TMP: IT'S EFFECT ON TURBINE HARDWARE

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ABSTRACT

The authors consider TMP as the deliberate mating of the deformation and heat treat processes to achieve specific mechanical property goals. The conception and birth of TMP is traced from the early 1950's involving the 16-25-6 and A-286 alloy turbine wheels through today's heavily TMP - dependent turbine components.

Major significance is associated with processing changes which occurred in Astroloy in the early 1960's. Both older and newer turbine alloys have been influenced by this revolution. The degree of change is quite evident in the mechanical property requirements for today's turbine hardware. The supremacy of tensile and stress-rupture strength is being challenged by creep, low cycle fatigue, fracture toughness and microstructure.

Examples are given of Inco 718, Waspaloy, Astroloy, and Rene' 95 in both solid and powder form.

DEFINITION

This session is devoted to thermomechanical processing. This is a loosely defined term and some of the subsequent speakers will describe research more "thermo" oriented than "mechanical". But in our review, we shall limit our remarks to the deliberate mating of deformation and heat treatment to achieve some mechanical property and/or microstructural goal that is unobtainable by heat treatment alone. In current U. S. technology, this goal is usually improved tensile and fatigue properties.

Optimization of the many TMP variables requires costly, time-consuming development. The major thrust has been directed at critical rotating components, i.e., compressor disks, shafts, turbine wheels. In these TMP cycles, the mechanical deformation is classified "warm" or "hot" work operation. Our attention will be directed toward this area. There is also an important field of "cold" deformation in sheet and bar TMP in which one speaker will report research.

Another thing we shall not do is discuss superplasticity. Available published research in this area utilizes a heat treat which eliminates the as-forged structure(1) and thus falls outside our definition of TMP.

HISTORY OF TMP USAGE

Thermomechanical processing of superalloys presumably originated with rolled rings of Timken 16-25-6 which were utilized by several engine manufacturers in welded turbine wheels. The potential of TMP was demonstrated but the reproducibility must have been a problem. This process was abandoned in the early 1950's with the advent of W545 and A286 precipitation hardening alloys. The metallurgist could analyze the chemistry, monitor the heat treatment, and ignore the deformation. Of course, we now know that these alloys may respond to TMP but the specification requirements of that time period were set low enough that the TMP need not be optimized.

Other alloys followed, i.e., Inco 901, M308, V57, D979. These followed the path of the A286. Perhaps it was their iron base that constricted our thinking about these "steels". Whatever the cause, the industry was immersed in nickel-base alloys before the benefits of TMP were rediscovered.

Udimet 700 arrived about 1958 as a wrought blade alloy, Figure 1. In 1960, this blade technology was applied to turbine wheels.(2) U700-Astroloy had a 2150° F. solution treat which was a complete γ' solution. It created a very coarse grain structure and effectively erased prior variations in processing. Once the alloy was heated to 2150° F., it was too brittle to quench. Some degree of overaging was inevitable as the turbine wheel air cooled after solution treat. A 1975° F. stabilize was required to spheroidize some of the carbides and borides dissolved at 2150° F. Unfortunately the 1975° F. stabilize continued the overaging. Even after subsequent low temperature aging, the tensile strength was low (although higher than other alloys of that time). In 1961, multiple vacuum melting was introduced and the ductilities significantly improved. However, the strengths achieved were unchanged.

The revolution began in 1963 with the issuance of a new Astroloy specification incorporating several remarkable features. First was the use of a 2050° F. "solution" which actually is within the upper portion of the γ' aging range. However, it was sufficiently close to the solution that most of the matrix was a solid solution. The lowered solution temperature permitted use of a quench and allowed the stabilize to be decreased from 1975° down to 1800° F. These reduced the overaging and raised the precipitation potential for the subsequent low temperature ages. The end result was a more uniform, finer grained structure (ASTM 3-4 instead of 00) with improved tensile strength and ductility and rupture ductility coupled with an acceptable decrease in static stress rupture life. This extensive development program has been partially reported.[3]

The crucial point was that the grain structure had to be established before the "solution" treat. This was implicit with the selection of a partial γ' solution as the specification heat treatment. The authors believe that this is the first time in U. S. superalloy technology where success depended upon the correct interlocking of deformation and heat treat practices. Since this initial breakthrough, TMP has received increasing attention in the evolution of jet engine rotor technology.

The evolution of TMP has not been an orderly development and many new obstacles have been encountered. Waspaloy serves as an example. It began life as a turbine blade alloy with a high γ' solution temperature of 1975° F. In 1964 experimental work by a material supplier had demonstrated that the R.T. yield strength could be varied at will from 120 to 175 ksi. as a function of forge practice combined with a lowered solution treatment.[4] This occurred at the expense of 1350° F. smooth bar stress rupture life, Figure 2. But as Figure 3 shows, no market existed for this new combination of properties. Before TMP, process development was the province of the producer. Since TMP, the engine manufacturers are required to fill a leadership role. The structure of our industry seems to be slowly changing to accommodate TMP.

PROBLEMS TO BE SOLVED IN THE DESIGN OF A TMP SEQUENCE

a. Temperature and Severity of Deformation

Temperature and severity are the initial basic questions. Subsequent speakers will be describing experimental work and we will not overlap their presentations. However, we wish to interject the measurement of "hot workability". Sellars and Tegart have published a marvelous analysis of this subject.[5] The normal objective is to measure some interrelation of strain to fracture, temperature, strain rate, etc. such as their Figure 17 and then select an appropriate peak ductility temperature (or other variable). With TMP it is still essential to obtain this information but the independent variable, i.e. temperature, is selected on the basis of its effect on properties, not on the basis of maximum achievable deformation.

One problem we note as suppliers of forgings is the inadequacy of describing the severity of work by "percent reduction". Perhaps this is sufficient for the rolling operation but it isn't for forgings. Equivalent true strain is a more unique value. It is more difficult to obtain in complex shapes but we would benefit by a lesser quantity of higher quality published research.

b. Strain Rate

Selection of strain rate influences radiation and conduction heat losses and the character of the adiabatic heating spike. We are seldom able to measure internal heating and are forced to deduce its potency by calculation or subsequent metallographic interpretation. Internal heating may be the major effect of strain rate on forgings where the die-chilled surface is removed by machining.

In the present state-of-the-art, the as-deformed microstructure is the measure from which we select strain rate. Capeletti, Jackman, and Childs recently summarized current knowledge about recovery and recrystallization in steels.[6] Luton and Sellars had studied the same phenomena in nickel and nickel-iron solid solutions[7]. With some combining and extrapolating, we may conclude that:

(1) dynamic recrystallization is the softening mechanism during hot-working.

(2) grain boundaries are preferred nucleation sites for recrystallization.

(3) the rate of recrystallization decreases as the temperature and/or the extent of deformation decreases.

(4) precipitation that may occur during the recrystallization can inhibit the softening process. Recrystallization can not be completed until the precipitate coarsens to a "relatively ineffective morphology".

Luton and Sellars further conclude that periodic recrystallization will result below a critical flow stress and continuous recrystallization above this critical value and that the recrystallized grain size will be uniquely determined by the flow stress and is not dependent upon the temperature of deformation. These last observations, perhaps true for solid solutions, have little applicability to the TMP of precipitation-hardening superalloys in the opinion of the current authors because of the overriding influence of the several coherent and noncoherent secondary phases.

Figure 4 shows the sequence observed in Inco 718. Comparison of Figures 4a and 4b show a refinement of grain size due to recrystallization. We are not dealing with recovery in these heavier strains. We cannot prove that recrystallization is dynamic but the observations agree with the theory that it is dynamic. Figure 4d illustrates the preferred nucleation along grain boundaries (and twin boundaries as well).

The rate of recrystallization decreases as the temperature decreases. The structure in Figure 4b was forged at an average upsetting rate of approximately 0.5 sec^{-1}. In Figure 17 of (5), it is calculated that a 0.3 sec^{-1} strain rate causes a 100-125 F° temperature rise. The metal remained above any precipitation temperature and recrystallization is essentially complete. In contrast, Figure 4d shows the structure after forging approximately 0.05 sec^{-1}. The total adiabatic heat was generated over a longer time span and the die chill was a greater factor. Consequently, recrystallization is incomplete. Figure 5 shows the electron microstructure in a fine grained recrystallized region. Due to the presence of carbide and/or Ni_3Cb particles at triple points, we infer that the temperature dropped during deformation and that the precipitate halted the progress of recrystallization.

Subsequent "solution" treatment at 1750° F. precipitates Ni_3Cb on all existing as-forged boundaries. Figure 4c shows no growth in the completely recrystallized fine-grained structure. Figure 4e shows relatively extensive growth restricted to the precipitate-free remnants of as-forged warm-worked grains. Thus the theories seem qualitatively correct.

Where do we stand quantitatively? Specifically, what is Capeletti et al's "relatively ineffective morphology"? Unfortunately, there is such a conglomeration of spheroids and films that the inhibition by noncoherent phases has no quantitative basis.(8) The closest relevant data known to the authors is that by Gladman, McIvor, and Pickering.(9) They report that (a) coarse particles accelerate recrystallization by supplying surfaces for nucleation,

(b) medium particles ($\sim$ 200 Å) don't affect recrystallization but may inhibit subsequent grain growth,

(c) fine particles ($\sim$ 50 Å) retard recrystallization when present in approximately 0.1 vol/o

> by stabliizing the subgrains and that the reduced nucleation rate may result in a relatively coarse recrystallized grain size.

Similar studies of the nickel-base system are needed.

In addition the particle size is not stable but may be continuously changing. Speight and Healey have examined this problem[10] but the theoretical approach is still much less complex than are the alloys in current usage.

The same lack of specific information limits precise knowledge of the effect of the coherent γ' phase upon recrystallization. Oblak and Owczarski[11] have made a tentative start on the study of cellular recrystallization but it is just a start. This will become a very important area if the industry moves toward direct forge and age.

We recognize that these differences in strain rate and the resultant differences in microstructure are important. In Inco 718, the stress rupture ductility is particularly sensitive, Table 1. In other alloys, it may be a different property or a different temperature range, etc. In the absence of understanding, we rely on empirically designed experiments.

c. <u>Phase Equilibria</u>

In the initial stages of TMP development, control of the boundary carbides was the prime objective and Rene' 41 is an example. Phase studies had shown that the γ' was dissolved above 1925° F. and the M_6C above 2050-2100° F. Figure 6a shows a ring rolled with much of the carbon in solution. During subsequent 1950° F. solution treatment, a thick layer of M_6C precipitated at grain and twin boundaries. Figure 6b shows a ring rolled at 1975° F. Due to reduced carbon in solution, the heat treated structure has clean twins and much decreased boundary precipitation. The ductility was vastly improved, Table 2.

Control of the carbide equilibria has since progressed furtherest in Waspaloy. Figure 7 shows the different $M_{23}C_6$ morphology in two forgings from one bar. The intercellular structure was obtained by use of a 1900° F. forge temperature where the spherical structure resulted from an 1875° F. temperature. But both structures could be obtained in an 1875° F. forging by a 25 F° difference in solution treat temperature. It is assumed that this variation is due to the amount of carbon in solution at the beginning of the 1550° F. stabilize cycle. This variation is too minute to verify by analysis. Actual practice is built on a hypothesis which rests on a foundation of quicksand. We must remain alert to new knowledge of phase equilibria resulting from new equipment, better experiment design, etc.

Kinetic data on the γ' equilibria is available on only one alloy, Astroloy/U700.[12] This is thought to be a sensitive function of chemistry[13] and is needed for every different alloy for intelligent selection of the range of heating temperatures and variation in permissible time at heat.

d. <u>Introduction of In-Process Anneals</u>

High temperature cycles have been widely employed in TMP cycles for a variety of reasons, i.e., more uniform γ' solvus, dissolve brittle intermetallics, control of in-process grain size. The industry has generally been able to ignore critical grain growth because sufficient carbides were present to limit the maximum size.

Powder products will force us to relearn the critical growth concept. High temperature anneals are tempting because they minimize the original powder boundary influence. Figure 8 shows two IN-100 compacts forged at 2150° F., one with a 2250° F. intermediate anneal and one without. The critical growth was disastrous to the tensile ductility. In the manufacture of contour shapes, some nonuniformity of strains will be inevitable and TMP development must build in some flexibility to accommodate this need.

e. Reproducibility of Manufacturing Practices

After specifying a TMP cycle, can the manufacturing process be adequately controlled? Table 3 lists data collected on one run of ten Waspaloy forgings in an 18,000 ton hydraulic press.

The heating time was much more closely controlled in the finish than in the upset. Transfer and waiting times had similar averages in finish and upset but the sigma is relatively large in both. The finish deformation cycle is much slower.

The factor of interest is the small sigma of the deformation cycle as compared to the handling cycle. The machine is much more reproducible than are the people. Semi-automated manufacturing processes can more fully achieve the total potential offered by TMP.

Heating facilities are another limiting factor in TMP control. Rene' 95 has an in-process anneal designed to control one of the two segments of the aim duplex structure. The aim is represented by the 2075° F. structure, Figures 9c and d. Note that a ± 15 F° variation results in significantly different structures. If one is working from a furnace with the usual ± 25 F° certification, the TMP cycle obviously will not be reproduced.

Another factor important in the design of TMP cycles is the economic penalty associated with mistakes. Table 4 shows the loss in ultimate strength that occurs in the stronger segment (the hub) of a Rene' 95 forging due to a restrike. The damage is not visible by optical microscopy but is revealed only by the destructive testing. TMP places a heavier burden on the mutual trust and integrity of relations between consumer and producer.

THEORY OF TMP

Oblak and Owczarski[14] have published the first rudiments of a coherent theory. Their extensive work on U700 demonstrates the information necessary to make an initial, relatively gross selection of an optimum cycle. These are:

(a) an initial structure described in detail,

(b) an investigation of the optimum aging cycle for each deformation cycle,

(c) a broad evaluation including tensile, rupture, creep, LCF, some high cycle fatigue, texture studies, and metallography.

The authors found that this stable γ' did not require an optimized age cycle for each deformation cycle. This agrees with Nutting's statement[15] and is contrast to the behavior of metastable γ'[15, 16].

Oblak and Owczarski conclude that a uniform array of γ' must first be produced. Then the warm work must be conducted below the recrystallization temperature but above the range in which planar slip predominates. They define the optimum stable structure as characterized by a polygonal cell alignment of the dislocations with the optimum cell size control resulting from the distribution of the γ' precipitate.

This constitutes a theory which can be measured and tested. To the current authors, it seems unlikely to be the whole theory because it doesn't cover the partially recrystallized structure desirable for the Rene' 95 chemistry. But it is a vast improvement upon the amorphous mass existing prior to this publication.

SUMMARY

We believe that the need for TMP superalloys is self-evident. The current market is limited to certain high-integrity components because the need for shotgun-style development causes expensive pilot-plant development. Rudiments of TMP theory are beginning to evolve. As the principles emerge, the reliability of manufacture will increase and the cost of development decrease. These trends forecast a broadening demand for TMP in elevated-temperature service applications. The trend appears inevitable but the rate depends upon our skills as scientists and engineers.

REFERENCES

1 M. M. Allen, R. L. Athey and J. B. Moore, Application of Powder Metallurgy to Superalloy Forgings, MEQ, February 1970, pp 20 - 30.

2 Various specifications of the Pratt & Whitney Aircraft Company.

3 R. L. Athey and J. B. Moore, Development of Astroloy Disks for Advanced Jet Engine Application, FRDC, PWA publication GP-67-21 of 5 May 1967 orally presented at AIME High Temperature Alloys Symposium in Los Angeles on 20 February 1967.

4 W. H. Couts, Jr. and J. F. Radavich, Correlation of Forge Temperature and Phase Equilibria during Processing of a Nickel-Base Superalloy, orally presented at AIME High Temperature Alloys Symposium in New York on 19 February 1964.

5 C. M. Sellars and W. J. M. Tegart, Hot Workability, International Metallurgical Reviews, 17, 1972, pp 1 - 24.

6 T. L. Capeletti, L. A. Jackman and W. J. Childs, Recrystallization Following Hot-Working of a High-Strength Low-Alloy (HSLA) Steel and a 304 Stainless Steel at the Temperature of Deformation, Metallurgical Transactions, 3 (1972), pp 789 - 796.

7 M. J. Luton and C. M. Sellars, Dynamic Recrystallization in Nickel and Nickel-Iron Alloys During High Temperature Deformation, Acta Metallurgica, 17, August 1969, pp 1033 - 1043.

8 M. J. Donachie, A. A. Pinkowish, W. P. Danesi, J. F. Radavich and W. H. Couts, Jr., Effect of Hot Work on the Properties of Waspaloy, Metallurgical Transactions, 1, September 1970, pp 2623 - 2630.

9 T. Gladman, I. D. McIvor and F. B. Pickering, Effect of Carbide and Nitride Particles on the Recrystallization of Ferrite, JISI, 209, May 1971, pp 380 - 390.

10 M. V. Speight and T. Healey, Effects of Temperature and Alloy Composition on the coarsening of Metal Carbide Particles, Metal Science Journal, 6, 1972, pp 11 - 12.

11 J. M. Oblak and W. A. Owczarski, Cellular Recrystallization in a Nickel-Base Superalloy, Transactions of the Metallurgical Society of AIME, 242, August 1968, pp 1563 - 1568.

12 E. H. VanDerMolen, J. M. Oblak and O. H. Kriege, Control of γ' Particle Size and Volume Fraction in the High Temperature Superalloy Udimet 700, Metallurgical Transactions, 2, June 1971, pp 1627 - 1633.

13 J. E. Bridge, Jr. and G. N. Maniar, Effects of Molybdenum on the Diffusion of Aluminum and Titanium from Ni_3 (Ti, Al) Intermetallic Compounds into a Ni-Mo Matrix, Metallurgical Transactions, 3, April 1972, pp 1005 - 1007.

14 J. M. Oblak and W. A. Owczarski, Thermomechanical Strengthening of a γ' Precipitation-Hardened Nickel-Base Alloy, Metallurgical Transactions, 3, March 1972, pp 617 - 626.

15 J. Nutting, The Influence of Plastic Strain Upon the Aging Characteristics of Alloys, Metallurgical Transations, 2, January 1971, p 48.

16 J. M. Oblak, W. A. Owczarski, and B. H. Kear, Heterogeneous Precipitation of Metastable γ'-Ni_3Ti in a Nickel-Base Alloy, Acta Metallurgica, 19, April 1971, pp 355 - 363.

TABLE 1

EFFECT OF STRAIN RATE ON INCO 718*

Furnace Temperature From Which Forged	Average Forge Closure	% Elongation in 1200° F. Stress Rupture
2025° F.	.5 sec^{-1}	6%
2025° F.	.05 sec^{-1}	25%
1825° F.	.5 sec^{-1}	15%
1825° F.	.05 sec^{-1}	25%

* These were pancake forgings, some of which are portrayed in Figure 4.

TABLE 2

COMPARISON OF ROLLING TEMPERATURES FOR RENE' 41

a. R. T. Tensile

Roll Temperature °F.	.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, %	Reduction of Area, %
2050	136.0	150.1	3.0	5.4
1975	138.5	203.1	21.0	21.1

b. 1400° F. Tensile

Roll Temperature °F.	.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, %	Reduction of Area, %
2050	117.3	160.1	8.0	10.9
1975	118.3	148.1	20.0	31.5

TABLE 3

SCATTER OF MANUFACTURING RESULTS

A. Upset Operation

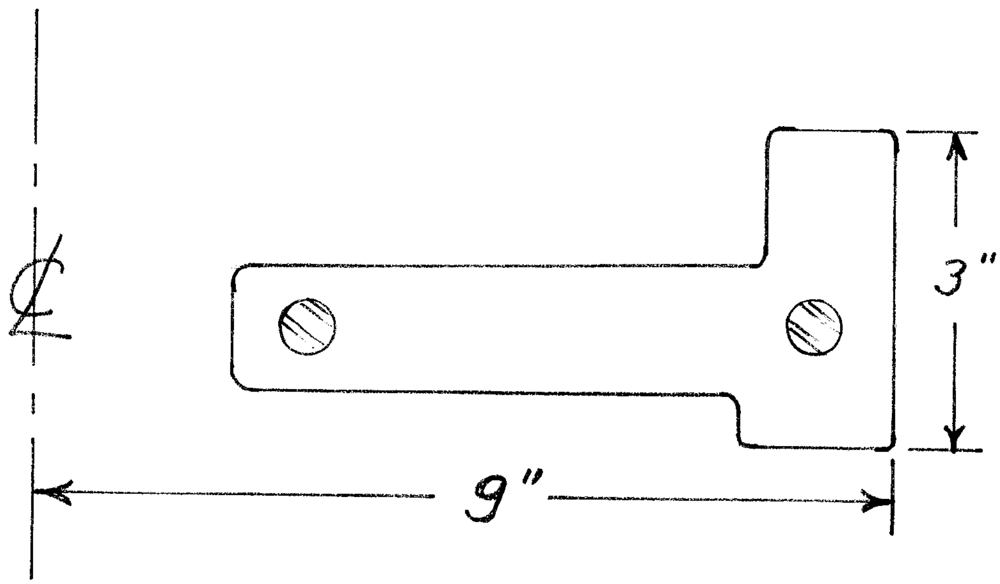
	Average	Sigma
Heating time (minutes)	116.	21.8
Transfer time, furnace to die (seconds)	28.	4.9
Time lapse awaiting top die contact (seconds)	15.	9.7
Duration of deformation cycle (seconds)	8.6	1.9

B. Finish Operation

	Average	Sigma
Heating time (minutes)	95.	3.2
Transfer time, furnace to die (seconds)	29.	3.0
Time lapse awaiting top die contact (seconds)	16.	5.7
Duration of deformation cycle (seconds)	19.	3.1

TABLE 4

EFFECT OF RE-STRIKING RENE' 95



a. R. T. Tensiles

Location	Restrike	.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, %	Reduction of Area, %
Hub tangential	No	190.	234.0	13.0	15.2
" "	Yes	193.6	224.8	11.5	13.7
Rim tangential	No	178.4	228.8	11.0	12.2
" "	Yes	177.8	224.6	13.5	13.7

b. 1200° F. Tensiles

Location	Restrike	.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, %	Reduction of Area, %
Hub tangential	No	174.0	223.0	15.0	18.1
" "	Yes	175.4	213.2	10.0	10.8
Rim tangential	No	161.6	216.6	10.0	15.2
" "	Yes	166.0	213.0	12.0	14.4

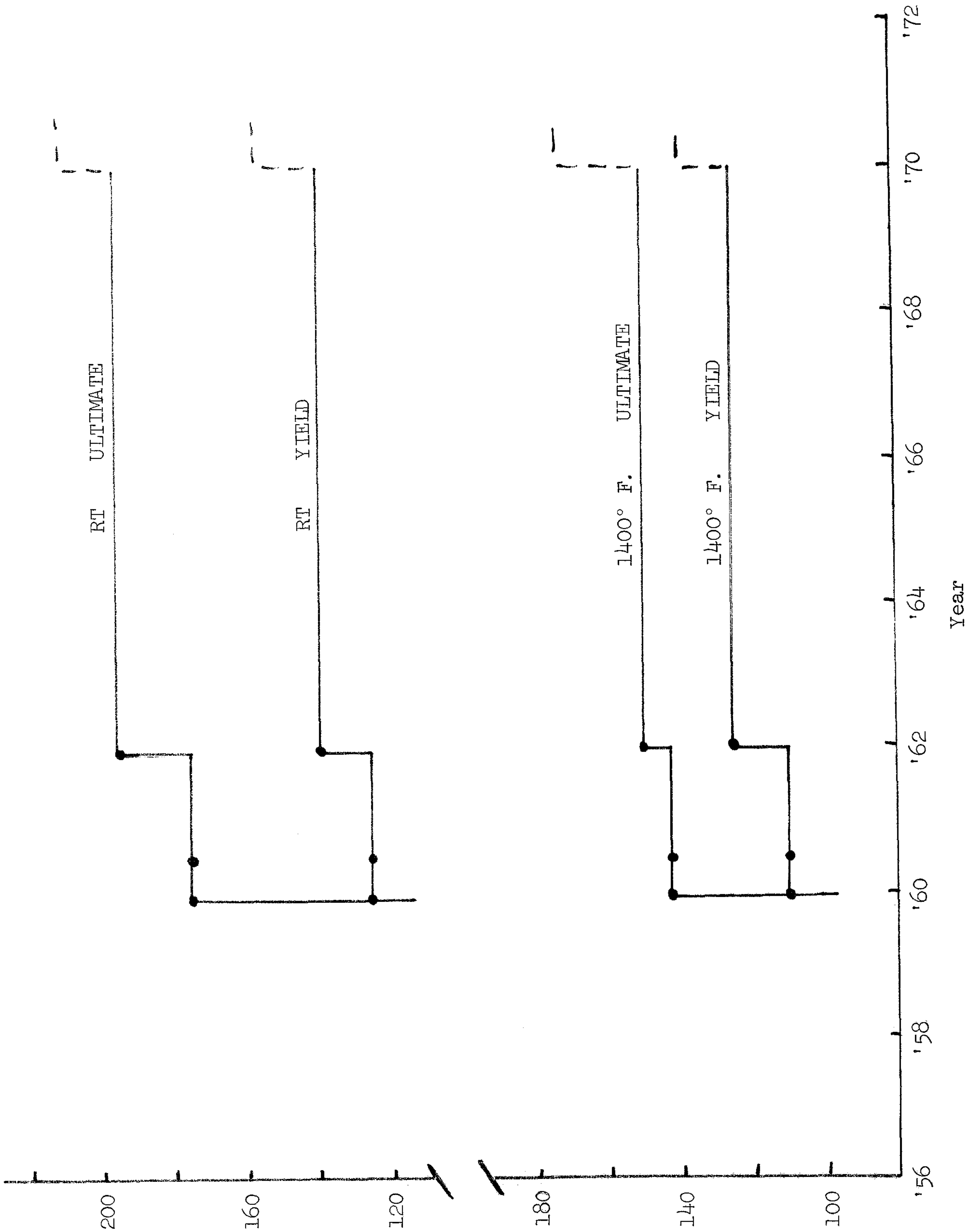


Figure 1. Required Tensile Strength of Astroloy Hardware

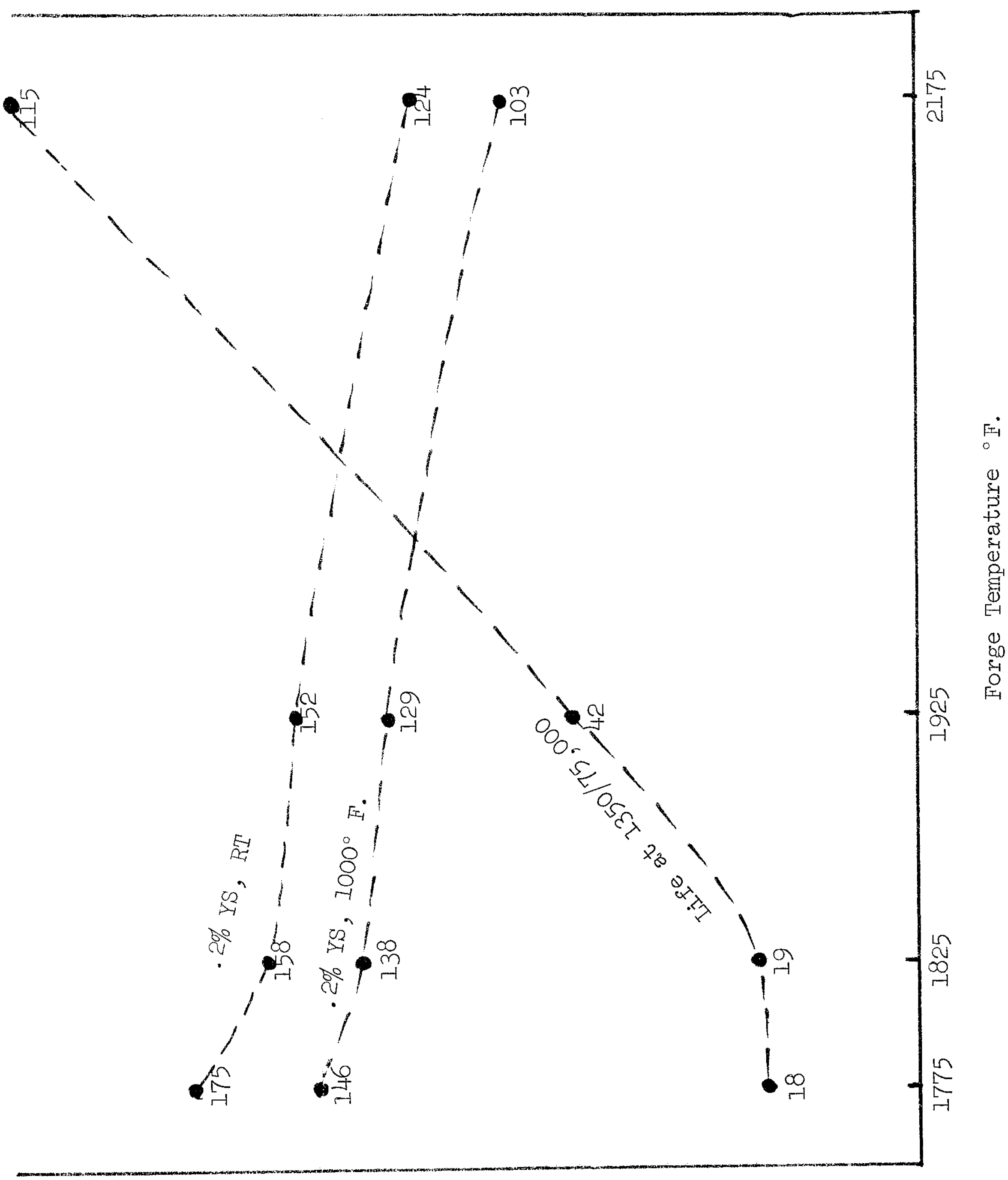


Figure 2. .2% Yield Strength, Ksi
Stress Rupture Life, Hours

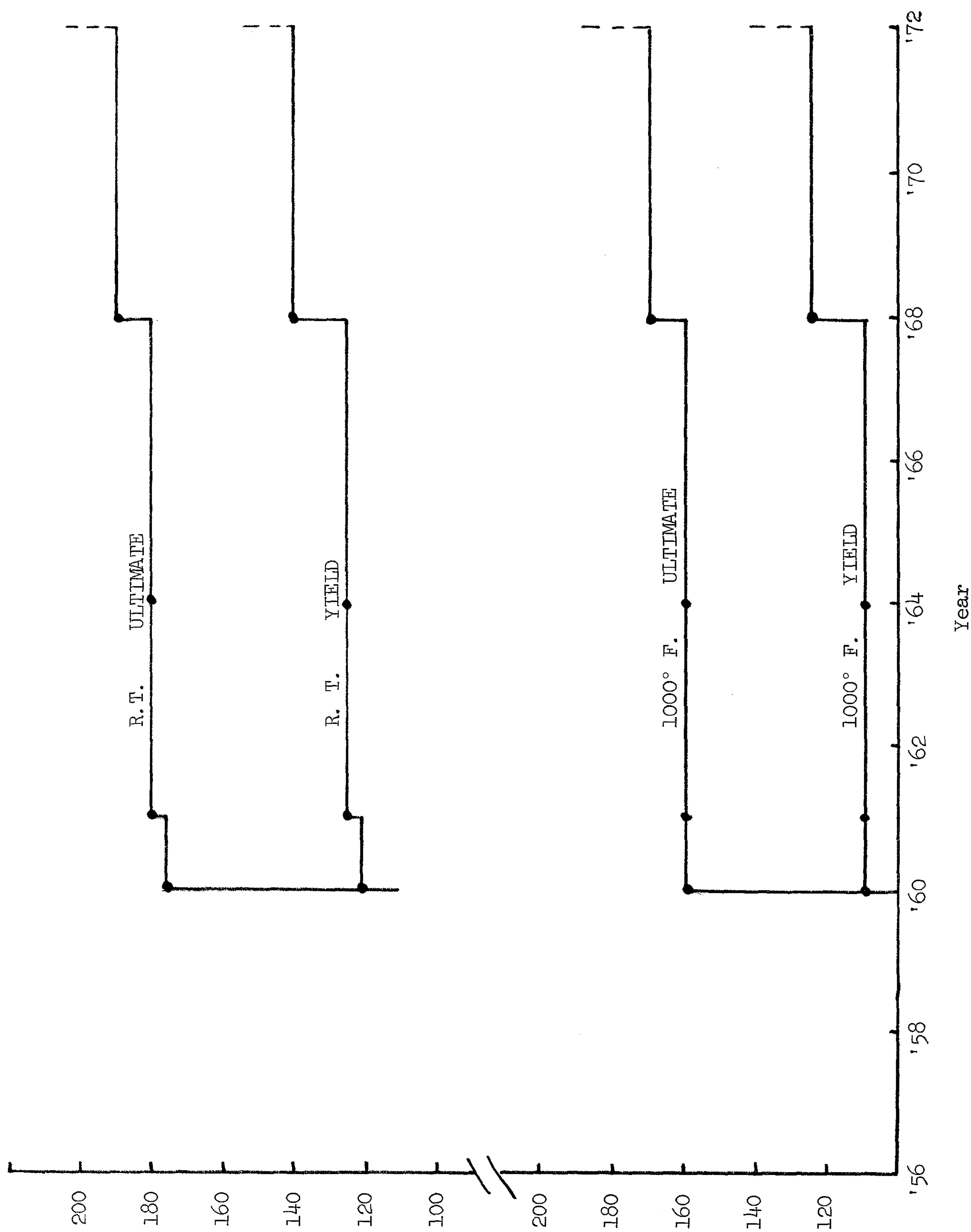
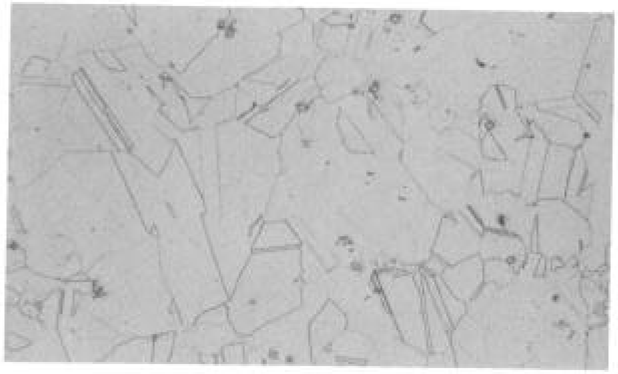
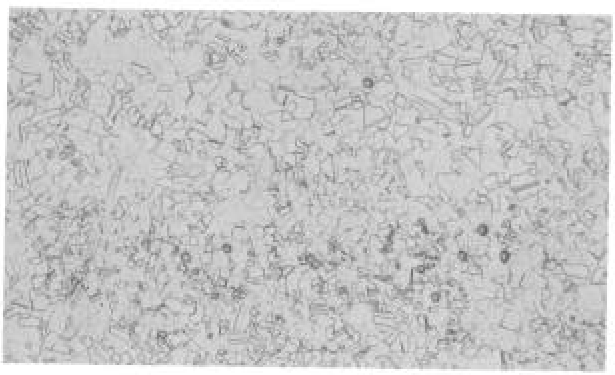


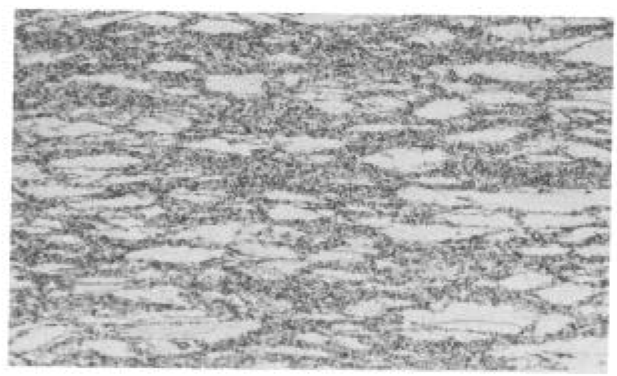
Figure 3. Required Tensile Strength of Waspaloy Hardware



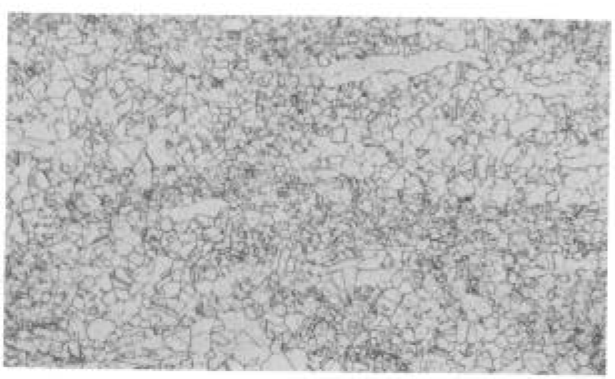
a. Heated 2 Hrs. at 1825° F. 100X
Tm $\sim$ .85-.90



b. High Strain Rate 100X



d. Slow Strain Rate 100X



c. Plus 1750° Solution 100X
15% Rupture Elongation

e. Plus 1750° Solution 100X
25% Rupture Elongation

Figure 4. Inco 718 Microstructures demonstrating the effect of adiabatic heating upon recrystallization.

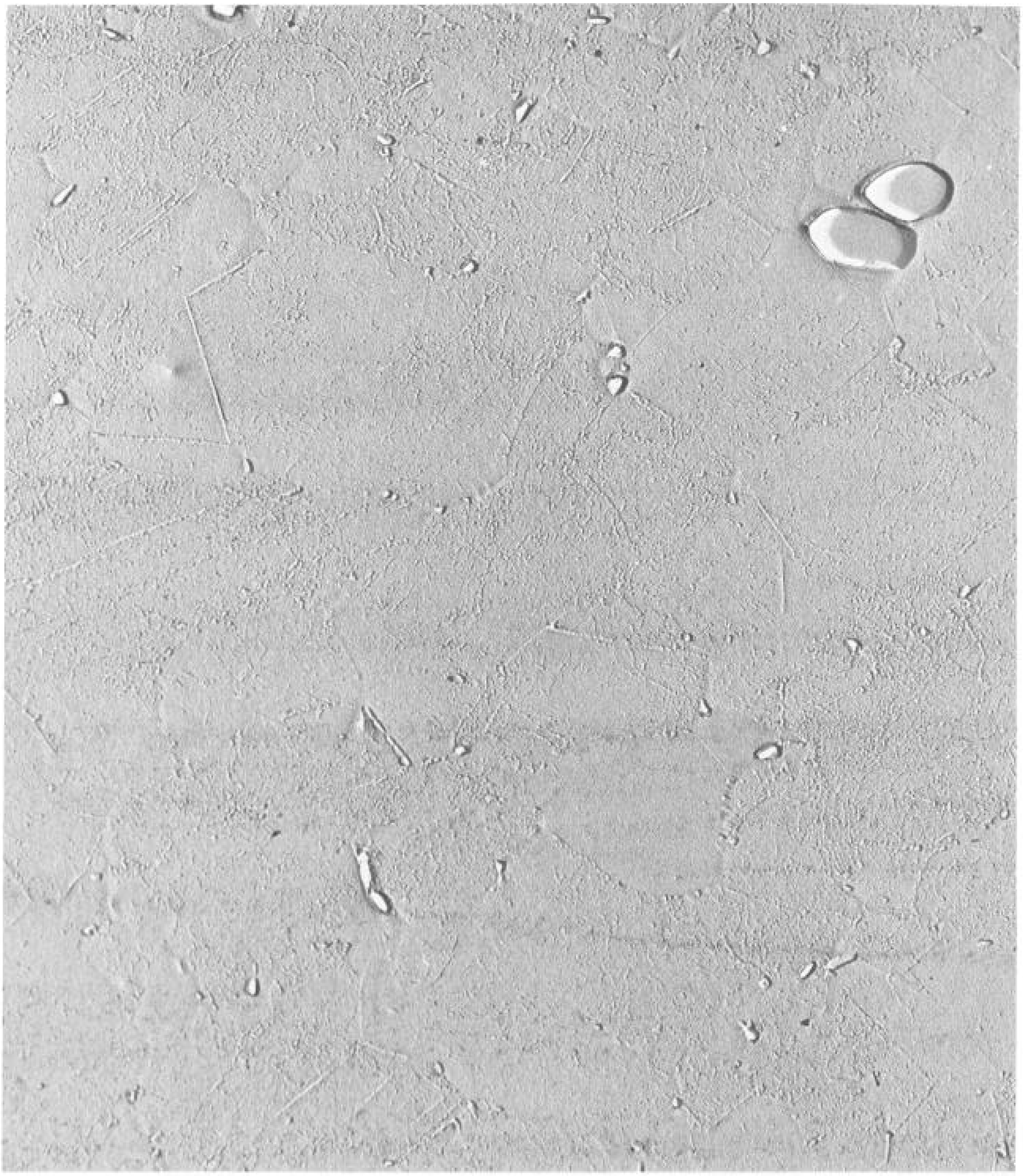
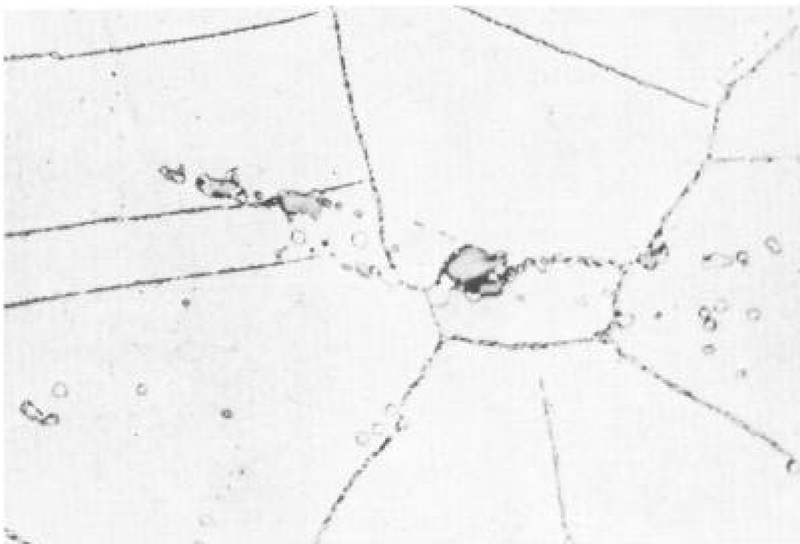
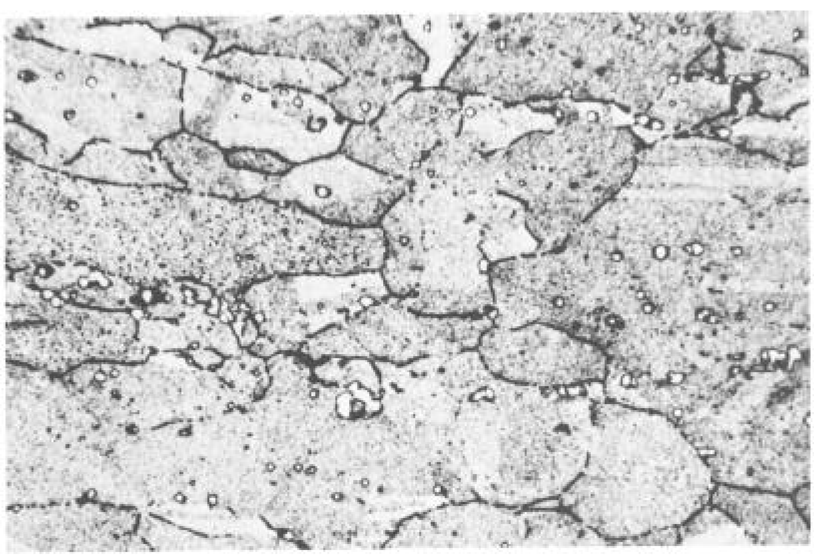


Figure 5. Electron Microstructure of Inco 718 as-forged at a slow rate from a 1825° F. initial temperature. 25,000X

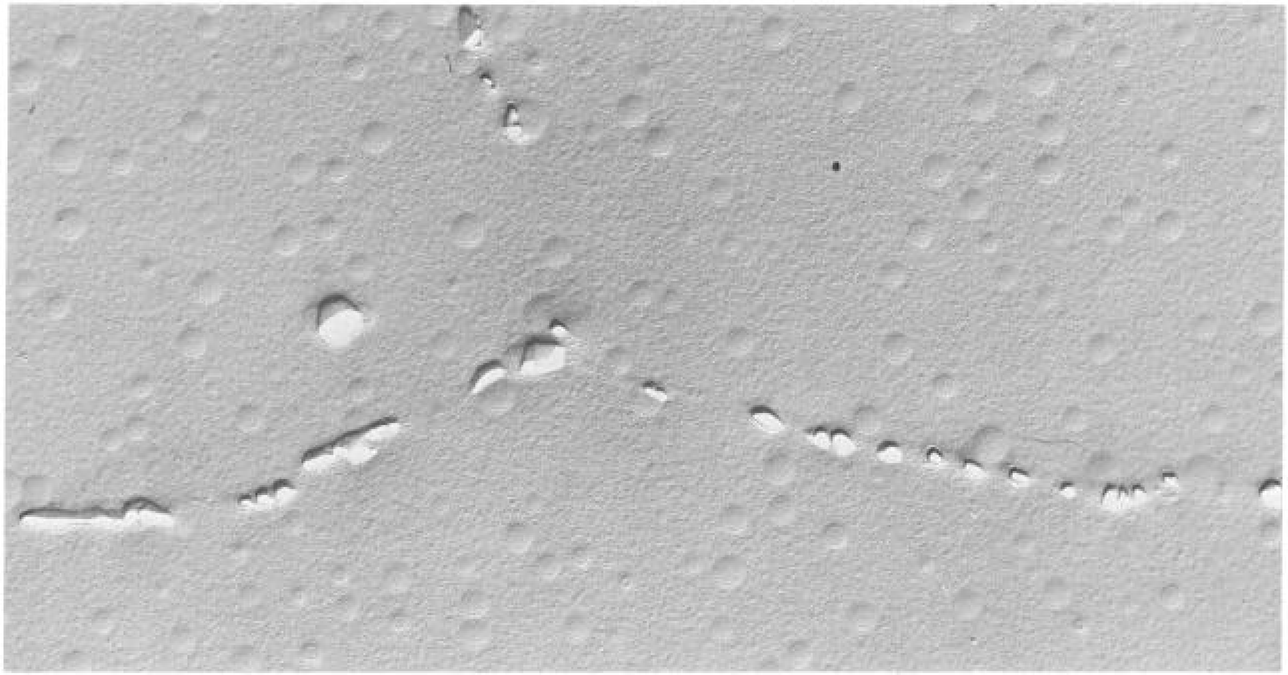


a. Ring rolled from 2050° F. furnace 500X

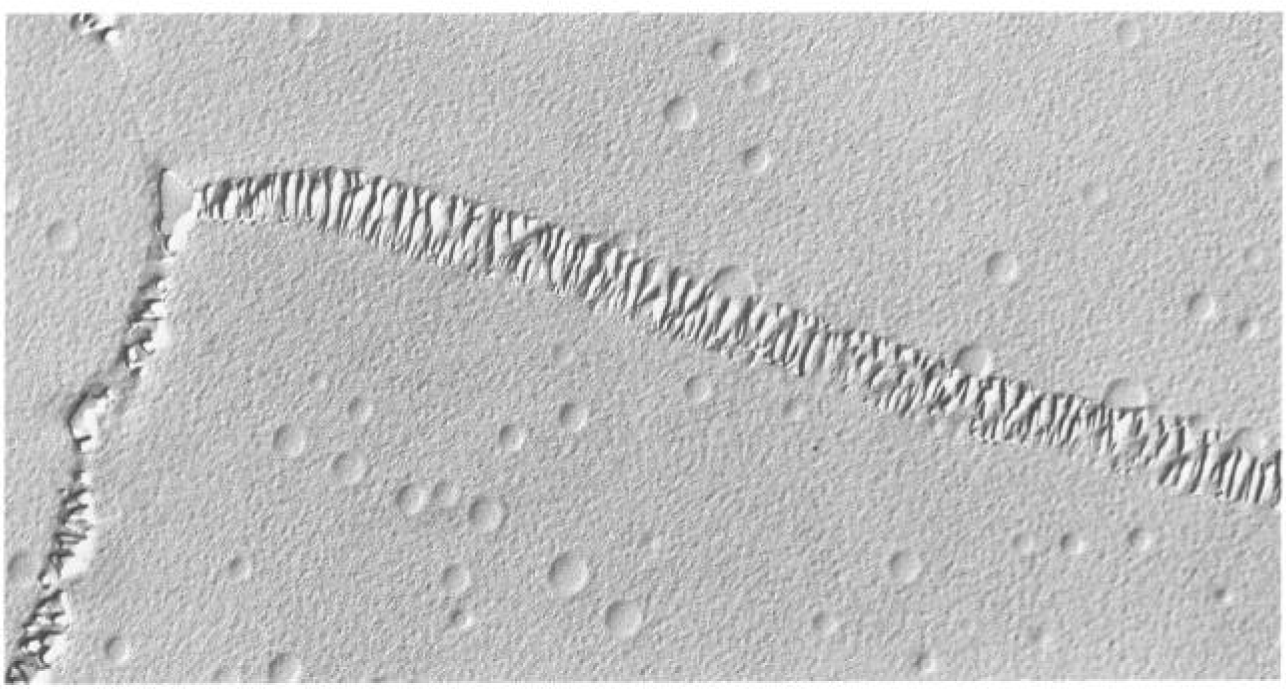


b. Ring rolled from 1975° F. furnace 500X

Figure 6. Micrographs at 500X of typical structures in Rene' 41 rolled rings heat treated 1950°-4 hrs.-AC; 1400°-16 hrs.-AC.

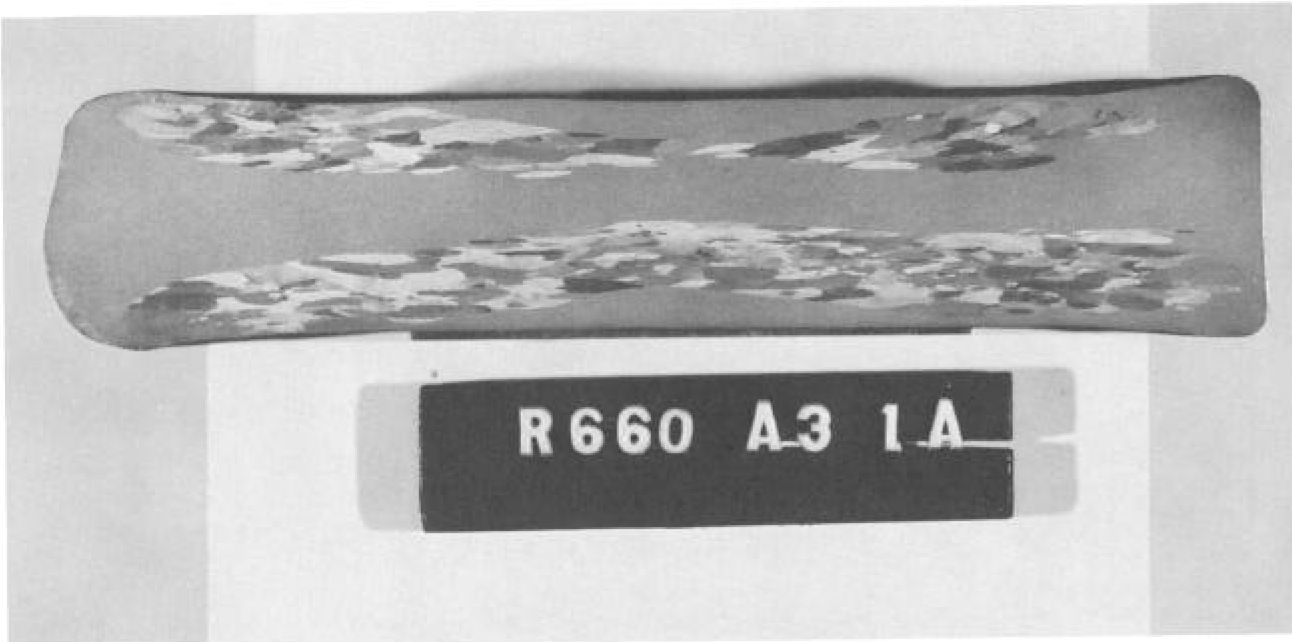


a. Spheroidized Grain Boundary $M_{23}C_6$

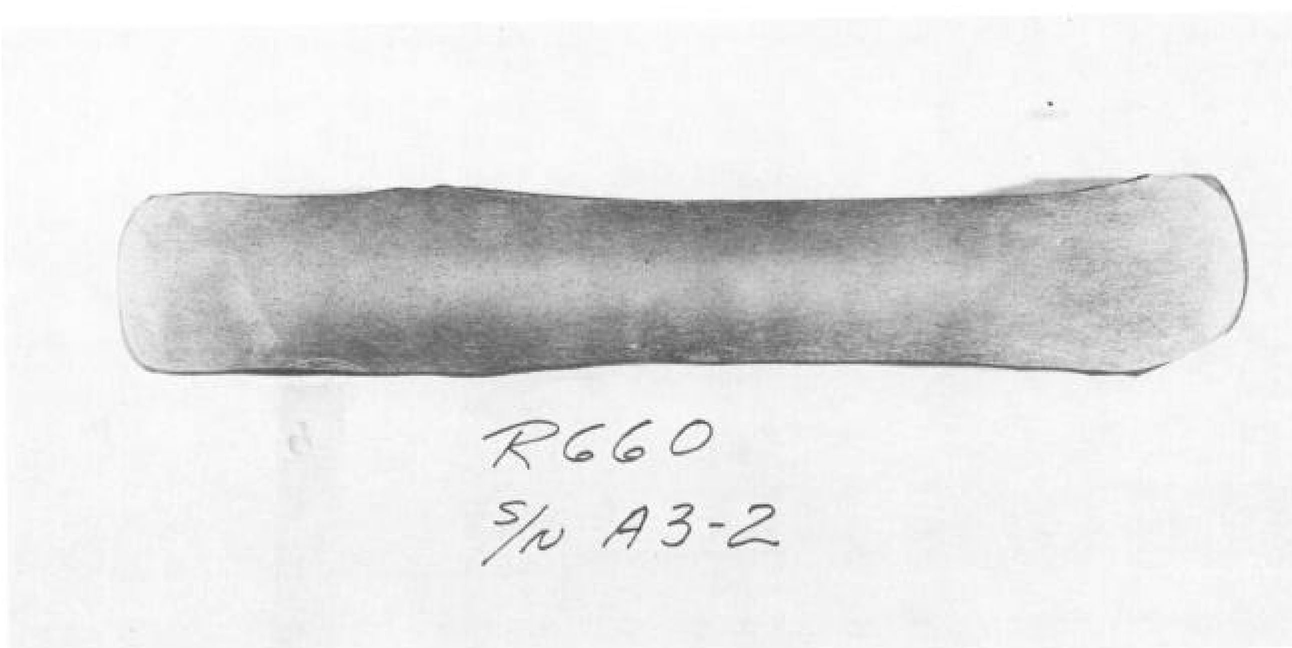


b. Cellular Grain Boundary $M_{23}C_6$

Figure 7. Varied morphologies of $M_{23}C_6$ in Waspaloy
(H.T. 1865°-4-OQ; 1550°-4-AC; 1400°-16-AC)



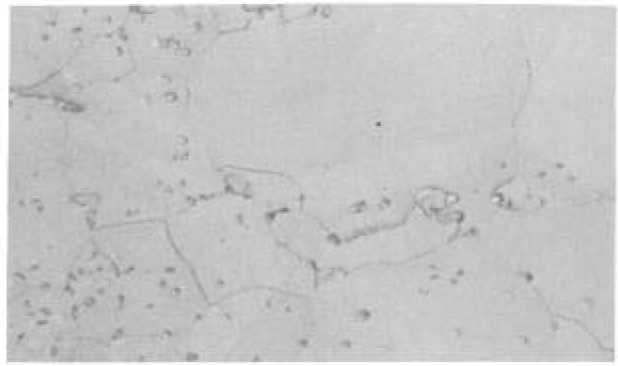
a. With Intermediate 2250° F. Anneal



b. Without any Intermediate Anneal

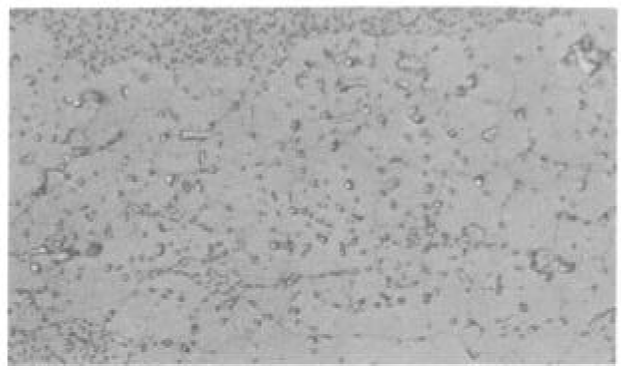
Figure 8. Macrosections at 1X of forged Astroloy powder compacts.

a. 2090° F. Anneal

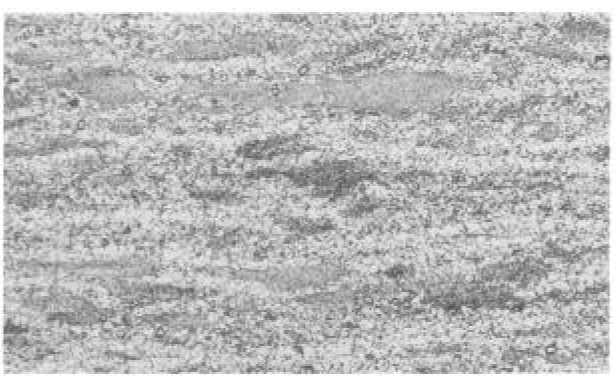


b. 2090° F. Anneal

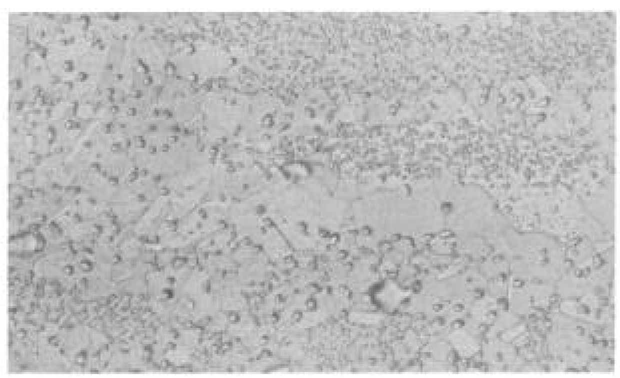
c. 2075° F. Anneal



d. 2075° F. Anneal



e. 2060° F. Anneal



f. 2060° F. Anneal

Structures at 100X

Structures at 500X

Figure 9. Sensitivity of Rene' 95 structure to the in-process anneal.