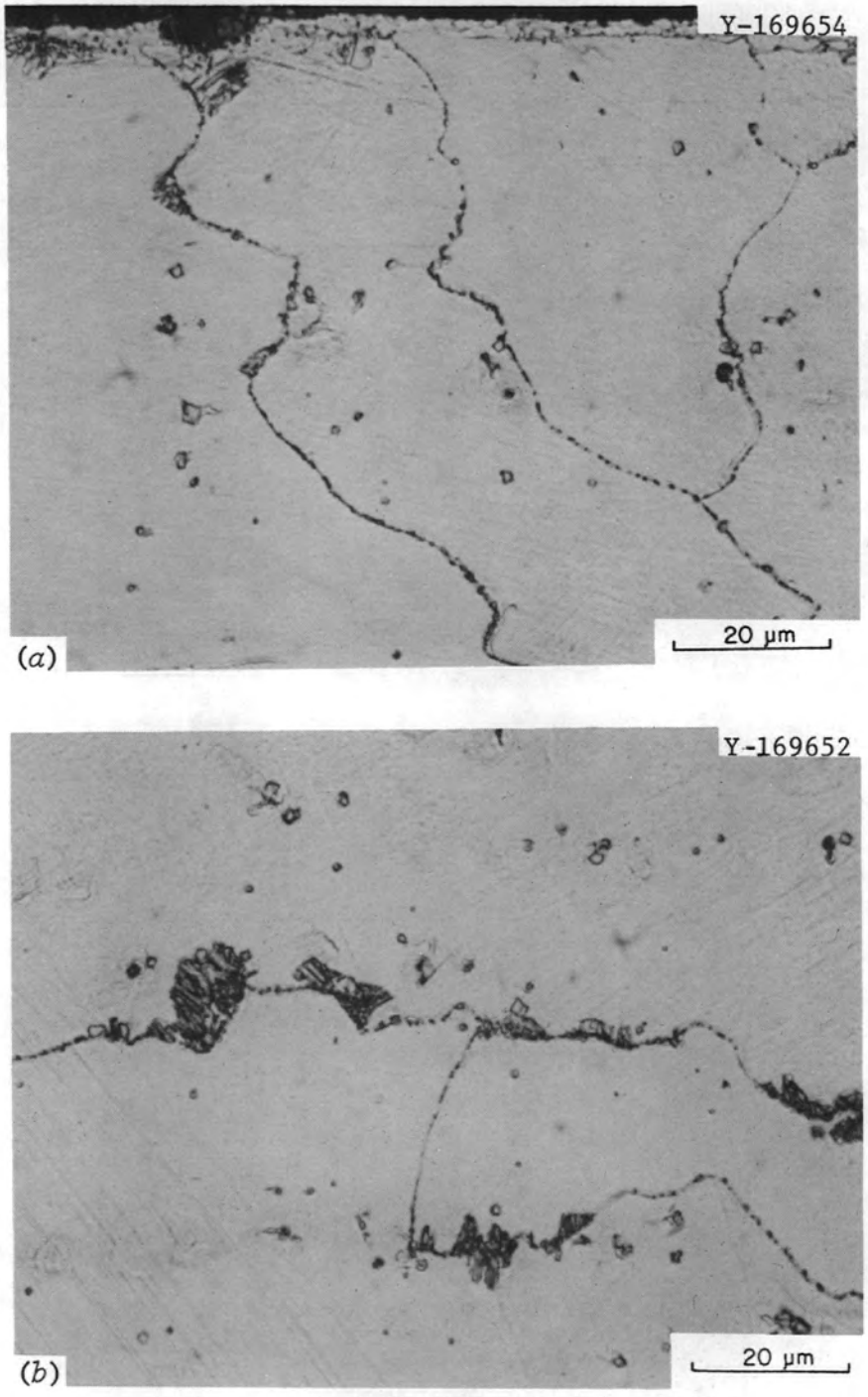

Fig. 17. Photomicrographs illustrating discontinuous precipitate formation in conjunction with grain boundary migration. Note relationship of discontinuous precipitate to the rows of matrix precipitate.

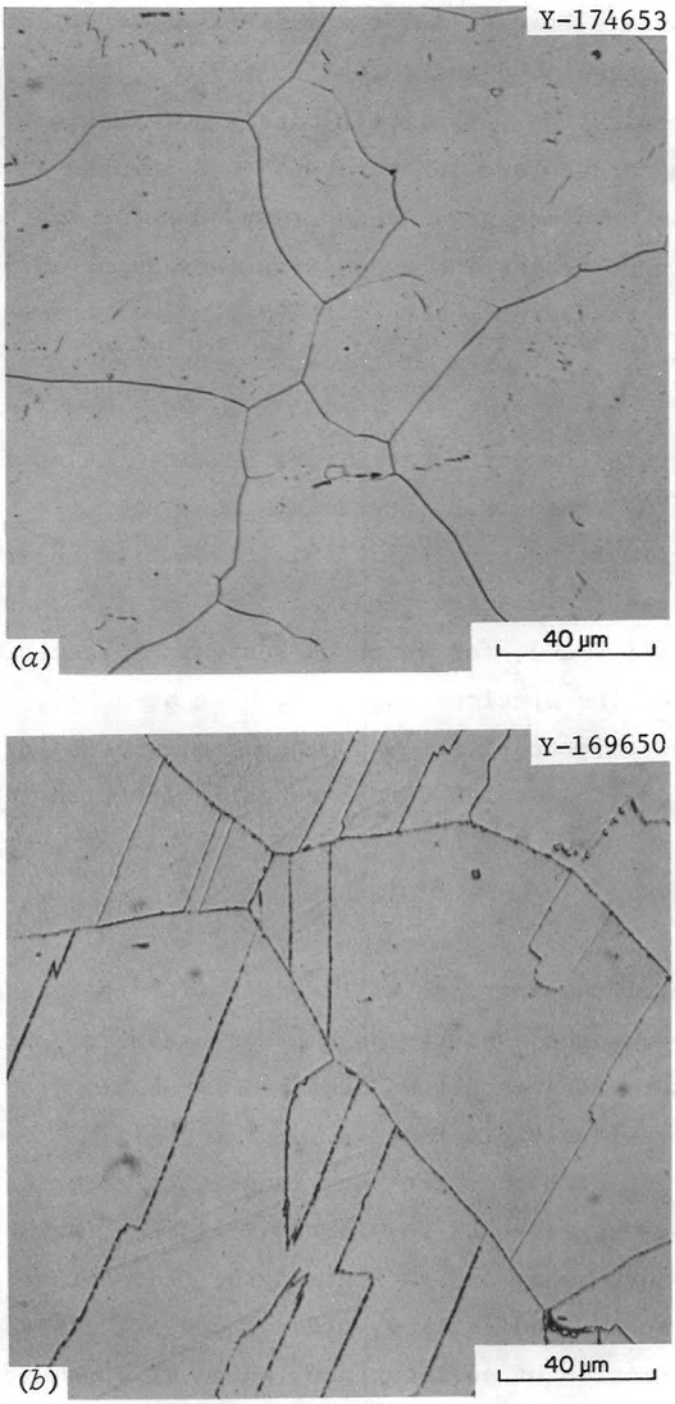

Fig. 18. Microstructure of ERNiCr-3 wrought alloy. (*a*) Unaged. (*b*) After aging for 1000 h at 677°C.

a few grain boundary cracks, but the failure was ductile transgranular. For the material aged 5000 and 10,000 h at 510°C, the specimens failed by intergranular separation with little matrix deformation.

All specimens aged and tested at 566°C developed grain boundary cracks along the specimen gage length perpendicular to the specimen axis. The number of these cracks increased with increasing aging time. Although there were transgranular aspects to the failure of the specimens aged 200 and 1000 h at 566°C, failures of the specimens aged for longer times were intergranular.

When the grain boundary cracks were examined in specimens aged 15,000 h at 566°C, they often appeared to be associated with the precipitates in the grain boundaries (Fig. 19). A similar observation was made for specimens aged for shorter times at 566°C or intermediate and longer times at 510°C (Fig. 20), for which the extent of precipitation was not as great as that for the specimens aged 15,000 h at 566°C. The relationship of the fractures to precipitate was much more evident for a long-time creep test (>12,000 h) at 566°C that recently ruptured (Fig. 21).

DISCUSSION

The origin of the increase in strength with thermal aging can be sought in two processes: short-range order and/or precipitation. The formation of short-range order in ERNiCr-3 (probably from nickel and chromium) was previously discussed.[3]

The formation of both short- and long-range order is expected to lead to changes in the strength of an alloy.[5,6] Fisher,[7] who first discussed the effect of short-range order on strength, stated that as a dislocation passes through an ordered lattice, order bonds are destroyed. This requires the formation of an interface, which will require energy above that required by a dislocation moving through the unordered lattice. Once the order on the slip plane has been reduced by the first dislocation, subsequent dislocations can move through the slip plane at a reduced stress, and the stress to move a series of dislocations along a slip plane should drop progressively.

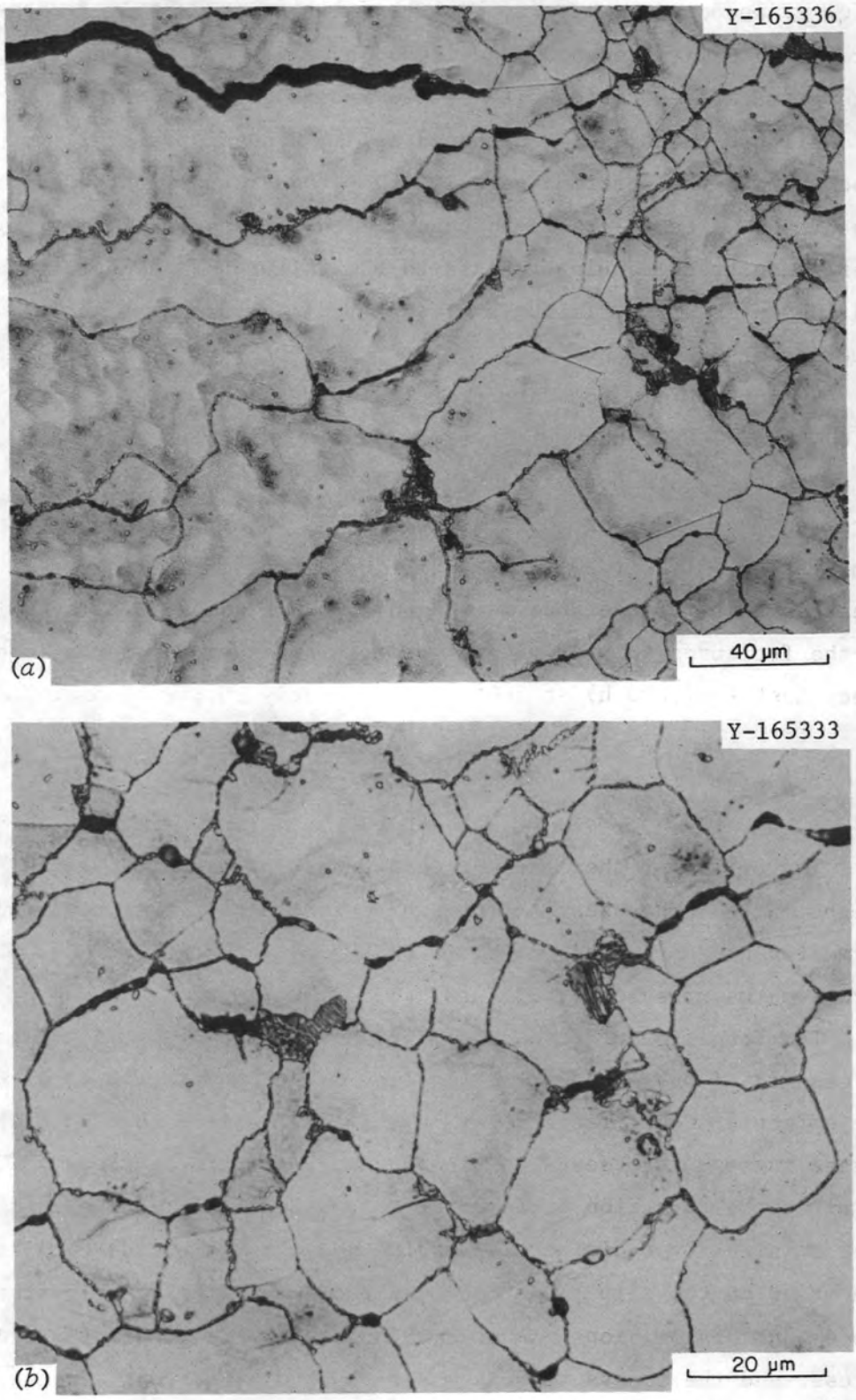

Fig. 19. Examples of grain boundary cracks that appear to be associated with the grain boundary precipitates in ERNiCr-3 weld metal aged 15,000 h at 566°C and creep tested at 566°C and 345 MPa.

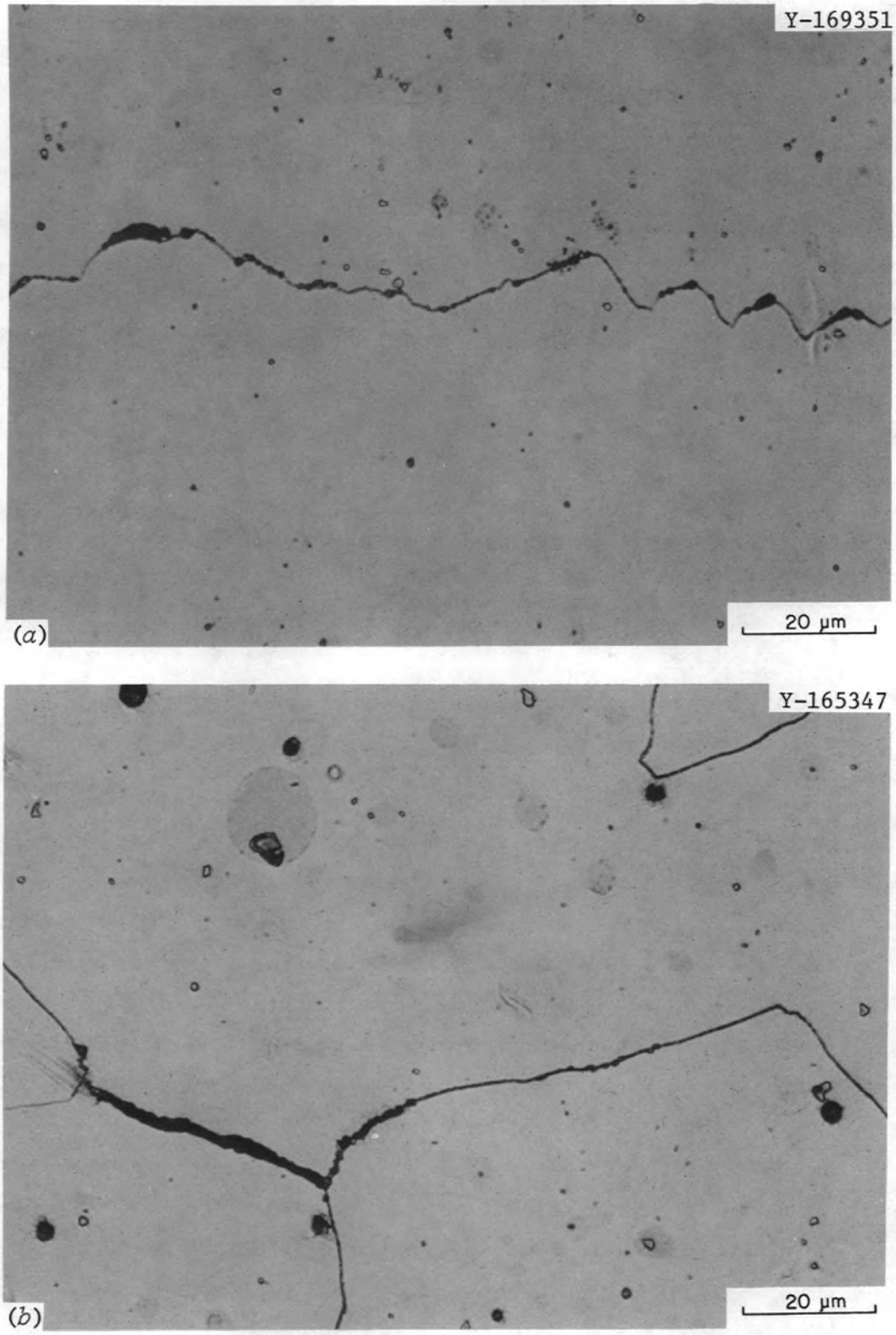

Fig. 20. Grain boundary cracks in ERNiCr-3 weld metal. (*a*) Aged 1000 h at 510°C and tested at 510°C and 434 MPa. (*b*) Aged 200 h at 566°C and tested at 566°C and 345 MPa.

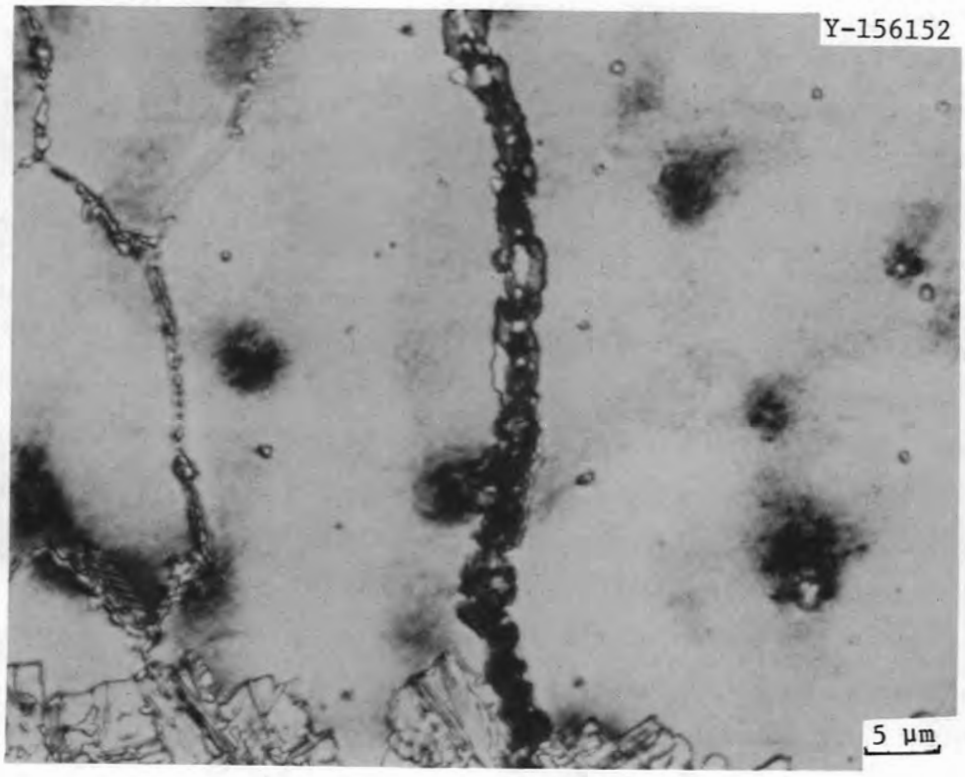

Fig. 21. Grain boundary cracks in ERNiCr-3 weld metal in long-time creep test (>12,000 h at 310 MPa and 566°C).

Cohen and Fine[8] put the Fisher mechanism on a more quantitative basis. Their calculations indicated that, when an alloy with short-range order is deformed, the passage of the first dislocation across a slip plane reduces the local order across the plane by about 20%. Subsequent dislocations require less stress to pass across the plane, as postulated by Fisher.[7] From their calculations, Cohen and Fine concluded that, after roughly two dislocations pass through a region, the next dislocation would appear to move through the lattice spontaneously because it should release energy.

A method for studying short-range order effects involves observations of slip bands on the external surfaces of deformed specimens.[6] As stated above, the first dislocation that passes across a slip plane will make the passage of subsequent dislocations across that plane easier. As a result, only a few slip planes will be active during deformation, giving rise to coarse slip. Because the amount of order is expected to increase with aging time, the slip should become progressively coarser with aging time.

To determine if coarse slip occurred, we electropolished tensile specimens after aging at 566°C. These specimens were then deformed to 4% at room temperature at a strain rate of 3×10^{-4}/s. A definite coarsening of slip occurred with increased aging time to 1000 h (Fig. 22). The specimens that were aged for 2500 and 5000 h at 566°C were similarly examined (Fig. 23). No further increase in slip-line spacing appeared to occur; if anything, a slight decrease in spacing occurred. This may have resulted from an increase in precipitate formation with time because a considerable amount of precipitate was present in the matrix of the specimens aged 5000 h at 566°C, which was not as apparent on the specimens aged for 1000 and 2500 h at 566°C (Fig. 24). As pointed out previously, this precipitate was nonuniformly distributed, being mainly confined to regions of solute rejection during solidification (Fig. 15).

Although the coarse matrix precipitate noted above could possible account for a change in the slip band morphology, no large amount of fine precipitate was detected; a high density of small precipitate particles would be required to account for the increased strength with aging time at 510 and 566°C. The precipitate at 677°C was also quite coarse. Likewise, our previous creep studies on these materials indicated that only nonuniformly distributed coarse matrix precipitates developed during the creep tests at 566, 621, 677, and 732°C (ref. 2). Thus, although the precipitate could possibly affect the slip-line spacing, it could not account for the increased strength.

The observation that the strength remained relatively unchanged after 1000 h at 677°C, even though a precipitate formed, would also seem to be an indication that the strength change at 510 and 566°C is related to short-range order. If a precipitate caused the strength increase at 510 and 566°C, an effect would also be expected at 677°C, at least for short aging times.

In previous studies we demonstrated behavioral differences in tensile[3] and creep[3,4] properties, which could be attributed to short-range order. We now feel that part of the aging response can also be attributed to short-range order. The increase in strength with aging time at 510 and 566°C is concluded to be the result of progressively more short-range order forming with time at temperature. Because the degree of short-range

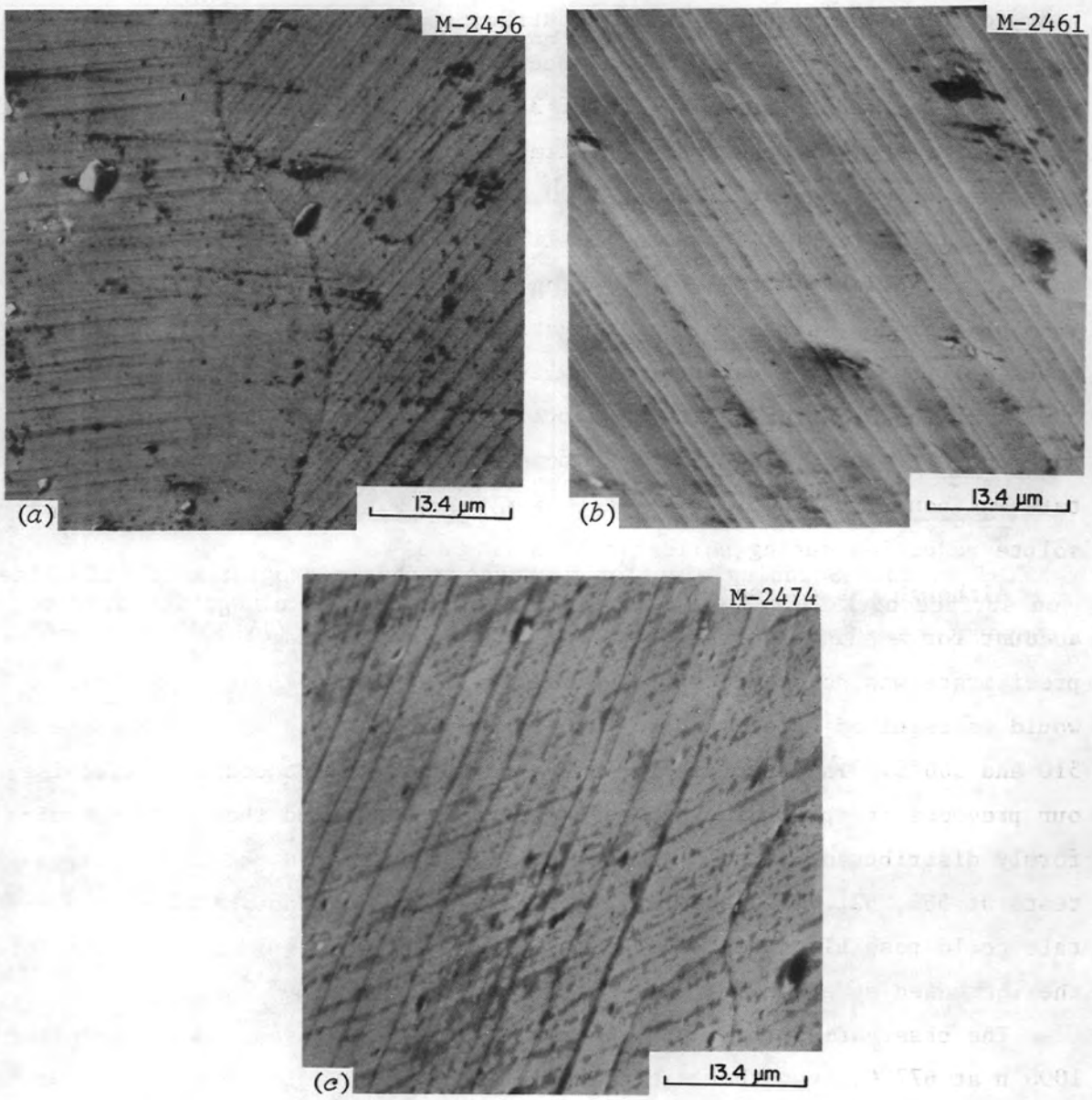

Fig. 22. Scanning electron microscope photomicrographs of slip lines on surface of ERNiCr-3 weld metal specimens after 4% elongation at room temperature. Specimens were aged for (*a*) 0, (*b*) 200, and (*c*) 1000 h at 566°C.

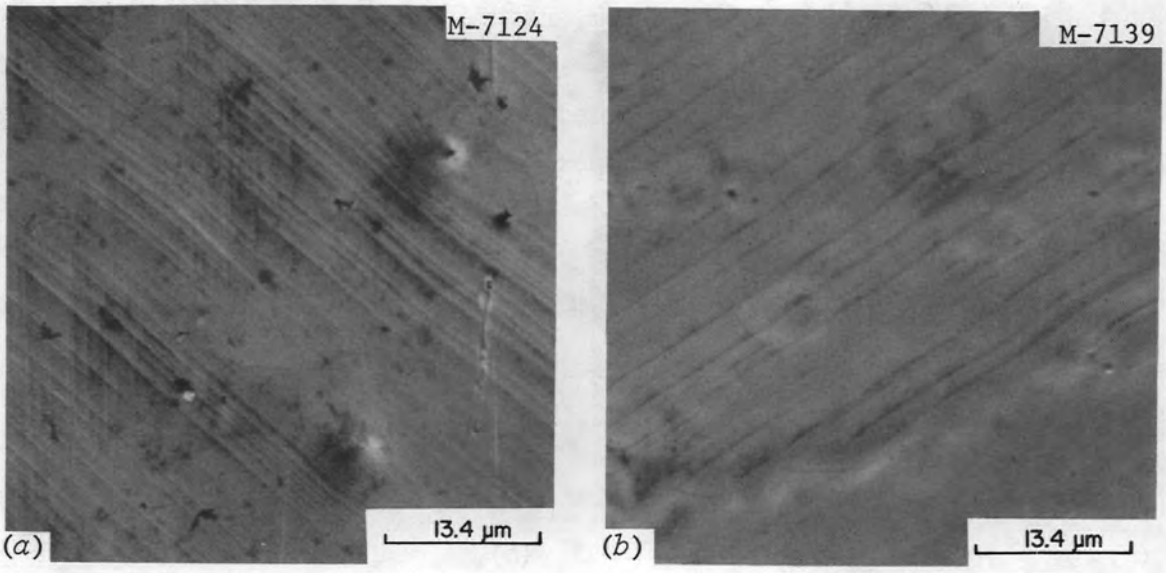

Fig. 23. Scanning electron microscope photomicrographs of slip lines on surface of ERNiCr-3 weld metal specimens after 4% elongation at room temperature. Specimens were aged for (*a*) 2500 h and (*b*) 5000 h at 566°C.

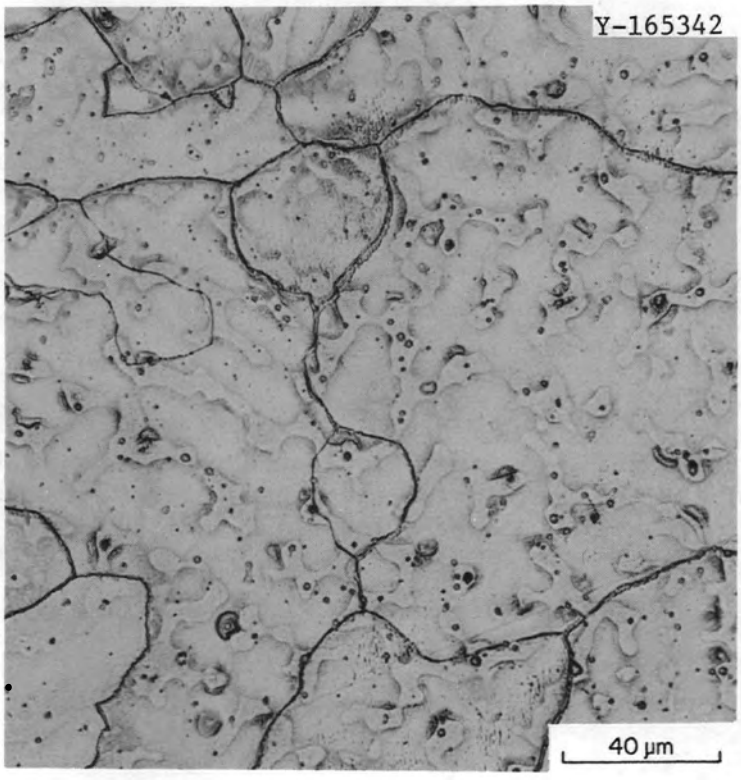

Fig. 24. Large matrix precipitates on ERNiCr-3 weld metal aged 5000 h at 566°C.

order decreases with increasing temperature, the lack of response to aging at 677°C is concluded to be the result of the decrease in the amount of short-range order with increasing temperature. In line with previous results,[2,3] we are assuming that nickel and chromium interaction cause the short-range order effects. Although the amount of short-range order will not decrease to zero above the critical temperature for long-range order, which is approximately 590°C for Ni_2Cr (ref. 9), a strong effect could be expected above this temperature, as noted in this study and previously in the tensile[3] and creep[2] studies.

A change in creep-rupture ductility and fracture mode could be expected to accompany the increase in the strength of the matrix because of short-range order formation. An increase in matrix strength relative to the grain boundary strength makes the material more susceptible to intergranular failure. Another factor that could contribute to the change in fracture mode is the precipitate that forms on the grain boundaries. If the grain boundary phase is weak, the subsequent loss in strength and in ductility would follow. The observations on the effect of thermal aging on creep-rupture strength and ductility agree with this explanation.

SUMMARY AND CONCLUSIONS

Specimens of ERNiCr-3 weld metal and wrought material were thermally aged for various times at 510, 566, and 677°C. After aging, room-temperature tensile tests were made for all aged conditions. Creep-rupture tests were made on specimens aged at 510 and 566°C. The specimens aged at 510°C were tested at 434 MPa and 510°C, while those aged at 566°C were tested at 345 MPa and 566°C. The following observations and conclusions were made:

1. The room-temperature yield strength of the weld metal aged to 10,000 h at 510°C and to 15,000 h at 566°C and the wrought alloy aged to 5000 h at 566°C showed relatively large increases in strength with aging time.

2. The room-temperature ultimate tensile strength of the weld metal aged to 15,000 h at 566°C and the wrought alloy aged to 5000 h at 566°C

showed progressive increases with aging time. The ultimate tensile strength of the weld metal aged at 510°C increased to 5000 h but decreased slightly (about 3%) between 5000 and 10,000 h.

3. The yield strength and ultimate tensile strength of the weld metal and wrought alloy aged at 677°C remained relatively unchanged.

4. The uniform elongation, total elongation, and reduction of area showed a general decrease with aging time at 510 and 566°C but showed little change at 677°C.

5. The creep-rupture strength of the weld metal went through a maximum at both 510 and 566°C; at 510°C an initial strength decrease preceded the maximum. The minimum creep rate reflected the rupture-life data (minimum creep rate decreased when rupture life increased and vice versa).

6. Although there were slight increases for intermediate aging times, the total elongation and reduction of area for the creep-rupture tests decreased for the longest aging times.

7. The increase in strength at 510 and 566°C was attributed to short-range order formation in the alloy. At 677°C, the lack of response to aging was concluded to be the result of the decrease in short-range order formation with temperature.

8. A grain boundary precipitate formed during thermal aging. This phase is believed to be the cause of the reduced creep-rupture strength and ductility at long aging times at 510 and 566°C.

ACKNOWLEDGMENTS

Special thanks are due to several people. J. L. Griffith and L. T. Ratcliff carried out the experimental work. C. W. Houck and T. J. Henson did the metallography. The manuscript was reviewed by S. A. David, B. C. Leslie, C. R. Brinkman, and G. M. Slaughter. Irene Brogden edited the manuscript, and Patsy Dickens prepared the final copy.

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