STRUCTURAL STABILITY OF UDIMET-500, A NICKEL-BASE SUPERALLOY

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ABSTRACT

The microstructure of Udimet-500 alloy was studied by optical microscopy, transmission electron microscopy, and electron diffraction after different heat treatments. These included short-time aging after different solution treatments and, also, exposure both without stress and under stress for up to 10,000 hours at various temperatures after a three- or four-stage heat treatment

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INTRODUCTION

Udimet-500 is a wrought nickel-base superalloy extensively used in gas turbines. The use of these engines at increasingly higher temperatures and for longer service requires a more detailed knowledge of the evolution of mechanical properties and, in turn, of microstructures under service conditions. In this study the microstructure of Udimet-500 was investigated by optical microscopy, transmission electron microscopy of replicas or thin foils, and electron diffraction. The heat treatments studied were

(1) Holding at 1175 C or 1050 for various lengths of time and cooling at different rates

(2) Aging at 760 C or 840 C after a one- or two-stage heat treatment at higher temperatures

(3) Exposure up to 5000 hours at 700-950 C after a three- or four-stage heat treatment

(4) Exposure up to 10,000 hours at 700-1000 C with different stress levels at each temperature after a three- or four-stage heat treatment.

The composition of the 5/8-inch bars of the alloys used in this study is given in Table 1.

Udimet-500 alloy is normally subjected to a three- and four-stage heat treatment. The three-stage heat treatment aims at producing the maximum high-temperature properties in forged material and consists of a solution treatment and a double aging as follows:

Solution treatment:	1080 C for 4 hours and air cooled
Aging treatment:	840 C for 24 hours and air cooled
Aging treatment:	760 C for 16 hours and air cooled

The four-stage heat treatment is used for bar stock with no subsequent forging. It involves, prior to the treatments quoted, an anneal for 2 hours at 1175 C, followed by air cooling.(1)**

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** References are shown on page 8.

PRELIMINARY INVESTIGATION ON HEAT-TREATING CONDITIONS

Solution Treatment

As pointed out above, the industrial heat treatment of Udimet-500 is rather complex and involves three or four stages. The preliminary investigation was carried out to gain a better understanding of the effect of heat-treating conditions on grain size and on the nature and distribution of precipitated phases.

Holding for 4 hours at temperatures from 850 to 1200 C followed by a water quench showed that the original ASTM 8 grain size remained unaffected to 1000 C. At 1100 and 1200 C the grain size increased considerably to ASTM 2 and ASTM 1, respectively. The initial γ' particle size of about 500 A increased to 3000-4000 A at 1000 C. At 1100 C and above, the γ' precipitate went into solution. The effect of cooling rate on hardness after holding for 4 hours at 1200 C is shown in Figure 1. Because of the rapid precipitation of γ' during cooling, the hardness increased as the cooling rate decreased. Furnace cooling resulted in a coalescence of the γ' particles and, hence, in a decrease in hardness.

The effect of temperature, time at temperature, and cooling rate on the hardness of Udimet-500 is shown in Figure 2. Hardness increases with increasing temperature or decreasing cooling rate, and remains unaffected with time at 1175 C. The higher hardness obtained after treatment at 1175 C is explained by the fact that, at this temperature, MC-type carbides go progressively into solution and, thus, more titanium is available for precipitation during cooling. The rapid precipitation of γ' during cooling is illustrated in Figure 3 for specimens water quenched from 1175 C. The γ' particles are coherent and about 40 A in diameter. The presence of the ordered γ' particles is also demonstrated by the occurrence of dislocation pairs, as shown in Figure 3b, and of superlattice spots on electron-diffraction diagrams. After air cooling from 1175 C the γ' particle size is about 500 A. Examination of the water-quenched specimen shows that no $M_{23}C_6$-type carbides are present. In air-cooled specimens, such carbides were identified to be in epitaxy relative to the fcc matrix.

For specimens heat treated for 4 hours at 1080 C, the same relationships are generally valid as for those treated at 1175 C. As stated above, the grain size is appreciably smaller after the lower solution temperature: ASTM 4-5 instead of ASTM 2-3. The γ' particles are about 50 A in diameter after water quenching and about 300-650 A after air cooling. Because of the small grain size, boundary networks of carbides have not been observed.

Specimens subjected successively to 2 hours at 1175 C/air and 4 hours at 1080 C/air shows a microstructure similar to that obtained after the single treatment at 1175 C except that discontinuous carbide networks are present at the grain boundaries. (Figure 4). The carbides have precipitated at grain boundaries from the saturated solid solution which resulted from the 1175 C treatment during cooling and during the 1080 C treatment. The γ' size is 300-600 A.

Aging

In the aging experiments, specimens were aged for up to 64 hours after three different solution treatments: 1175 C/2 hr/air 1080 C/4 hr/air, and 1175 C/2 hr/air + 1080 C/4 hr/air.

Figures 5 to 10 reproduce the corresponding optical micrographs. Examination of these micrographs shows that the solution treatment at 1175 C results,on holding both at 760 and 840 C, in the heterogeneous precipitation of carbides at preferred areas: grain or twin boundaries, massive MC-type carbides, etc. (Figures 5 and 6). Specimens subjected to a treatment at 1080 C do not exhibit this behavior. In this case, the carbides precipitate mainly at the grain boundaries. Specimens subjected to the industrial three- or four-stage heat treatment show a similar behavior. The heterogeneous precipitates of carbides after the treatment at 1175 C result, presumably, in unsatisfactory mechanical and oxidation properties.

The hardness variation after the aging treatments described above is reproduced in Figures 11 to 13.

After the two-stage solution treatment and aging for 16 hours at 760 C, the γ' particle size is 450-700 A. For the same solution treatment and aging for 16 hours at 840 C, the γ' size is 600-1200 A. The $M_{23}C_6$ carbides precipitate at the grain boundaries in the form of rods. The fcc carbides with a lattice parameter approximately three times that of the matrix are in epitaxial relationship with the latter (Figure 14). Specimens subjected, after the two-stage solution treatment, to the two-stage aging treatment exhibit a γ' particle size of 350-1650 A. It is apparent that, during the 760 C aging, small γ' particles have precipitated. In this case also, $M_{23}C_6$ carbides form at grain boundaries in epitaxy with the matrix.

PROLONGED AGING REACTIONS

Three-Stage Heat Treatment

Specimens subjected to the three-stage heat treatment were exposed for up to 5,000 hours at temperatures from 700 to 950 C. The hardness variation during the prolonged treatment is reproduced in Figure 15.

Figures 16, 17, and 18 reproduce the microstructures obtained at different temperatures after 100, 1000, and 5000 hours, respectively.

At 700 C no structural modification is apparent, even for the longest time considered. As shown in Figure 19 the particle size is 350-1700 A, i.e., the same as after the initial heat treatment. For the longer aging time the grain boundaries become more readily etched because of the increase in the amount of precipitated carbides. As the temperature increases the precipitation reactions become more pronounced. The γ' precipitates become visible on the optical micrographs, whereas the grain-boundary reactions results in their depletion of γ'.

The γ' particles reach a size of 1800-14,000 A after 100 hours at 950 C and of 7,000-18,000 A after 5,000 hours at this temperature. Figure 20 shows that the large γ' particles have lost their spherical shape. They still have the same lattice parameter and remain apparently coherent. Along with the larger γ' particles, much smaller particles (50-150 A) are visible, which precipitated during the cooling from the aging temperature. $M_{23}C_6$ carbides have coarsened at the grain boundaries and also within the grains.

Four-Stage Heat Treatment

Specimens subjected to the four-stage heat treatment were submitted to the same prolonged aging reactions as above.

Figure 21 reproduces the variation of hardness after the aging treatments.

Figures 22, 23, and 24 shows the microstructures of the alloy after 100, 1,000, and 5,000 hours, respectively, at different temperatures. In all series of specimens the grain size is ASTM 2-3 instead of ASTM 4-5 as after the three-stage treatment. In these specimens the precipitation pattern is similar to that observed for the specimens subjected to the three-stage heat treatment. The grain-boundary reactions appear, however, to be more pronounced. Grain-boundary $M_{23}C_6$ carbides appear to be embedded within a phase that cannot be identified on the optical micrographs. Thin-foil transmission dark-field images showed that the carbides are surrounded by a continuous γ' phase. The $M_{23}C_6$ carbides also remain coherent with the matrix because of the parallel orientation of the lattices. Figure 25 shows typical views of carbides at grain boundaries and within the grains in specimens aged at 700 C.

The effect of prolonged exposure at elevated temperature on the room-temperature mechanical properties is illustrated in Table 2.

The prolonged exposure at 700 C results in a slight increase in the strength properties and a corresponding slight decrease in elongation. For the higher aging temperatures, a decrease in strength properties and a corresponding increase in elongation are observed. These observations are in agreement with the structural observations.

Prolonged Aging Under Stress

Three-Stage Heat Treatment

Specimens subjected to a three-stage heat treatment were maintained at 700 C for 1000 hours under a stress of 7 or 35 kg/mm^2 and, for the same period, at 850 C under a stress of 3 or 15 kg/mm^2. Metallographic examinations were made of the reduced section and of the practically unstressed head of the creep specimen.

The microstructures obtained of the specimens aged at 700 C are shown in Figure 26. There is a slight tendency for the grain boundaries to widen as the stress level increases. Figure 27 shows thin-foil transmission electron micrographs of the specimen stressed with 7 kg/mm^2. The γ' particles remained at practically the same size (500-1750 A) as before the stress-aging treatment. The grain-boundary carbides are in epitaxy with the matrix. Dislocation networks showed that they have crossed the γ' particles.

The effect of this treatment on the room-temperature properties is shown in Table 3. Increasing the stress had no appreciable effect on the room-temperature tensile properties except for a decrease in elongation for the alloy that was creep tested at 850 C.

The effect of aging treatment under stress at 850 C on the microstructures is shown in Figure 28. Compared with the microstructure obtained without stress (Figure 17), the grain boundary appears to be wider. The difference between the reduced section and that of the head of the specimen is only slight. The γ' particles have reached a size of about 5,000 A as shown in Figure 29. Comparing the results obtained without stress, it is apparent that coarsening of the γ' particles is due only to the temperature. The grain boundaries were shown to consist of discontinuoue $M_{23}C_6$ particles surrounded by γ'. A similar result was observed in optical micrographs of specimens exposed for 1,000 hours at 950 C (Figure 17).

The room-temperature tensile strength (Table 3) is lower after exposure for 1000 hours at 850 C under stress than after exposure under similar conditions at 700 C.

Four-Stage Heat Treatments

Specimens were subjected first to the four-stage heat treatment and then creep-rupture tested at temperature between 650 and 1,000 C and at stresses between 6 and 60 kg/mm^2. Rupture time varied, depending on the stress and temperature, from a few hours to about 10,000 hours.

Examination of the reduced section and of the head of the ruptured specimens revealed that the γ' particle size was dependent only on temperature and rupture time. This confirmed the findings (reported in the previous section) concerning the specimens subjected to the three-stage heat treatment. Figure 30 shows the relative increase in γ' particle size as a function of creep-rupture time for different temperatures.

The grain-boundary reactions were assessed by their thickness as measured on optical micrographs obtained on similarly etched specimens. It was shown clearly that the carbide reactions were influenced largely by the stress conditions. Figure 31 illustrates the relative increase in grain-boundary thickness for the tests at 850, 900, and 950 C. The precise knowledge of the structural evolution of the alloy may prove useful in assessing the stress and temperature in a turbine blade when its actual time in service is known.

DISCUSSION OF THE RESULTS

For all the treatments considered in this study, the only phases identified were γ' or $Ni_3(Al, Ti)$, MC carbides, and $M_{23}C_6$ carbides. These observations are in agreement with those of Hagel and Beattie.[2] However, no evidence was found for the presence of a fcc X phase, reported by Palty and Kaufmann[3] to precipitate in the grains. Hagel and Beattie[2] have also observed the presence of sigma phase in a double-aged Udimet-500 alloy that was creep tested for 5,000 hours at 1350 F (732 C). In this study, no evidence was obtained of the presence of sigma phase either in the specimens creep tested for 1,000 hours or in the specimens aged for 5,000 hours. This behavior is in agreement with predictions from electron-vacancy calculations[4], according to which Udimet-500 should not be considered as a sigma-prone alloy. The morphology of γ' and $M_{23}C_6$, as affected by heat treatments, to a large extent governs the properties of the alloy. Several investigations reported that optimum creep properties in nickel-base alloys are achieved by a large grain size, the precipitation of $M_{23}C_6$ carbides at the grain boundaries, and the precipitation of γ' by suitable aging treatments.[2,3,and 5] An anneal treatment at 1175 C gives rise to a large grain size (ASTM 2-3) and results in the solutioning of all phases except that of MC carbides. Aging in this condition results, however, in the heterogeneous or cellular precipitation of $M_{23}C_6$ carbides from the supersaturated solid solution which is considered to be deleterious.[4]

An additional treatment at the intermediate temperature 1080 C results in the partial precipitation of $M_{23}C_6$ carbides, preferable at the grain boundaries, and prevents, during subsequent aging, the heterogeneous precipitation of these carbides. In this respect the solution temperature of $M_{23}C_6$ carbides should be slightly higher than 1080 C rather than between 980-1080 C as reported by Kaufmann and Palty.[3] The solution treatment at 1080 C, practiced alone, results in a smaller grain size (ASTM 4-5) which is desirable when a high fatigue resistance is required.

The double-aging treatment gives rise to an optimum volume fraction of γ' and particle sizes ranging from 350 to 1700 A.

Udimet-500 first subjected to a three- or four-stage heat treatment and then exposed for periods up to 10,000 hours, without stress or under stress, and at various temperatures does not form any compounds in addition to those formed during the initial heat treatment. Temperature affects both the increase in γ' size and the grain-boundary reactions, whereas stress appears to affect only the increase in grain-boundary thickness. Grain-boundary recations are more pronounced in the alloy initially subjected to the four-stage treatment, which results in a larger grain size. Gamma prime particles are initially spherical and, as the size increases, become cubic. Up to a particle size of 18,000 A the γ' particles maintain practically the same lattice parameter and remain coherent with the matrix.

The fcc $M_{23}C_6$ carbide often is observed to grow at grain boundaries or within the grains in epitaxy with the matrix. Its lattice parameter is three times that of the solid solution. In many instances the grain boundaries where carbides have precipitated are surrounded with a γ'-depleted matrix. Also $M_{23}C_6$ carbides appear to be surrounded with a γ' phase (see Figure 24). As discussed by Sims, chromium diffuses to the boundaries in order to form $M_{23}C_6$ carbide.[4] In the chromium-depleted zones the solubility of γ' increases, or the concentration of nickel and aluminum increases, and a continuous γ' phase surrounding the carbides is formed. The grain-boundary reactions are enchanced by temperature, time at temperature, and stress level. The knowledge of the evolution of γ' size and grain-boundary thickness as a function of temperature, stress, and time permits the determination of the service conditions of a turbine component when its operation time is known.

Prolonged aging without stress or under stress of Udimet-500 subjected to a three- or four-stage heat treatment does not adversely influence the room-temperature elongation. At low temperature ($\sim$700 C) a strengthening effect and a slight decrease in elongation were observed. Stress appears also to have only a small effect on room-temperature properties. The effect of stress becomes more appreciable as the creep temperature increases.

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(3) M. Kaufmann and A. E. Palty, The Relationships of Structure to Mechanical Properties in Udimet-500, Trans. AIME, 218, 1960, 107.

(4) C. T. Sims, Contemporary View of Ni-Base Superalloys, J. Metals, 1119, October 1966.

(5) W. Betteridge, The Nimonic Alloys, Edward Arnold Ltd, London, 1959.

TABLE 1. CHEMICAL COMPOSITION OF THE EXPERIMENTAL ALLOY

Component	Actual, weight percent	Nominal, weight percent
C	0.08	0.15 max
Mn	0.10	0.75 max
Si	0.10	0.75 max
Cr	18.9	15-20
Co	19.3	13-20
Mo	4.15	3-5
Ti	3.00	2.5-3.25
Al	2.97	2.5-3.25
Fe	0.15	4 max
S	0.005	0.015 max
Cu	0.1	-
Ni	Bal	Bal

TABLE 2. EFFECT OF PROLONGED AGING ON ROOM-TEMPERATURE MECHANICAL PROPERTIES

Treatment	Ultimate Tensile Strength, kg/mm^2	0.2 Yield Strength, kg/mm^2	Elongation, percent
Four-stage heat treatment	94.5	80.4	4.1
Ditto + 700 C/5000 hr/A.	96.8	86.2	3.3
" + 800 C/5000 hr/A.	83.1	66.1	5.1
" + 900 C/5000 hr/A.	70.6	52.0	6.8

TABLE 3. EFFECT OF AGING UNDER STRESS ON ROOM-TEMPERATURE TENSILE PROPERTIES OF UDIMET-500

Conditions	Ultimate Tensile Strength, kg/mm^2	Yield Strength, kg/mm^2	Elongation, percent
Three-Stage heat treatment	132.0	93.0	15.0
Ditto + 700 C/1000 hr/7 kg/mm^2	127.6	98.7	9.8
" + 700 C/1000 hr/35 kg/mm^2	128.9	97.2	10.4
" + 850 C/1000 hr/3 kg/mm^2	125.2	74.8	18.4
" + 850 C/1000 hr/15 kg/mm^2	115.8	79.8	12.9

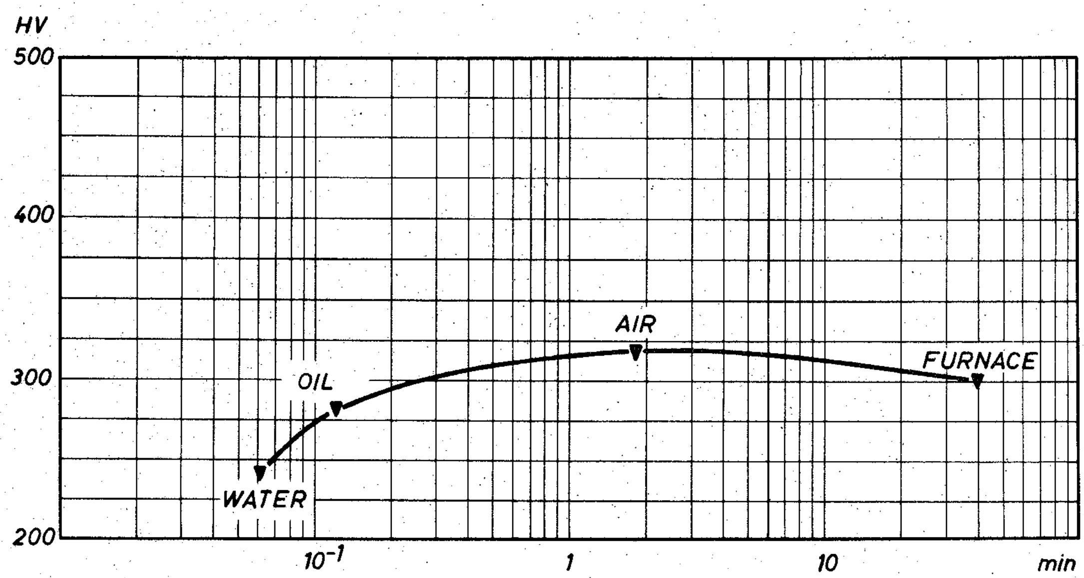


Fig. 1. Effect of Cooling Rate from 1200°C/4 hours on Hardness of Udimet 500

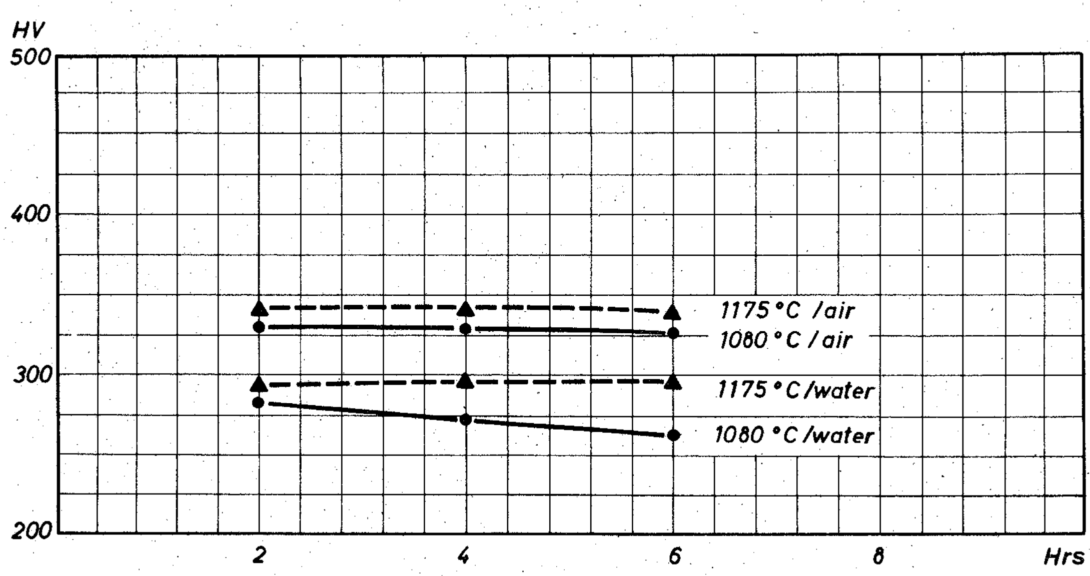
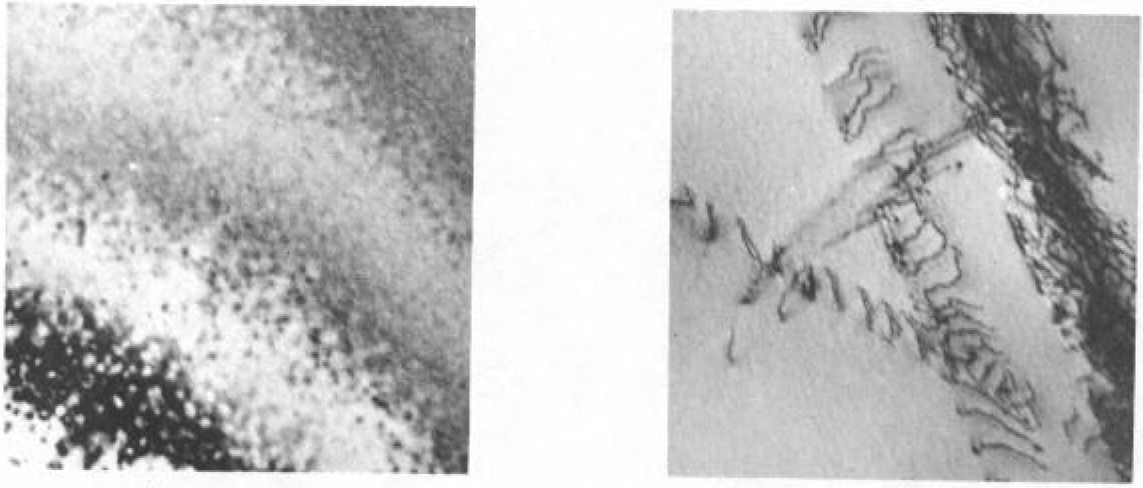
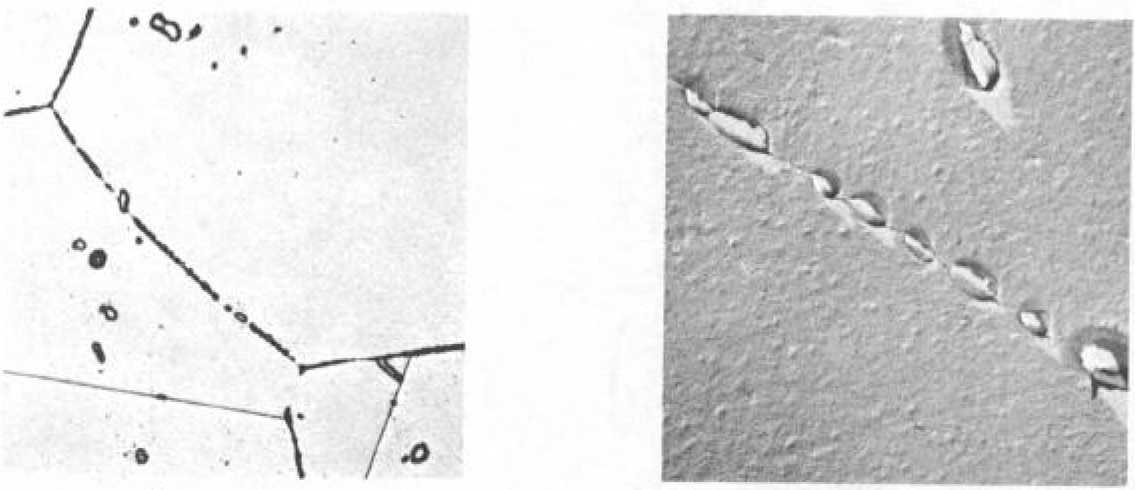


Fig. 2. Effect of Temperature, Time at Temperature and Cooling Rate on Hardness of Udimet 500



a) X 100.000 b) X 32.000

Fig. 3.. Thin Foil Electron Micrographs of Udimet 500 Water Quenched from 1175°C/2 hours.
a) γ' Ni_3 (Al-Ti) precipitates
b) Dislocations pairs.



a) X 500 b) X 5000

Fig.4. Microstructure of Udimet 500 Subjected to a Two Stage Solution Treatment : 1175°C/2 hours/air + 1080°C/4 hours/Air $M_{23}C_6$

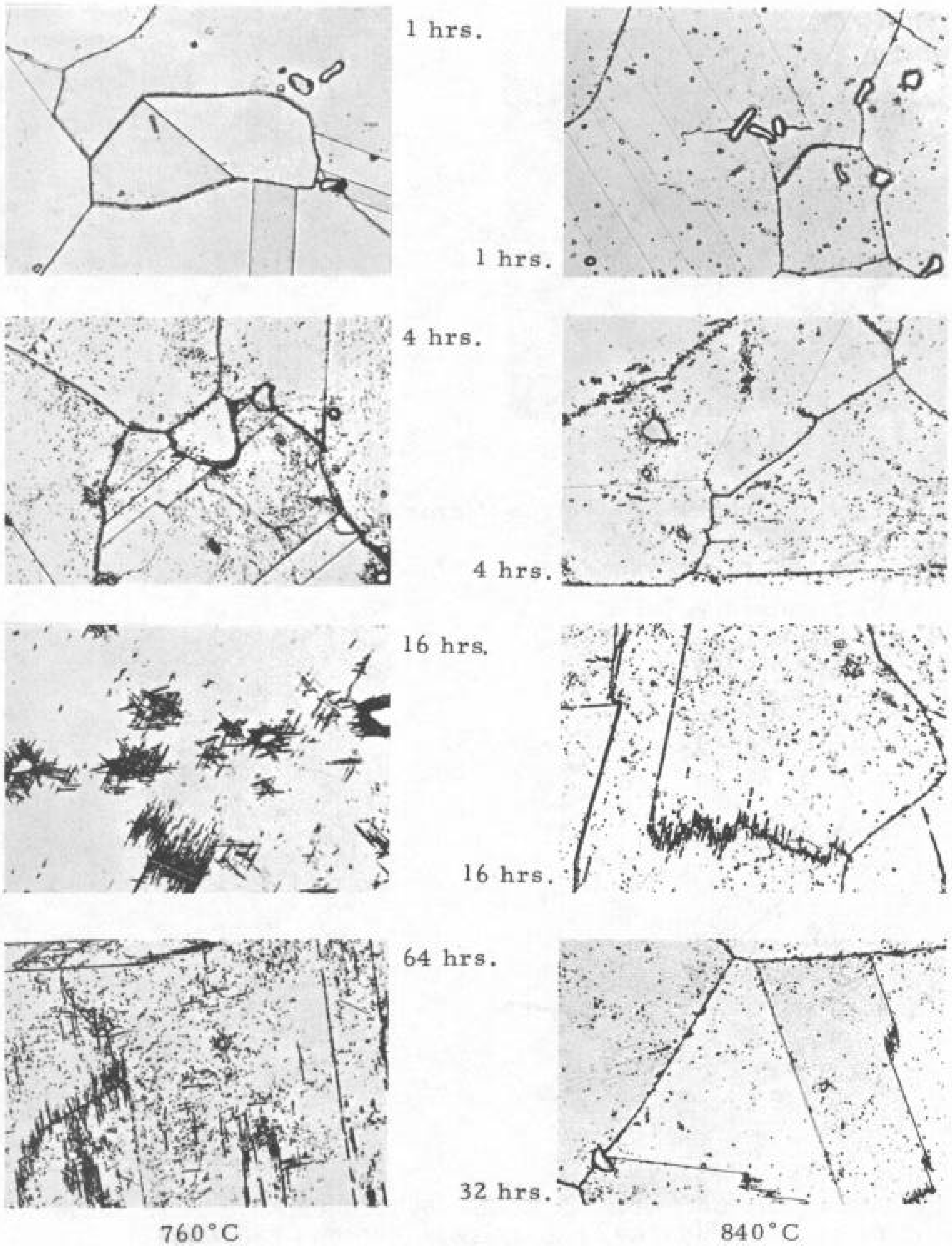


Fig. 5 and 6. Microstructures of Udimet 500 Aged for Indicated Times at 760°C and at 840°C after a Solution Treatment of 1175°C/2H/Air, Magnification X 500

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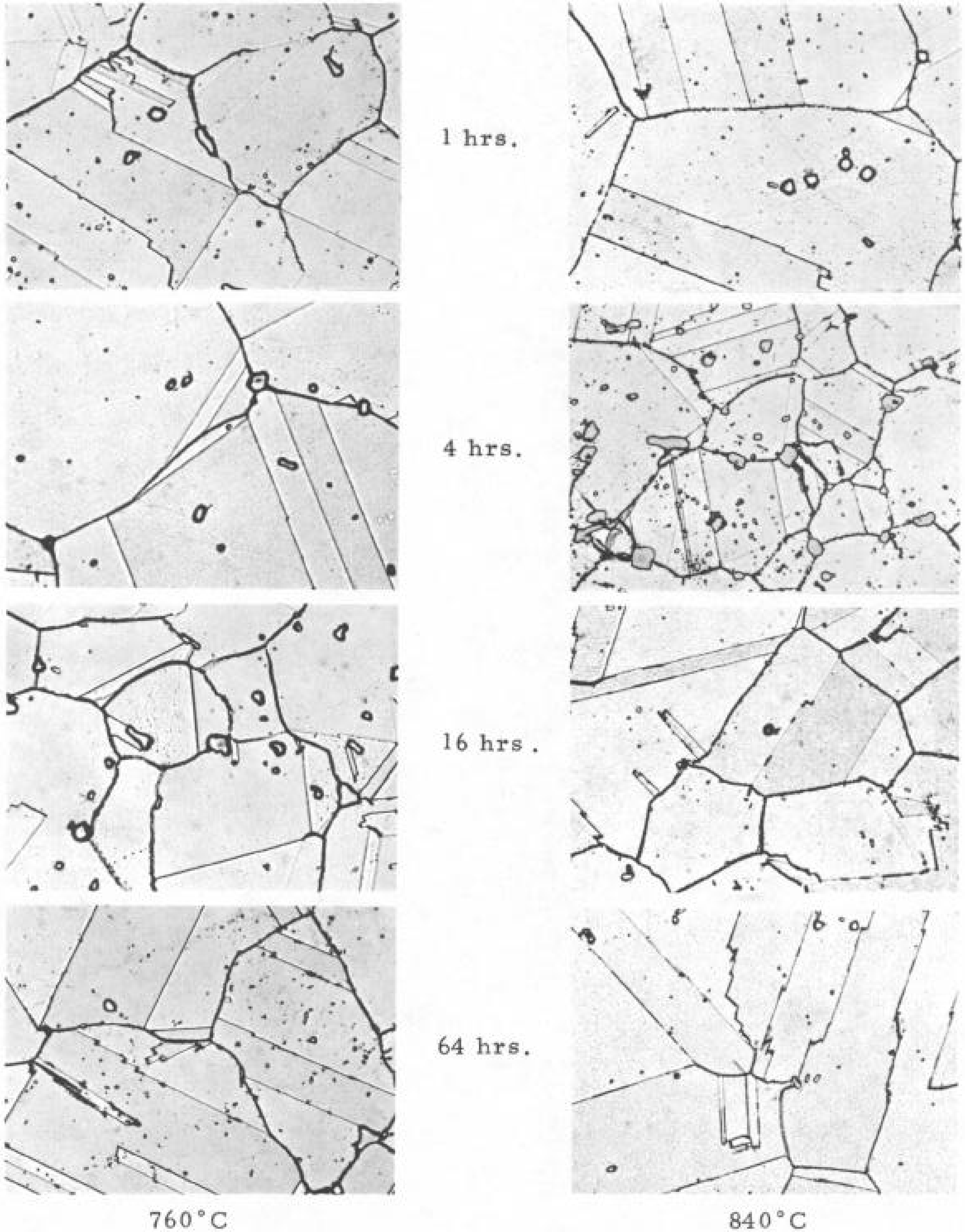

Fig. 7 and 8. Microstructures of Udimet 500 Aged for Indicated Times at 760 and at 840°C after a Solution Treatment of 1080°C/4H/Air, Magnification X 500

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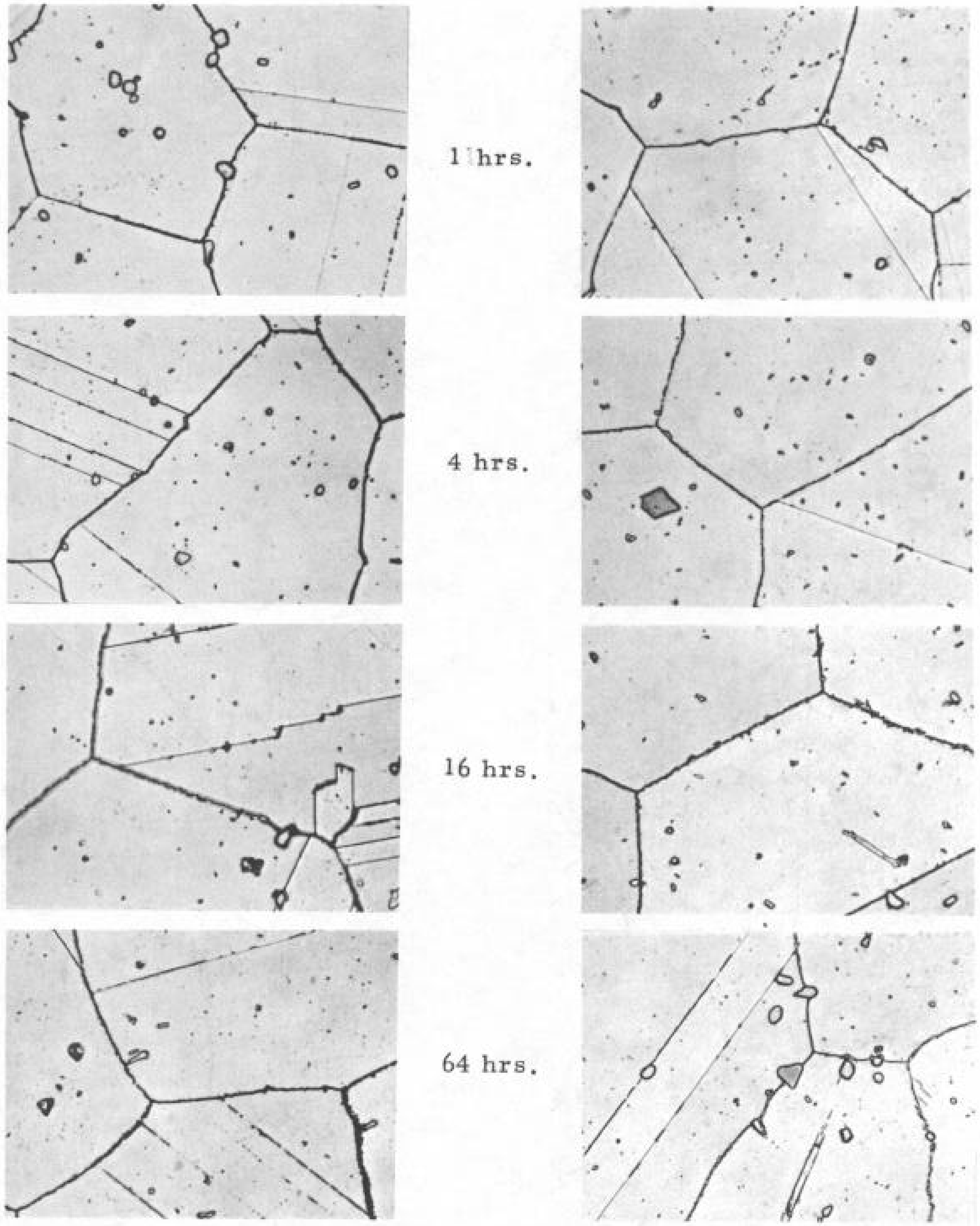

Fig. 9 and 10. Microstructures of Udimet 500 Aged for Indicated Times at 760°C and 840°C after a Solution Treatment of 1175°C/2H/Air + 1080°C/4H/Air, Magnification X 500.

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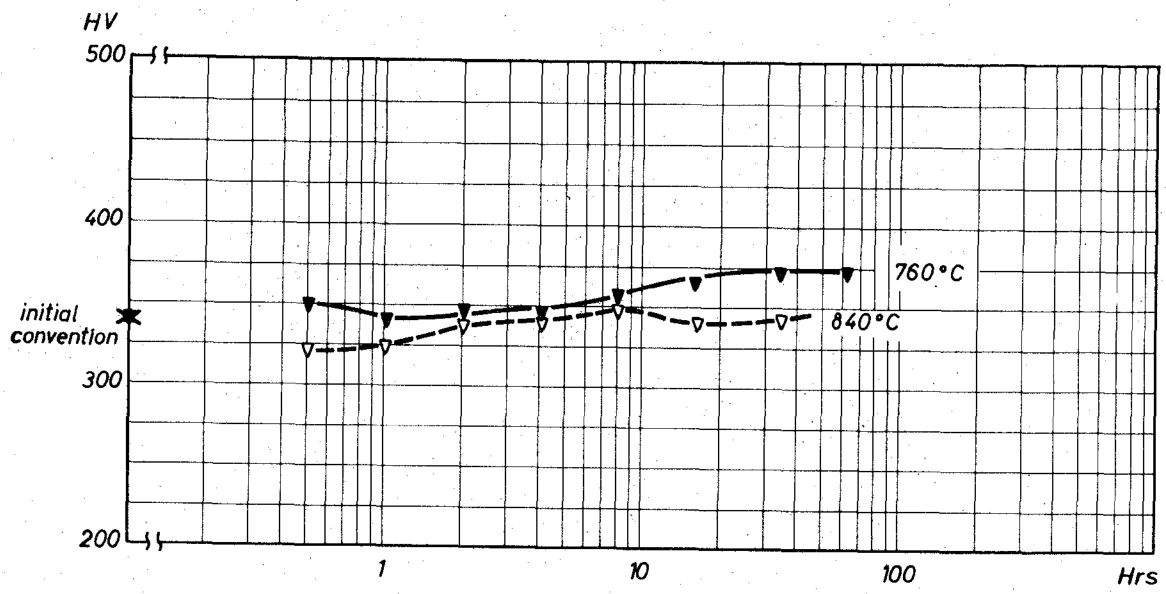

Fig. 11. Hardening Response of Udimet 500 Aged at 760°C or 840°C after a Solution Treatment of 1080°C/4H/Air.

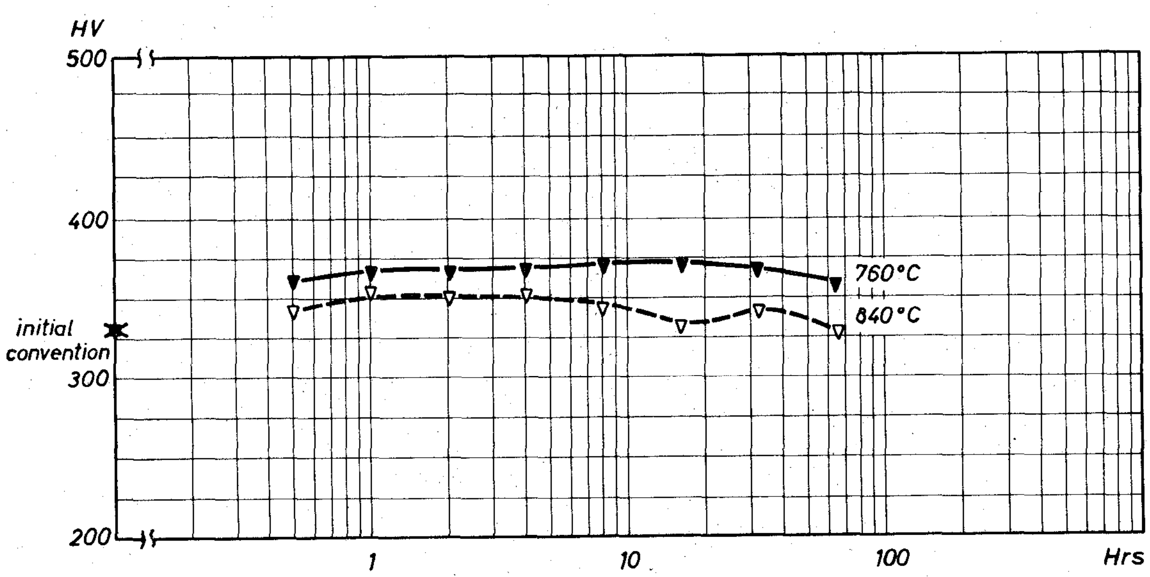

Fig. 12. Hardening Response of Udimet 500 Aged at 760°C or 840°C after a Solution Treatment of 1175°C/2H/Air.

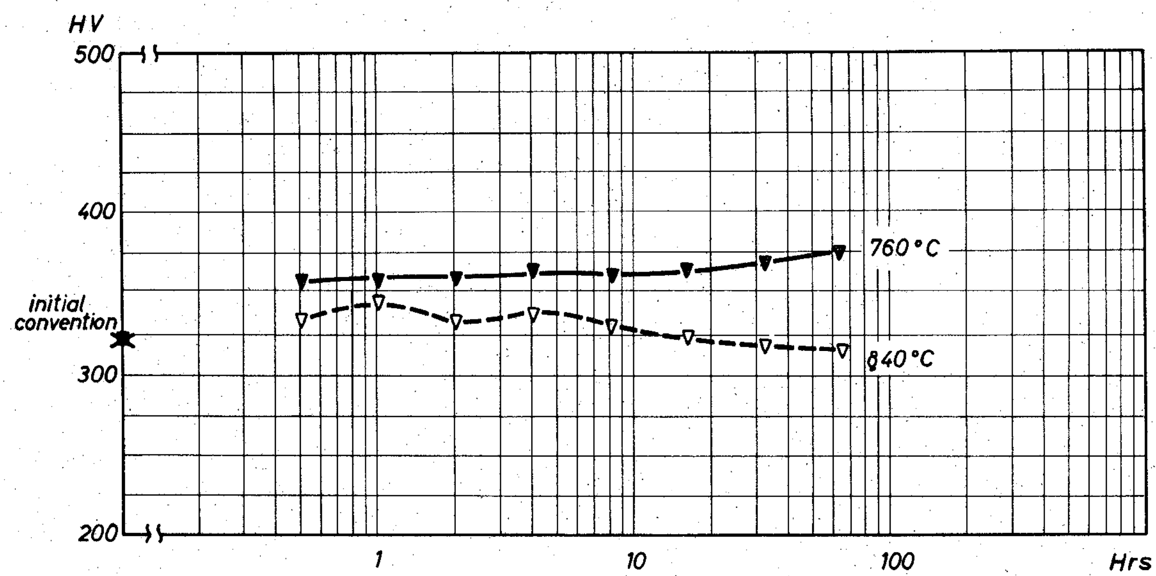

Fig. 13. Hardening Response of Udimet 500 Aged at 760°C or 840°C after a Solution Treatment of 1175°C/2H/Air + 1080°C/4H/Air

X 41,000

X 32,000

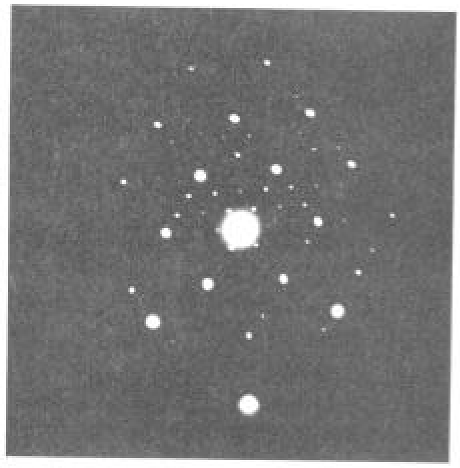

Fig. 14. Thin Foil Transmission Electron Micrographs and Electron Diffraction Pattern Showing the Epitaxy between Matrix and $M_{23}C_6$ Carbides. Heat Treatment 1175°C/2H/Air+1080°C/4H/Air + 840°C/16H/Air

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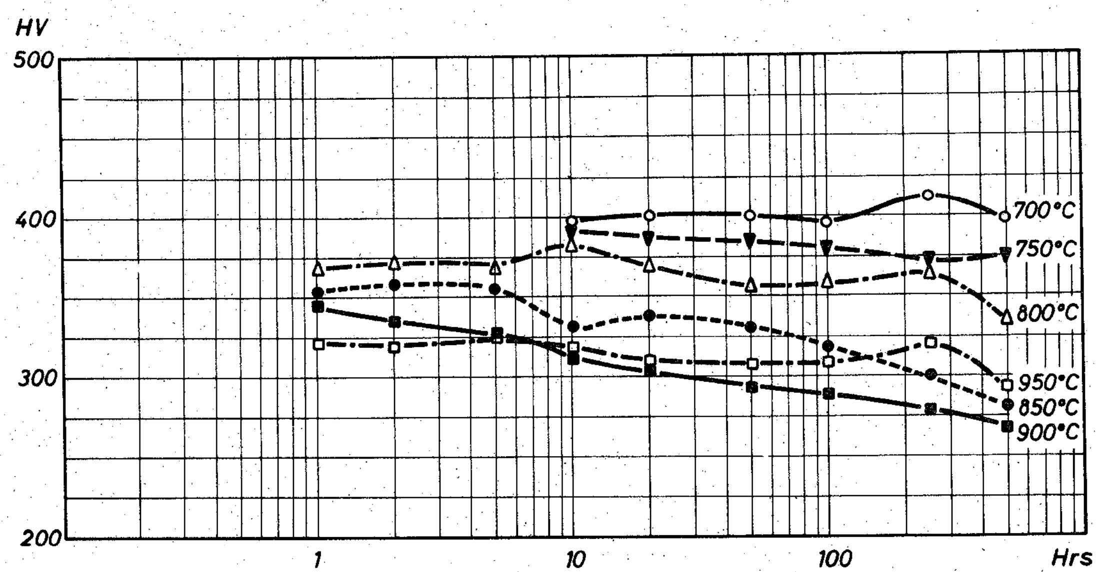


Fig. 15. Hardening Response of Udimet 500 Aged for up to 5000 Hours at 700-950°C Initial Heat Treatment 1080°C/4H/Air - 840°/24H/Air - 760°/16H/Air.

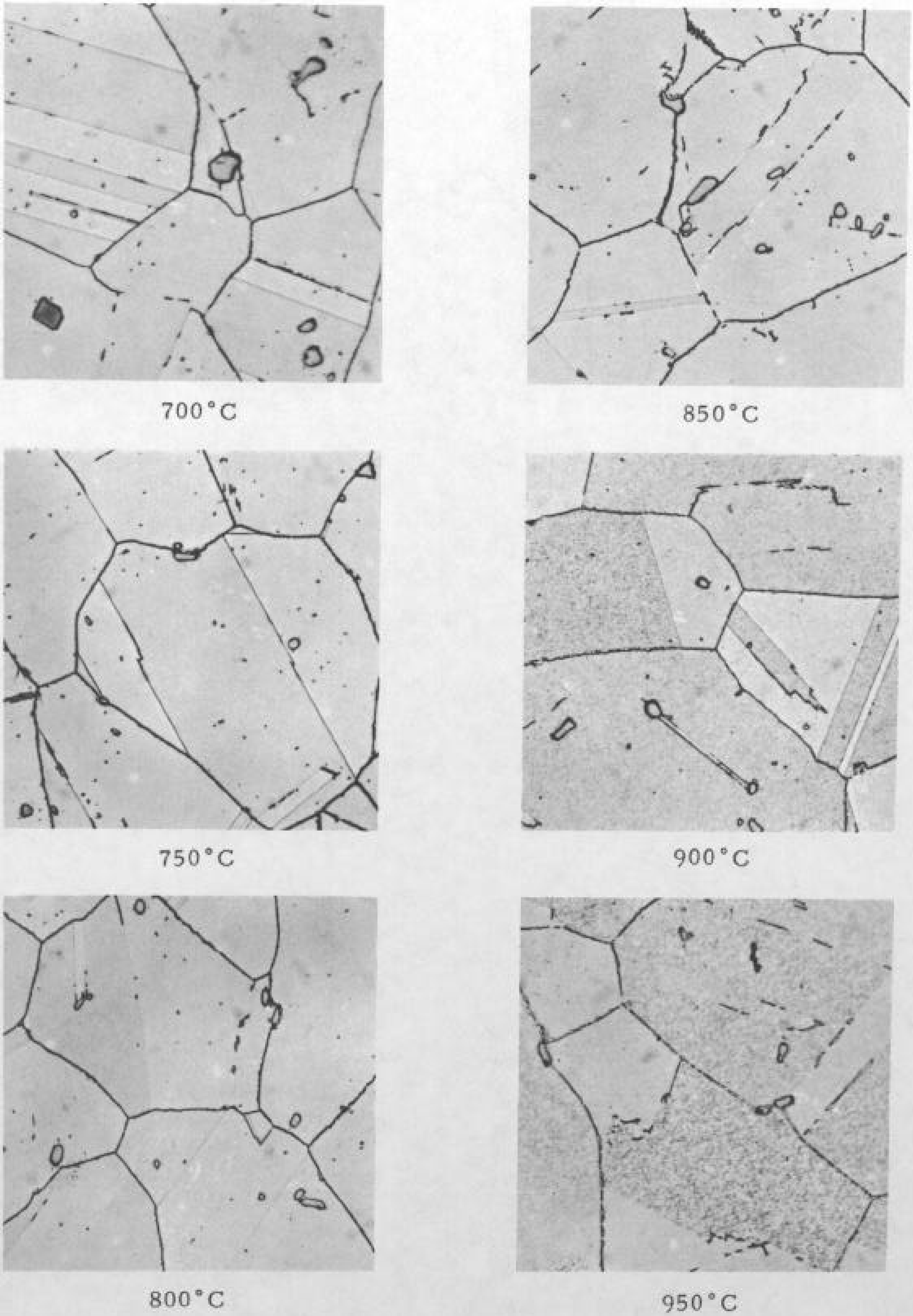
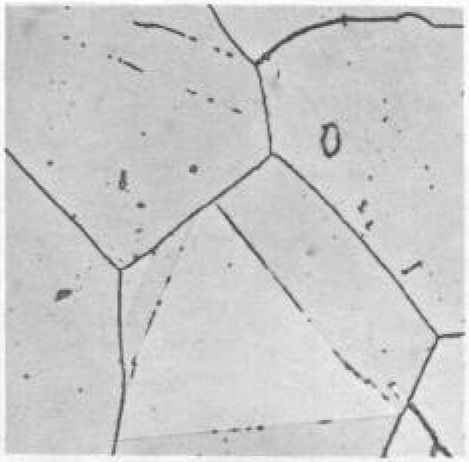
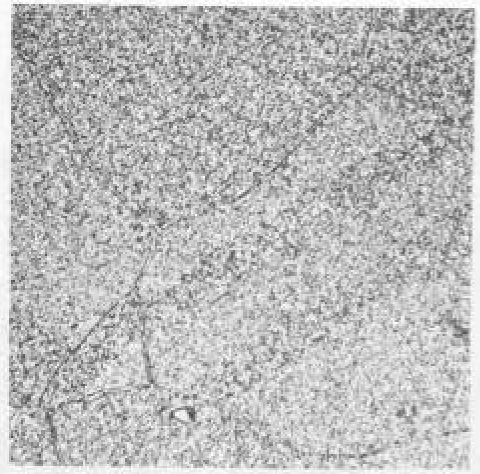


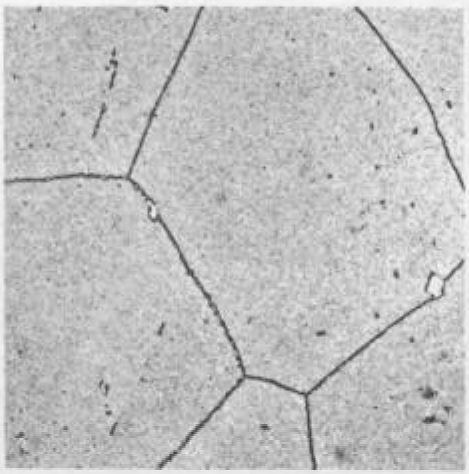
Fig. 16. Microstructures of Udimet 500 Aged for 100 Hours at Indicated Temperature after a Three Stage Heat Treatment. 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air, Magnification X 500



700°C

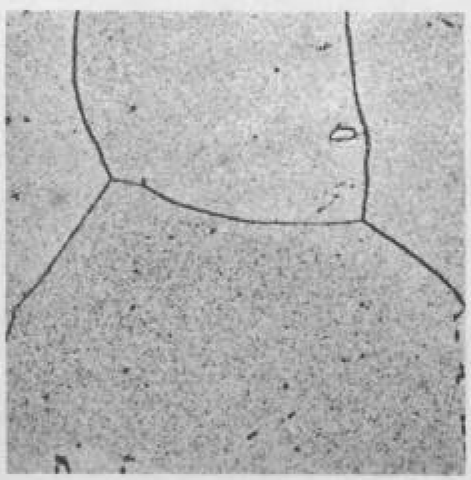


850°C

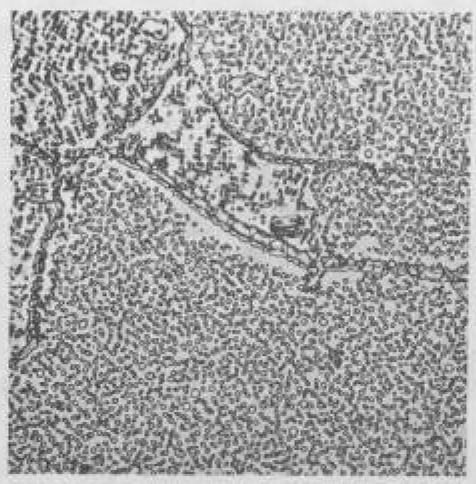


750°C

900°C



800°C



950°C

Fig. 17. Microstructures of Udimet 500 Aged for 1000 Hours at Indicated Temperature after a Three Stage Heat Treatment, 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air. Magnification X 500

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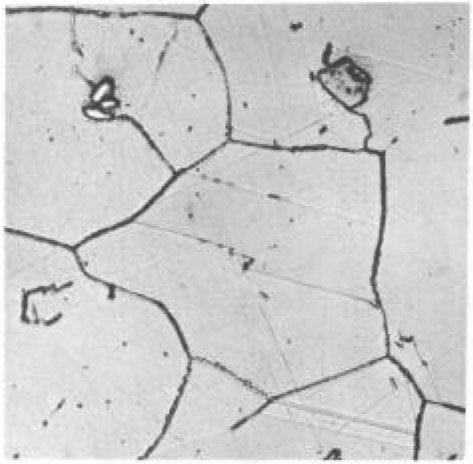
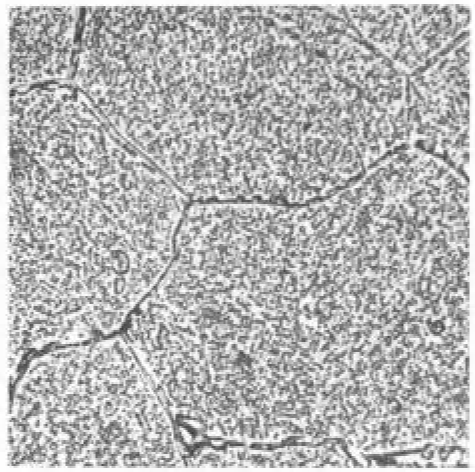
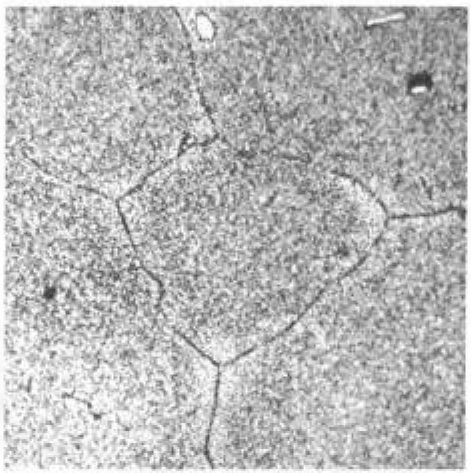


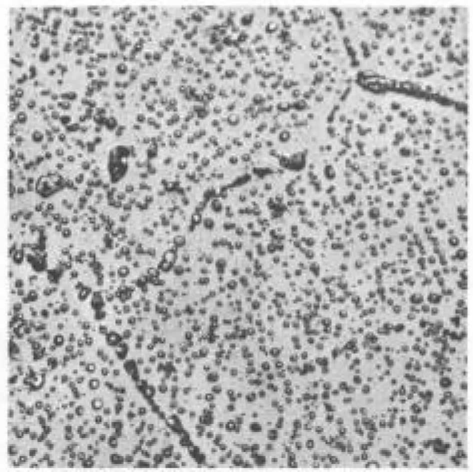
700°C



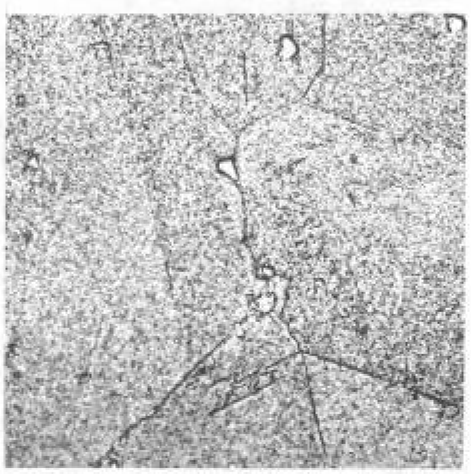
850°C



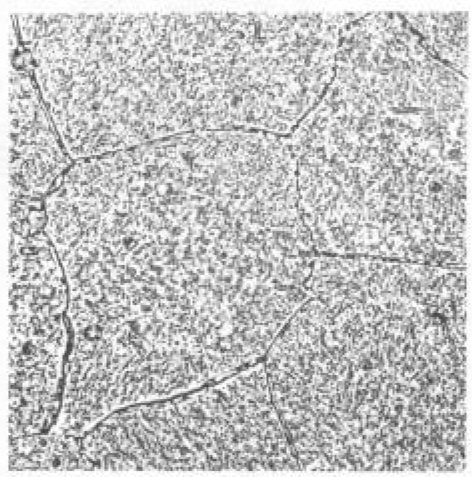
750°C



900°C



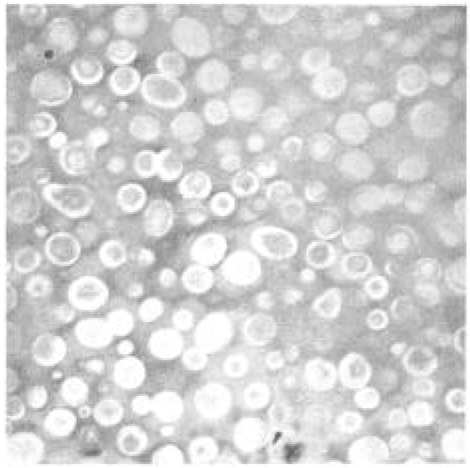
800°C



950°C

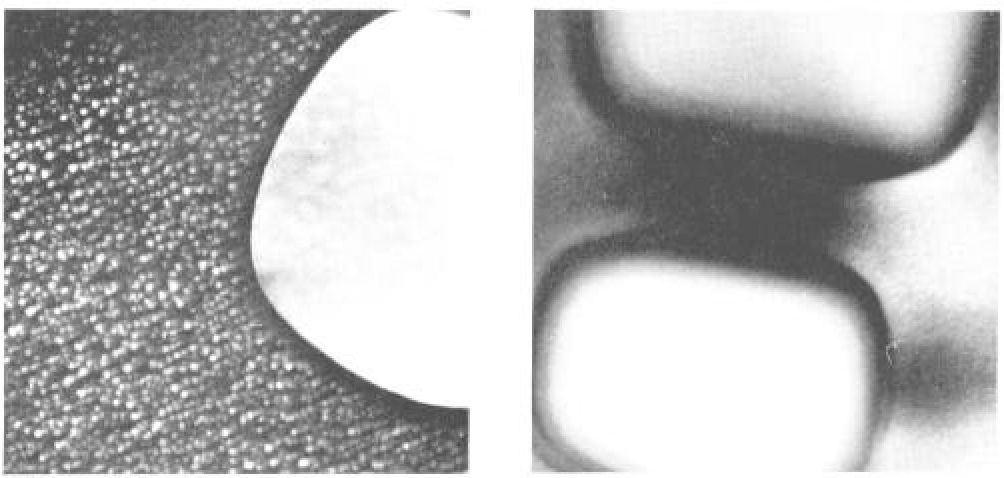
Fig. 18. Microstructures of Udimet 500 Aged for 5000 Hours at Indicated Temperature after a Three Stage Heat Treatment. 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air. Magnification X 500

H-72189



X 32.000

Fig. 19. Thin Foil Transmission Micrographs of Udimet 500 Aged for 5000 Hours at 700°C, after a Three Stage Heat Treatment



X 64.000 X 40.000

Fig. 20. Thin Foil Transmission Electron Micrographs of Udimet 500 Aged for up to 5000 Hours at 950°C, after a Three Stage Heat Treatment.

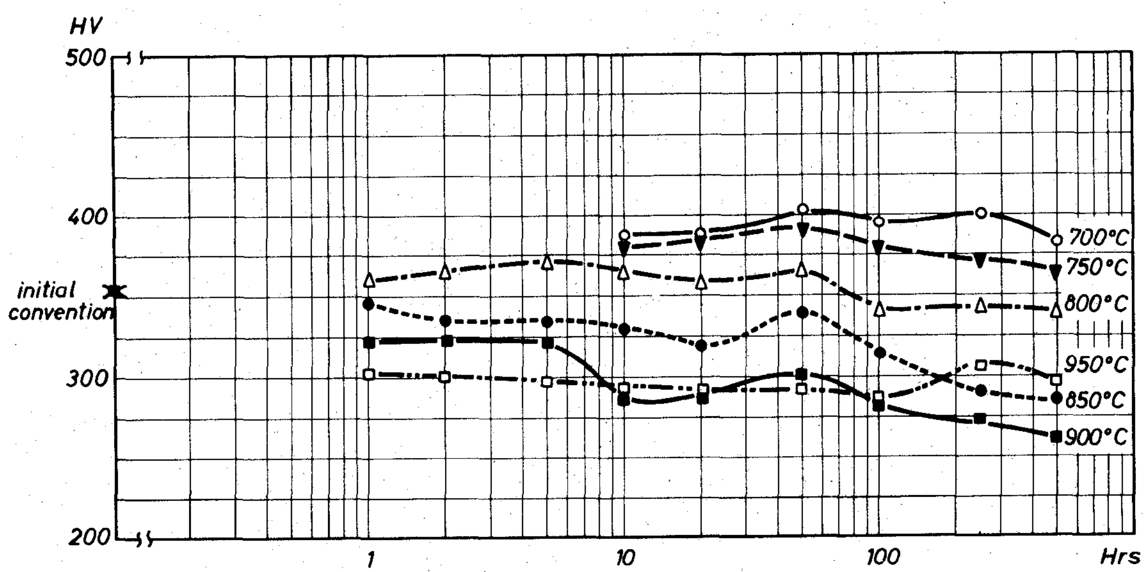


Fig. 21. Hardening Response of Udimet 500 Aged for up to 5000 Hours at 700-950°C Initial Heat Treatment. 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air.

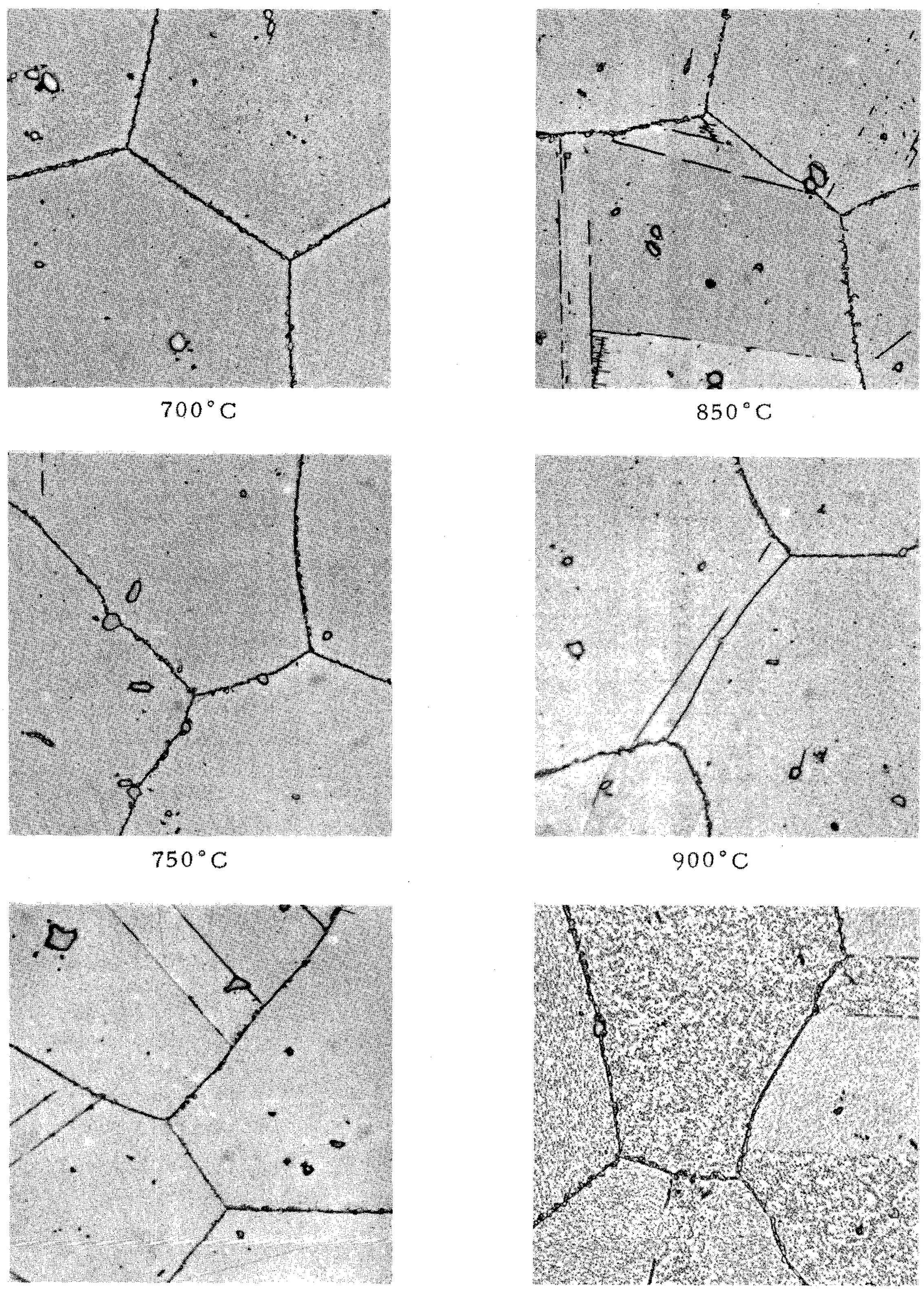


Fig. 22. Microstructures of Udimet 500 Aged for 100 Hours at Indicated Temperatures after a Four Stage Heat Treatment.
1175°C/2H/Air - 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air
Magnification X 500

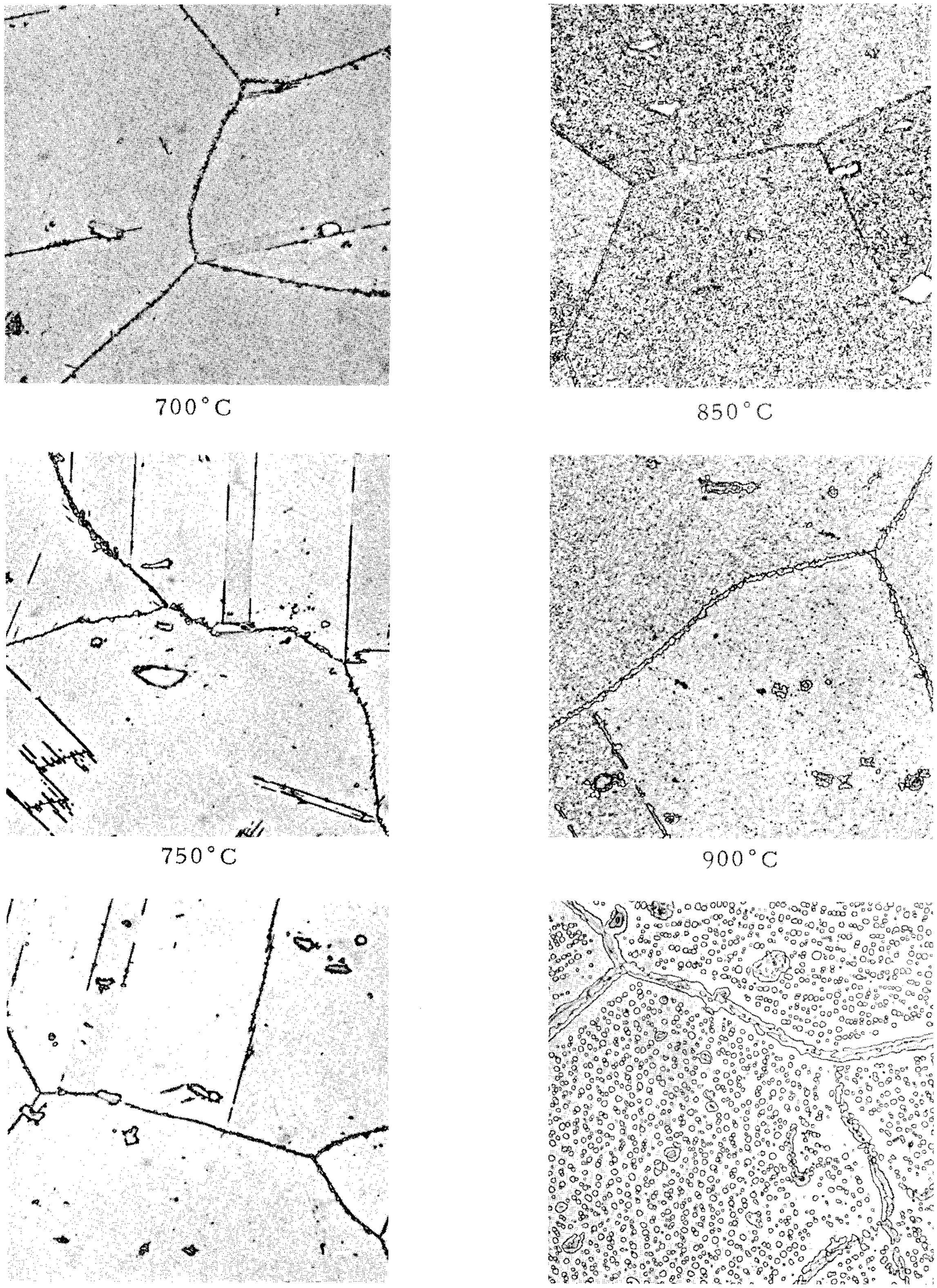
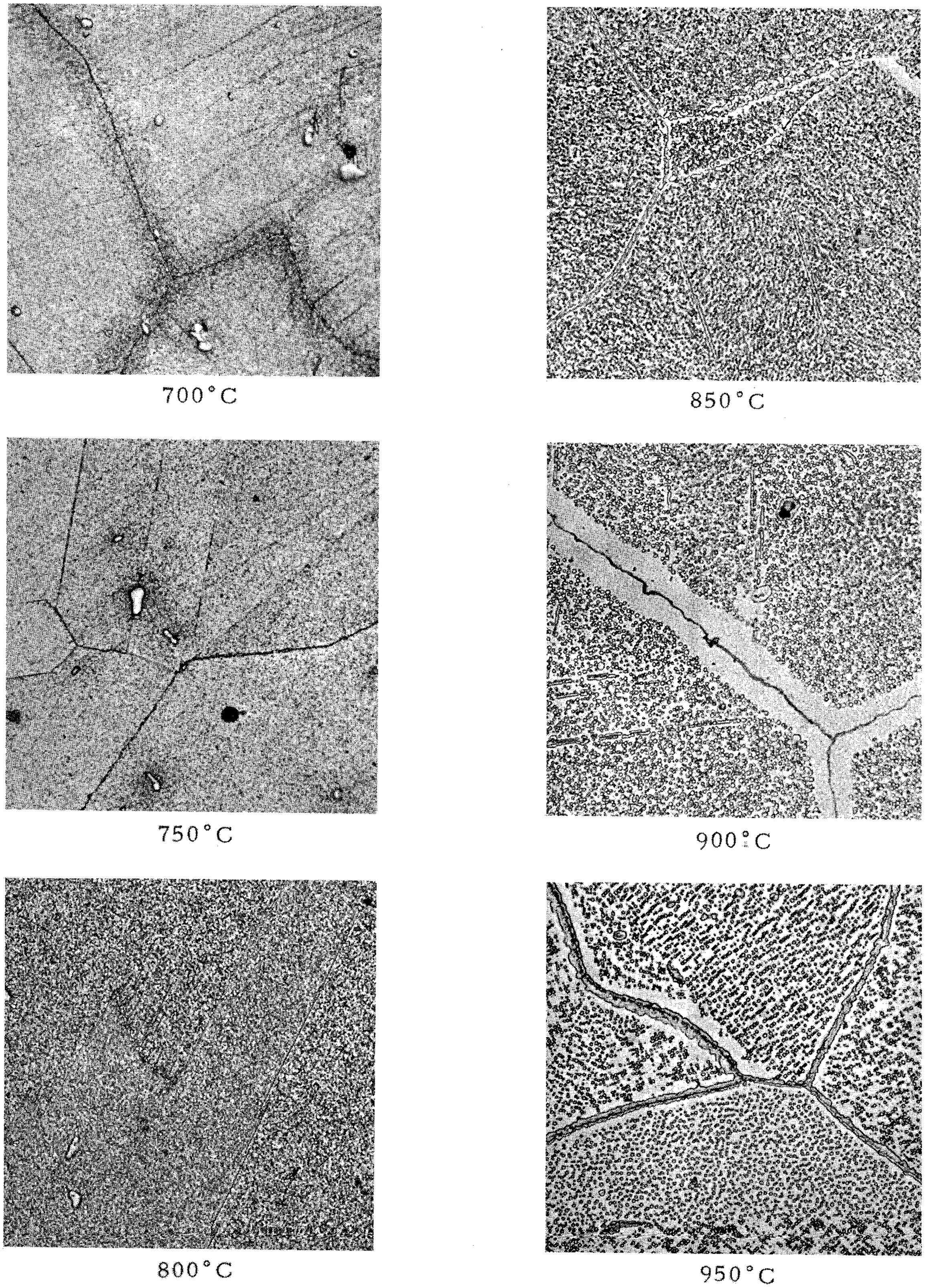


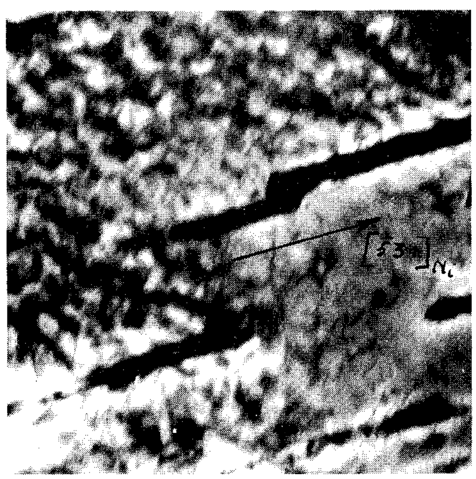
Fig. 23. Microstructures of Udimet 500 Aged for 1000 Hours at Indicated Temperatures after a Four Stage Heat Treatment. 1175°C/2H/Air - 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air Magnification X 500



H-22193

Fig. 24. Microstructures of Udimet 500 Aged for 5000 Hours at Indicated Temperatures after a Four Stage Heat Treatment. 1175°C/2H/Air - 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air Magnification X 500

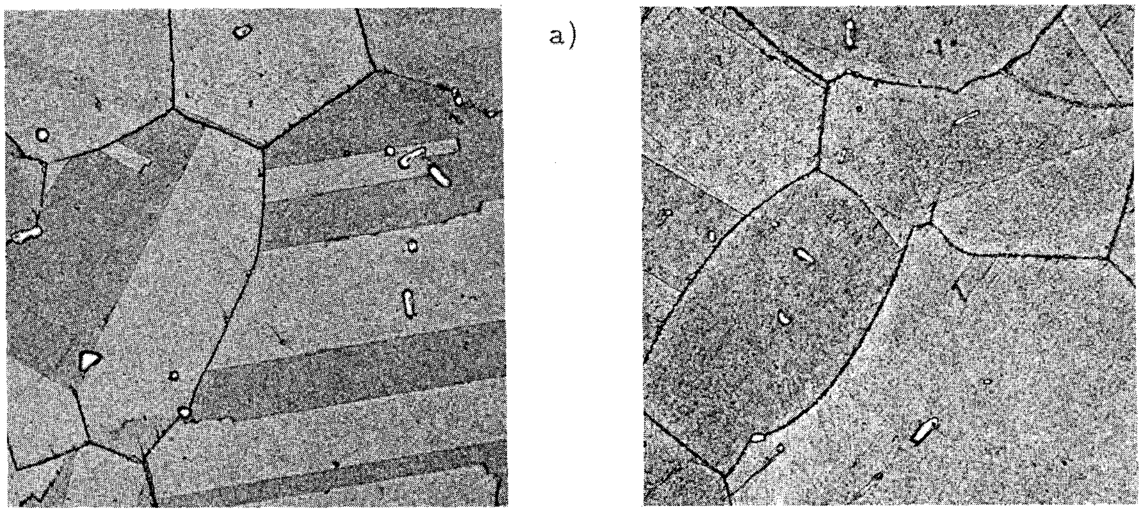
X 40,000



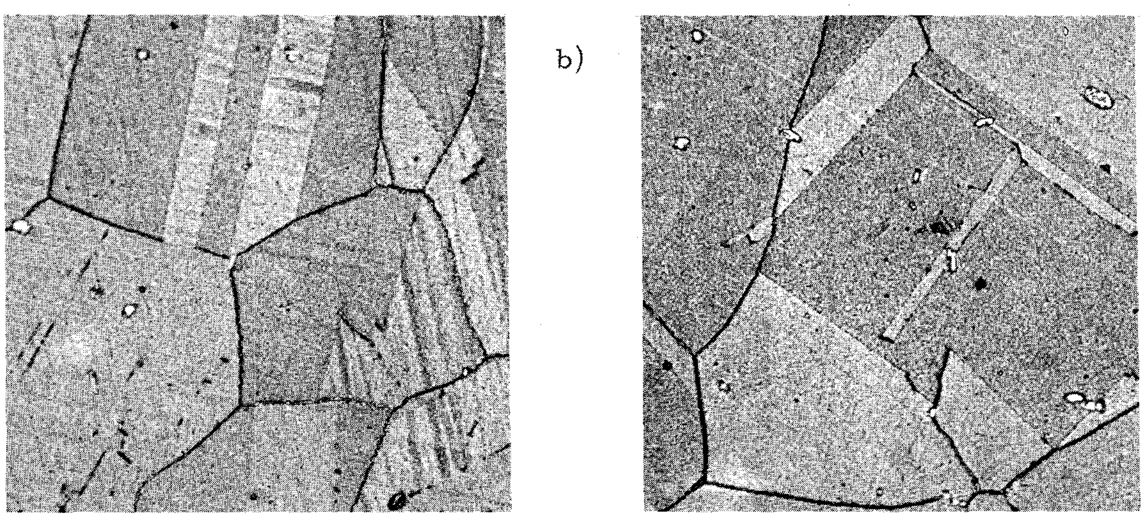
X 32,000

Fig. 25. Thin Foil Transmission Electron Micrographs of Udimet 500 Aged for 1000 Hours at 700°C Shown Intragranular and Grain Boundary $M_{23}C_6$ Carbides. Initial Treatment.
1175°C/2H/Air - 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air

a)

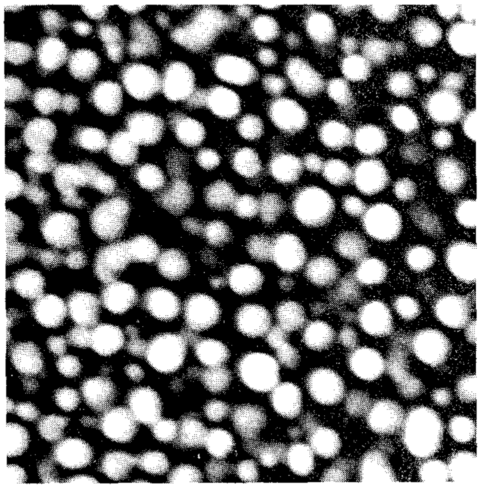


b)

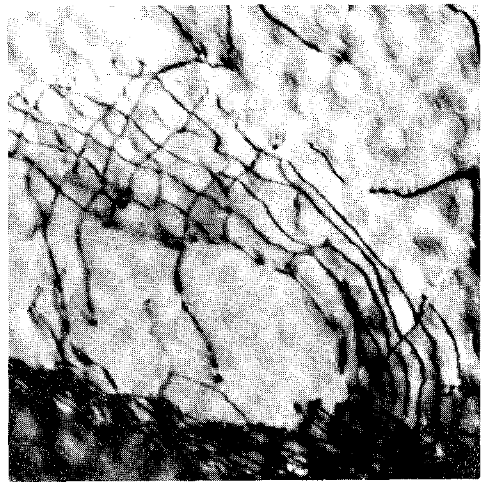


H-72195

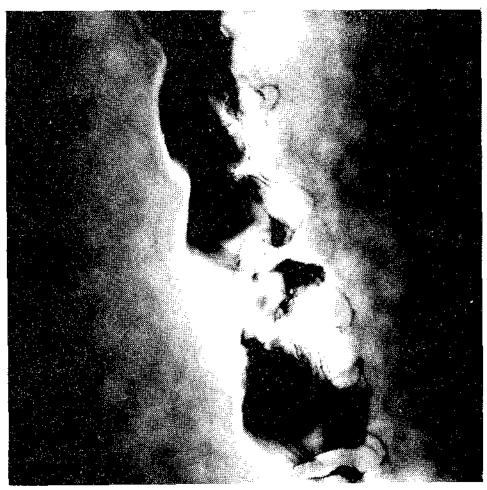
Fig. 26. Microstructures of Udimet 500 creep Tested :
a) at 700°C for 1000 Hours with a Stress of 7 Kg/mm2
b) at 700°C for 1000 Hours with a Stress of 35 Kg/mm2
Initial Heat Treatment. 1080°C/4H/Air - 840°C/24H/Air - 760°C/16H/Air. Magnification X 500



a)



b)



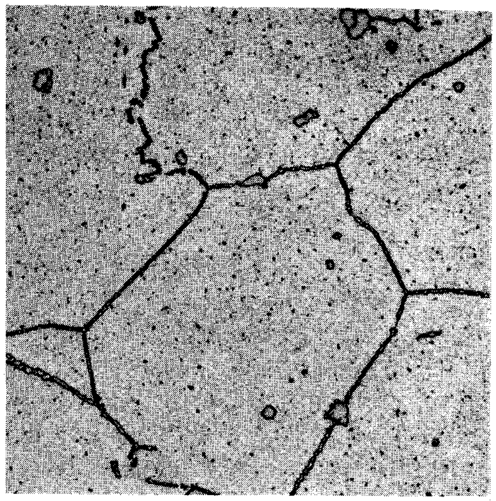
c)

Fig. 27. Thin Foil Transmission Electron Micrographs of Udimet 500 Aged for 1000 Hours at 700°C under a Stress of 7 Kg/mm2. Initial Three Stage Treatment.

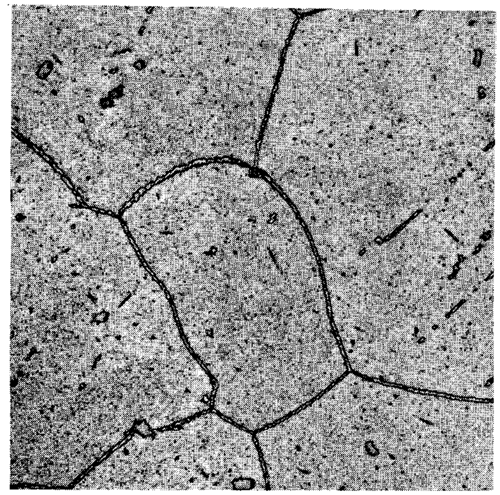
a) Virtually Unaffected γ' size

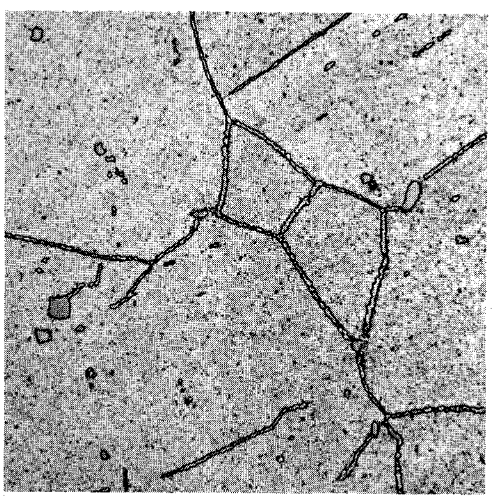
b) Dislocations Networks

c) Grain Boundary $M_{23}C_6$ Carbides. Magnification X 40.000

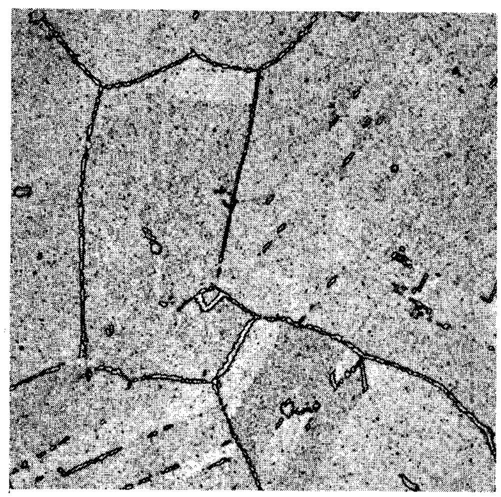


a)



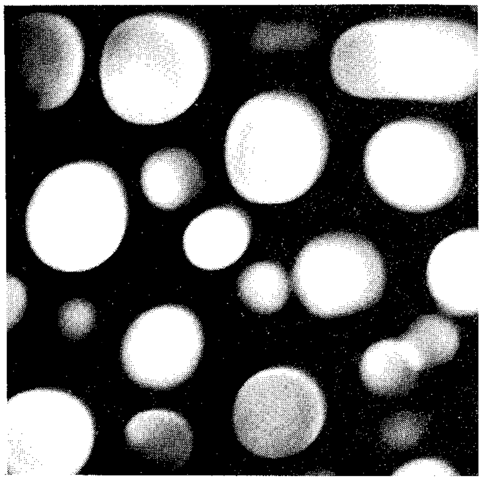


b)

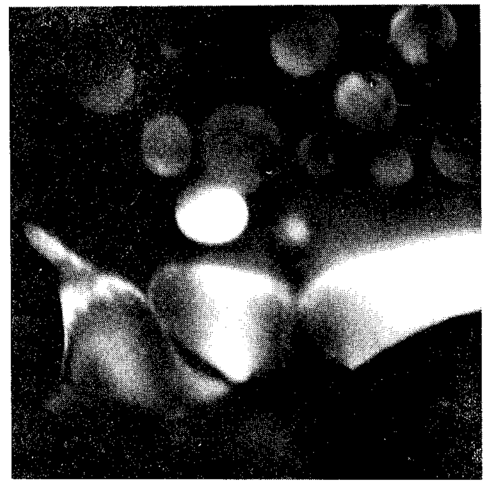


H-72/97

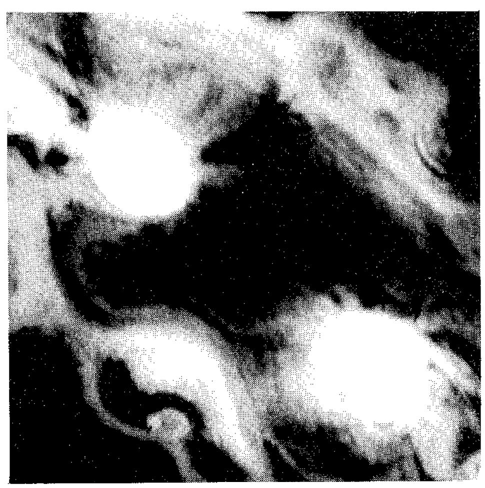
Fig. 28. Microstructures of Udimet 500 Creep Tested :
a) at 850°C for 1000 Hours with a Stress of 3 Kg/mm2
b) at 850°C for 1000 Hours with a Stress of 15 Kg/mm2
Magnification X 500



a)



b)



c)

Fig. 29. Thin Foil Transmission Electron Micrographs of Udimet 500 Aged for 1000 Hours at 850°C :
a) Under 3 Kg/mm2
b) Under 15 Kg/mm2
c) Initial Three Stage Treatment. Magnification X 40,000

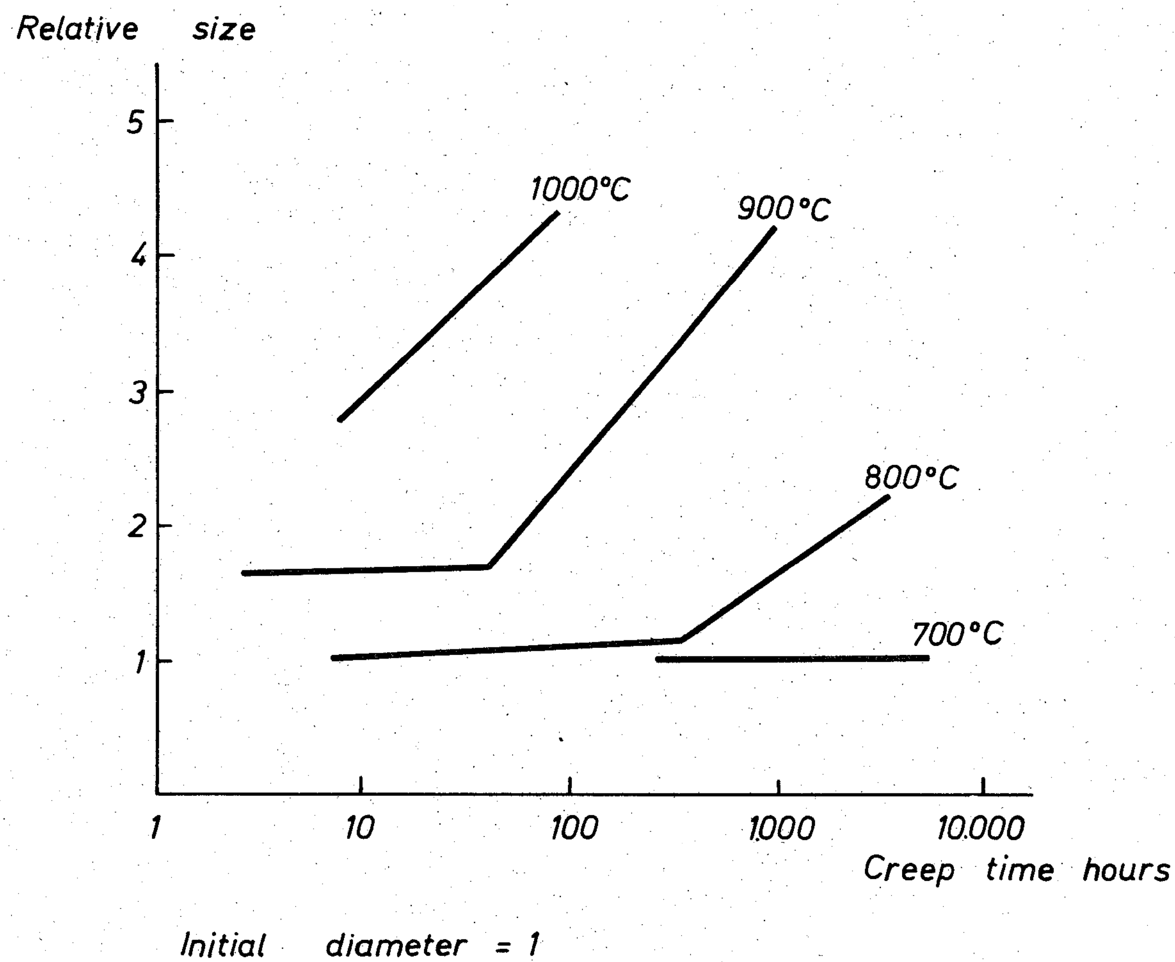
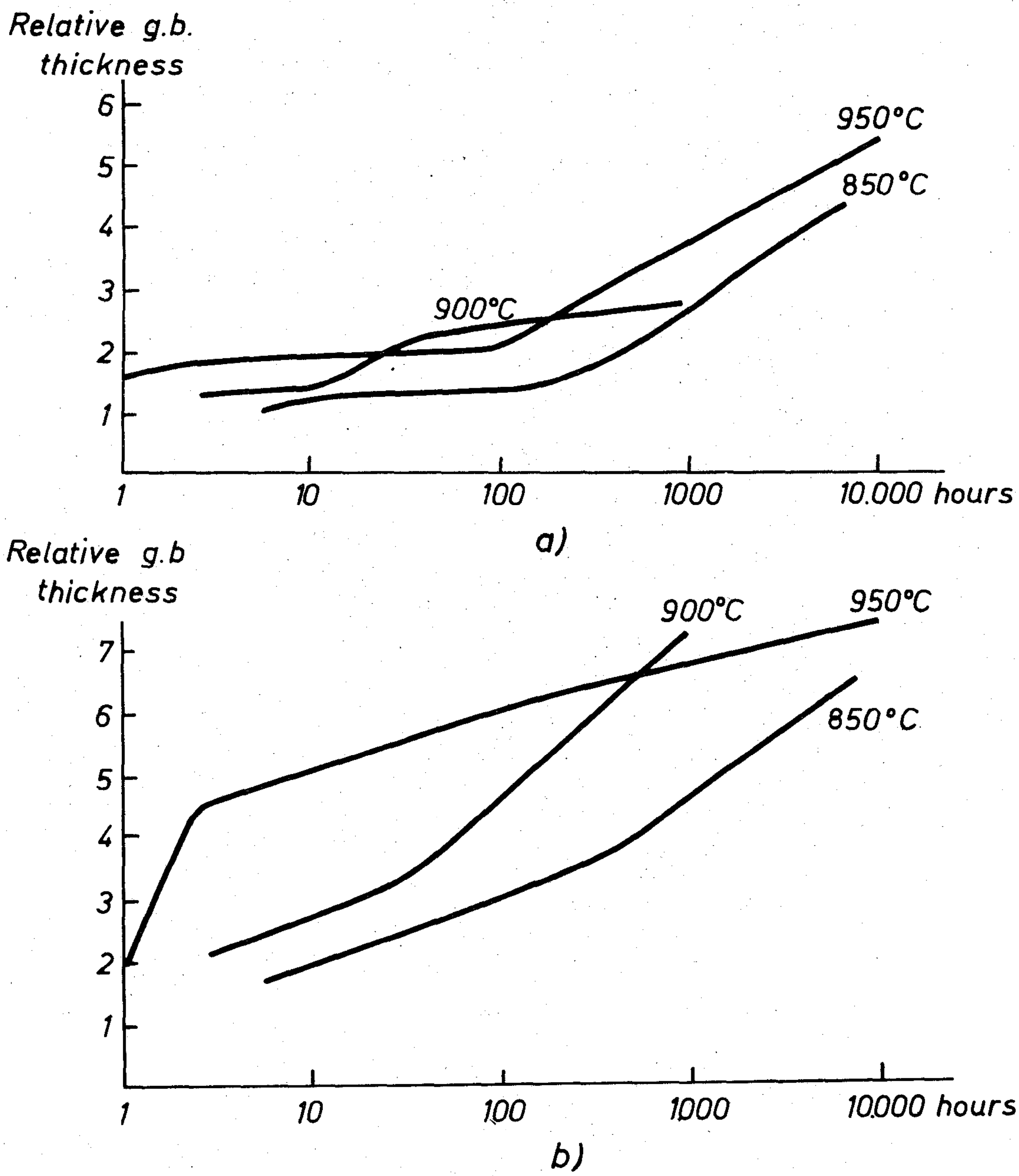


Fig. 30. Increase of γ' size of in Udimet 500 after Prolonged Creep Exposure. Initial Four Stage Treatment.



Thickness of initial grain boudary : 1

Fig. 31. Increase of Grain Boundary Thickness in Udimet 500 afer Prolonged Creep Exposure
a) in the Head of Creep Rupture Specimens.
b) in the Reduced Section of Creep Rupture Specimens
Initial Four Stage Heat Treatment.