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EXAMINATION of the INFLUENCE of COATINGS on THIN SUPERALLOY SECTIONS

Volume I

DESCRIPTION and ANALYSIS

by
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GENERAL ELECTRIC COMPANY

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16. Abstract

The effects of an aluminide coating, Codep B-1, and of section thickness were investigated on two cast nickel base superalloys, Rene 80 and Rene 120. Cast section thicknesses ranged from 0.038 cm to 0.15 cm. Simulated engine exposures for 1000 hours at 899C or 982C in a jet fuel burner rig with cyclic air cooling were studied, as were the effects of surface machining before coating and re-machining and re-coating after exposures. The properties evaluated included tensile at R.T., 871C and 982C, stress rupture at 760C, 871C, 982C and 1093C, high cycle mechanical fatigue at R.T., and thermal fatigue with a 1093C peak temperature.

Thin sections had tensile strengths similar to standard size bars up to 871C and lower strengths at 982C and above, with equivalent elongation. Stress rupture life was lower for thin sections at all test conditions. The aluminide coating lowered tensile and rupture strengths up to 871C, with greater effects on thinner specimens. Elevated temperature exposure lowered tensile and rupture strengths of thinner specimens at the lower test temperatures. Surface machining had little effect on properties, but re-machining after exposure reduced thickness and increased metallurgical changes enough to lower properties at most test conditions.

Volume II, Detail Procedures and Data, CR-134792, is a detailed description of materials and test procedures, complete tabulation of all test results and summary.

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FOREWORD

This report is Volume I of a two volume report prepared by the Aircraft Engine Group of the General Electric Company under National Aeronautics and Space Administration Contract No. NAS3-16759, evaluating the effects of protective coating and section thickness on the mechanical properties of cast nickel-base superalloys. Volume I includes the introduction, description of tests and results with full discussion, summary, conclusions and recommendations. Volume II is a detailed description of materials and test procedures and a complete tabulation of all test results.

The program was sponsored by the Lewis Directorate, U.S. Army Air Mobility Research and Development Laboratory, National Aeronautics and Space Administration, Lewis Research Center, Cleveland, Ohio with Mr. John P. Merutka as Program Manager. The General Electric Company Program Manager was Mr. E. J. Kerzicnik. The author gratefully acknowledges the support of the Program Managers and the General Electric Company personnel who contributed to the program.

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UNIT CONVERSIONS

S.I. units are used throughout the report. Tables and Figures (where possible) have S.I. units with standard American units in parentheses. In the text, only S.I. units are shown. The conversions for commonly encountered values are listed below. The principal measurements were taken with instruments calibrated in American units and converted to S.I. units for the report.

	S.I.	American
Linear measure	0.00254 cm	0.001 inch
	0.038 cm	0.015 inch
	0.075 cm	0.030 inch
	0.15 cm	0.060 inch
	0.64 cm	0.25 inch
Temperature	760°C	1400°F
	871°C	1600°F
	899°C	1650°F
	982°C	1800°F
	1093°C	2000°F
Stress	689.5 MN/m^2	100.0 ksi
	634 MN/m^2	92.0 ksi
	565 MN/m^2	82.0 ksi
	310 MN/m^2	45.0 ksi
	262 MN/m^2	39.0 ksi
	145 MN/m^2	21.0 ksi
	117 MN/m^2	17.0 ksi
	55 MN/m^2	8.0 ksi
	34.5 MN/m^2	5.0 ksi

SUMMARY

The effects of an aluminide coating, Codep B-1, and of section thickness were investigated on two cast nickel base superalloys, Rene 80 and Rene 120. Cast section thicknesses ranged from 0.038 cm to 0.15 cm which are in the region of many present turbine blade thicknesses. The coating is necessary to extend turbine blade life by providing oxidation and hot corrosion protection. Simulated engine exposures for 1000 hours at 899C or 982C in a jet fuel burner rig (0.05M) with cyclic air cooling were studied, as were the effects of surface machining before coating and re-machining and re-coating after exposures. The properties evaluated included tensile at R.T., 871C and 982C, stress rupture at 760C, 871C, 982C and 1093C, high cycle mechanical fatigue at R.T., and thermal fatigue with a 1093C peak temperature. In the following summary of results, the data from an earlier program[1] on Rene 80 is incorporated wherever possible. All coated specimen data discussion is based on stresses calculated from the original metal thickness before coating.

Tensile Properties. The effects of coating, section thickness and exposure are quite similar for both alloys. Fig. A shows the variation in 0.2% Y.S. and U.T.S. with section thickness and temperature. The strength is expressed as % of the average strength of bare and coated standard size (0.64 cm diameter) bars. Bare specimen Y.S. is not affected by thickness at R.T., but is higher for the 0.038 cm specimens at 871C and lower at 982C. Coating lowers Y.S. about 5% at R.T. and 871C, and raises Y.S. for the thinner sections at 982C. The 982C exposure lowers Y.S. in all cases except for Rene 80 thicker sections at 871C (the only point of difference from Rene 120). In all cases the exposure causes lower Y.S. for thinner sections.

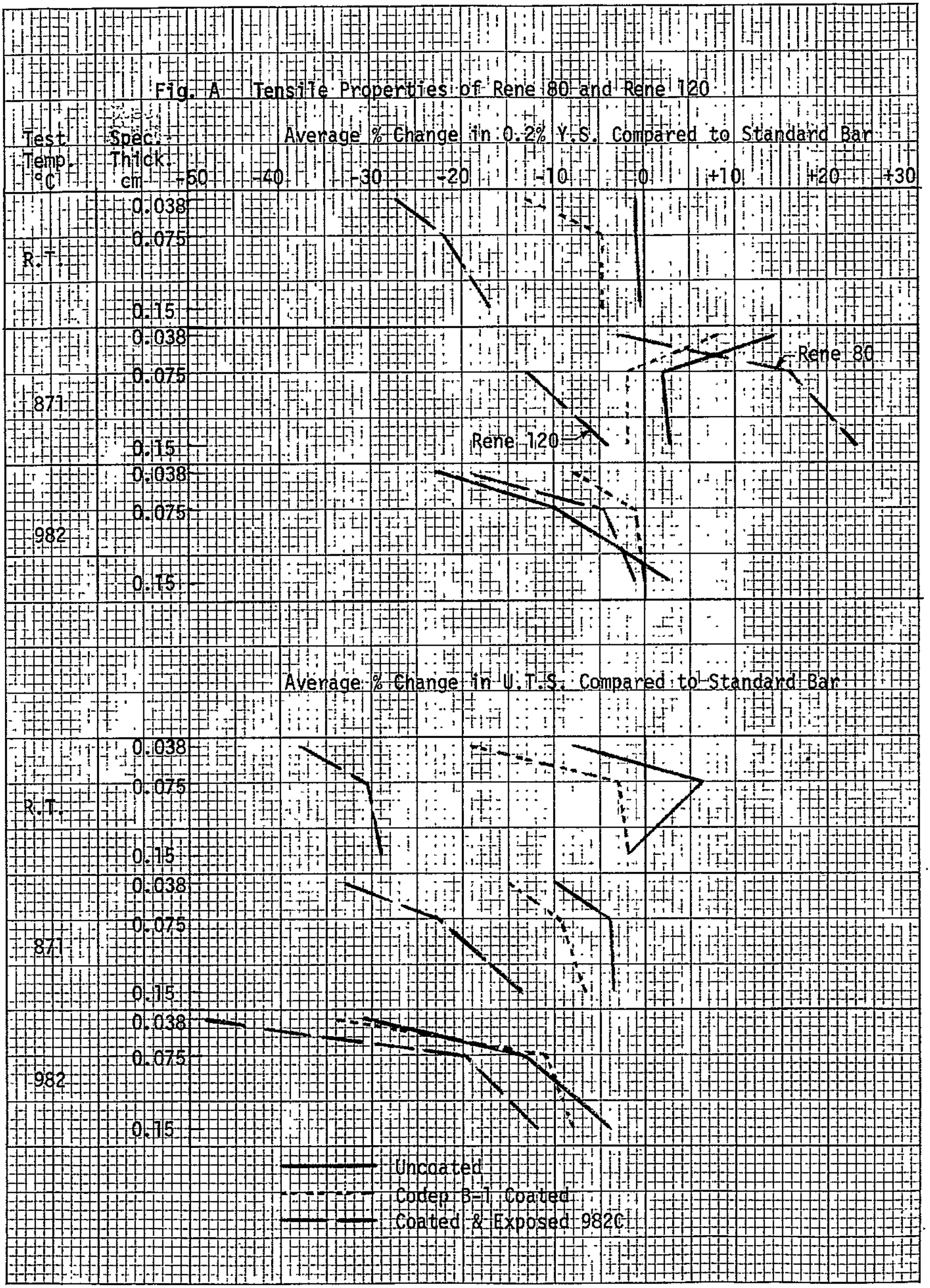


Fig. A Tensile Properties of Rene 80 and Rene 120
Test Temp. °C
Spec. Thick. cm
Average % Change in 0.2% Y.S. Compared to Standard Bar
-50
-40
-30
-20
-10
0
+10
+20
+30
R.T.
871
982
0.038
0.075
0.15
Rene 80
Rene 120
Average % Change in U.T.S. Compared to Standard Bar
Uncoated
Codep B-1 Coated
Coated & Exposed 982C

Except for the 0.075 cm uncoated specimens at R.T., the thinner the section, the lower the U.T.S. for all conditions. For bare and coated specimens, the higher the temperature, the lower the strength compared to standard size bars. Coating reduces U.T.S. at R.T. and 871C, except for the 0.075 cm specimen at 982C. Exposure at 982C significantly reduces U.T.S. at R.T., with smaller losses as the temperature increases, and smaller losses for thicker specimens.

Elongations of the thin sections are equivalent to those of standard size bars. Coating lowers elongations of the 0.038 cm specimens, with lesser effects on the thicker specimens. The 982C exposure reduces elongation at all temperatures, with the greatest loss at 871C.

A limited number of coarse grained specimens were tested at R.T. and 871C (0.075 cm thickness of Rene 80 and 0.15 cm thickness of Rene 120). Grain size was 2 to 4X the normal specimens. The coarse grain material has higher properties at 871C than the normal grain size material of the same thickness for both alloys, and at R.T. for Rene 80. The coarse grain Rene 120 has lower properties at R.T.

The effects of removing the as-cast surface of the 0.075 cm specimens by machining, before and after exposure are shown in Fig. B for bare, coated and coated plus ~1000 hour exposure at 982C specimens. The data is "corrected" for thickness back to the original dimension. On this basis, the initial surface machining (SM) has little effect on strengths for bare or coated specimens. After exposure, the SM specimens have higher R.T. strengths and lower elevated temperature strengths. After exposure re-SM results in lower R.T. and 871C strengths, but improves U.T.S. for both alloys and Y.S. for Rene 80 at 982C. The re-treatment and coating returns Y.S. of Rene 80 to original levels at all temperatures, while Rene 120 is not improved. The second 982C exposure causes relative strength losses, particularly at elevated temperature. Elongations are less affected, and in most cases are somewhat higher for the surface machined samples compared to the as-cast 0.075 cm conditions.

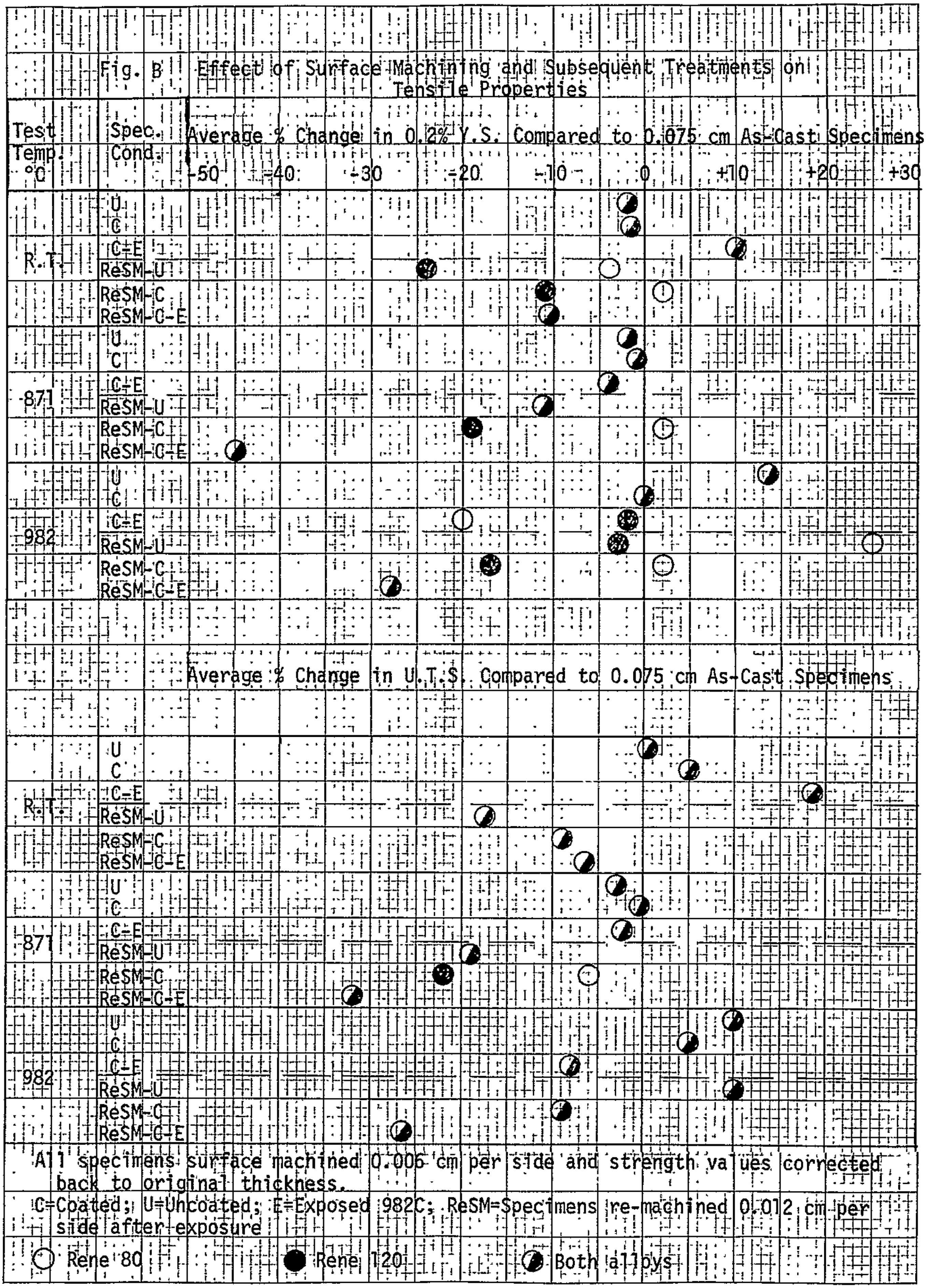


Fig. 8 Effect of Surface Machining and Subsequent Treatments on Tensile Properties
Test Temp. °C
Spec. Cond.
Average % Change in 0.2% Y.S. Compared to 0.075 cm As-Cast Specimens
-50 -40 -30 -20 -10 0 +10 +20 +30
R.T.
871
982
U
C
C+E
ReSM-U
ReSM-C
ReSM-C-E
Average % Change in U.T.S. Compared to 0.075 cm As-Cast Specimens
All specimens surface machined 0.006 cm per side and strength values corrected back to original thickness.
C=Coated; U=Uncoated; E=Exposed 982C; ReSM=Specimens re-machined 0.012 cm per side after exposure
Rene 80
Rene 120
Both alloys

Stress Rupture Properties. The rupture lives of the two alloys expressed as a percentage of the standard size bar life are shown in Fig. C. The differences in behavior are greater than noted in tensile strengths. For both alloys, bare or coated, the thinner the specimen, the lower the rupture life.

Maximum life for the 0.038 cm Rene 80 bare specimens is ~30% at 760C and 1093C, while the minimum is ~2% at 982C. The fact that life was lower at 760C for the thinner sections in spite of their beneficial fine grain size indicates that the geometry effect on fracture is of considerable magnitude.

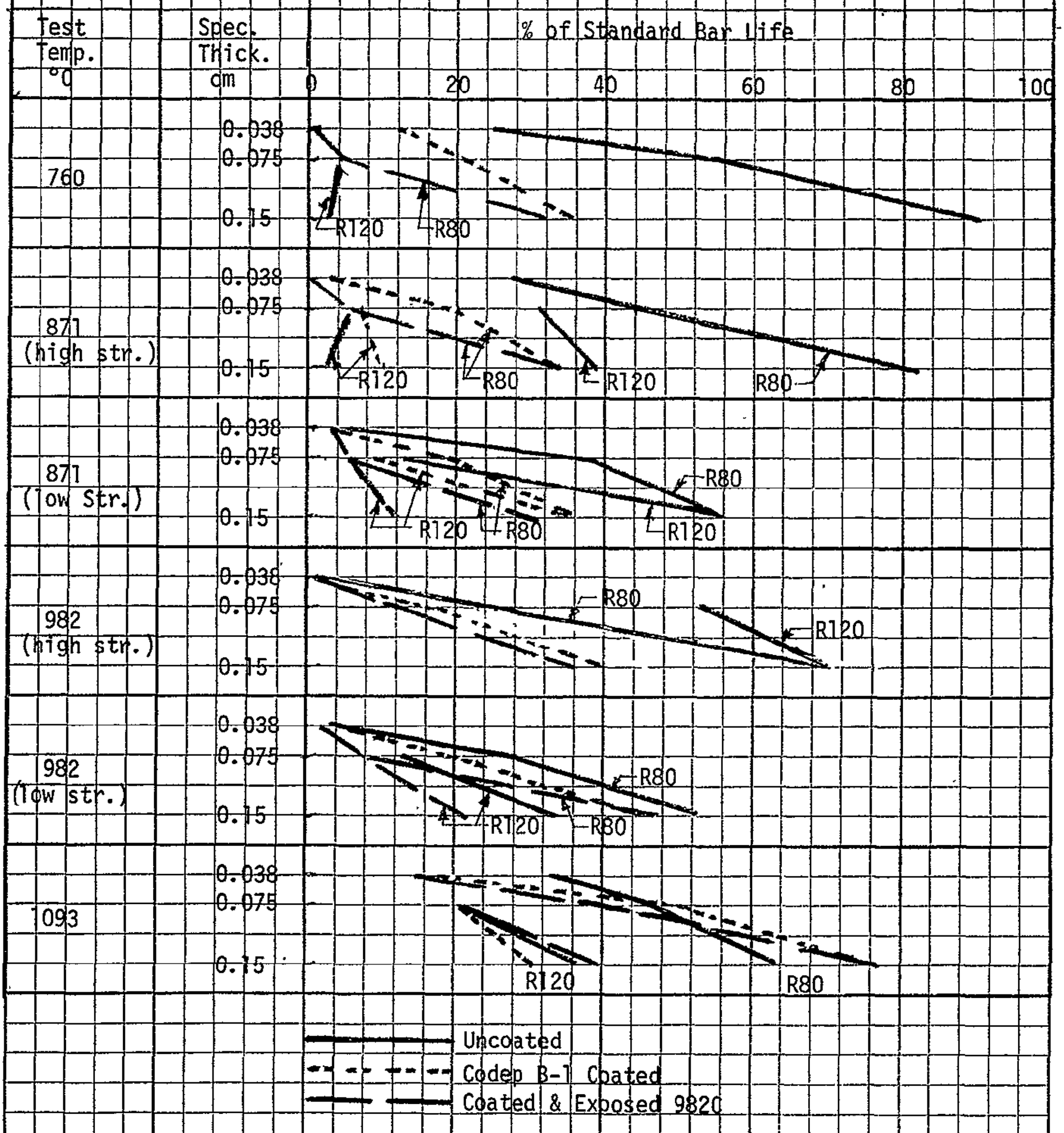
Coating lowers life for both alloys at all thicknesses at 760C and 871C and for high stress at 982C. The low stress at 982C and the 1093C tests show little effect. It should be noted that if the stress is based on the sound metal thickness below the coating, there is no reduction in life due to coating except at 760C.

Exposure further lowers the life of both alloys at 760C and 871C, with smaller, or no losses at 982C (low stress) and 1093C. Rene 120 thin sections generally have lower relative lives than Rene 80 except for the 982C, high stress, bare condition.

Coarse grained specimens, as described previously, were stress rupture tested at 871C and 1093C. Bare or coated, the coarse grained material have lower life at 871C (by 8 to 38%) and greater life at 1093C (by 20 to 100%) compared to standard bar life.

The effects of SM and re-treatments (described previously) on stress rupture life are shown in Fig. D. Lives are corrected for section thickness and compared to as-cast 0.075 cm specimen lives for the equivalent condition. Because of the greater variability in the rupture data compared to the tensile data, and the greater effect of section thickness on rupture life, there is a greater spread in the results.

Fig. C Stress Rupture Life of Rene 80 and Rene 120 Compared to Standard Bar Life



General conclusions are that the SM lowers lives at 760C and 871C and raises lives at 982C and 1093C, bare or coated. Relative life after exposure is lower for Rene 80 and higher for Rene 120 at the three higher temperatures, and the reverse at 760C. Re-surface machined specimens after exposure, whether bare, re-treated and coated or re-exposed at 982C, have lower lives than the respective specimens before re-SM. Although the 0.075 cm specimens were not initially aided in rupture life by the re-coating treatment, the improved life occurring in thicker specimens (>0.075 cm) with exposure, combined with the additional environmental protection afforded by the new coating make the re-treatment process more desirable.

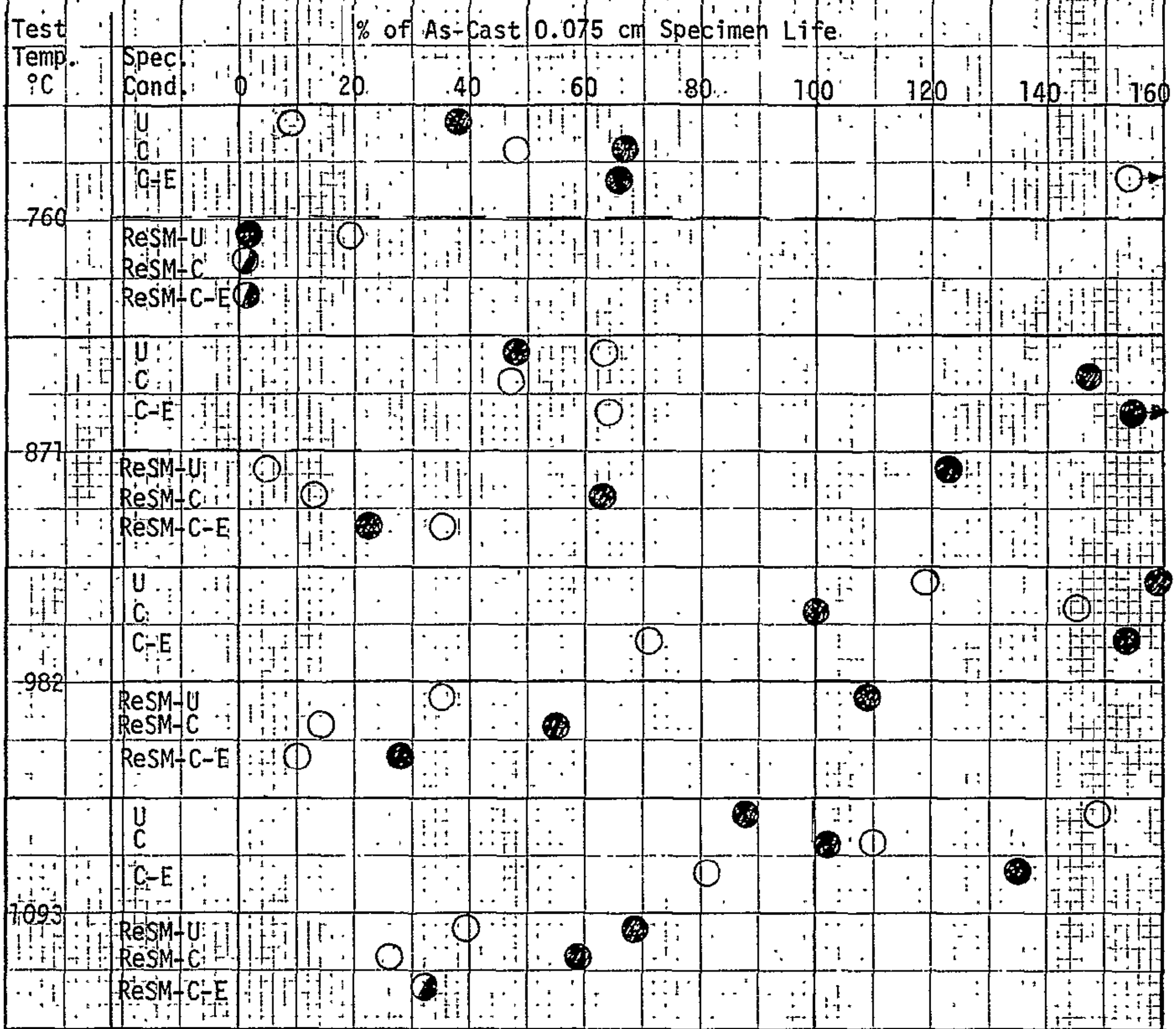
Thermal Fatigue. With a 1093C peak, 204C minimum temperature, 10 second heating time, 50 second hold time and 20 second cooling time thermal cycle, no normal thermal fatigue cracks could be produced in the round 0.075 cm thick coated or coated/982C exposed specimens of either alloy up to 4000 cycles.

High Cycle Mechanical Fatigue. The 10^7 cycle, reversed bending fatigue strengths for Rene 120 are:

Uncoated	269 MN/m^2
Coated	269 "
Coated, exposed 982C	286 "

The coating and exposure have virtually no effect (with stress measured at exterior of coated samples). These values are lower than previously found for Rene 80 by 30% (uncoated) to 7% (coated and exposed).

Fig. D Effect of Surface Machining and Subsequent Treatments on Stress Rupture Life



All specimens surface machined 0.006 cm per side and strength values corrected back to original thickness.
C=Coated; U=Uncoated; E=Exposed 982C; ReSM=Specimens re-machined 0.012 cm per side after exposure

○ Rene 80 ● Rene 120 ◐ Both alloys

INTRODUCTION

Mechanical property data for use in the design of cast aircraft gas turbine blades generally have been obtained from testing cast standard size bars, usually about 0.64 cm (1/4") diameter or from similar bars machined from sections of large blades. However, small engine blades and hollow and air-cooled sections of large engine blades have thicknesses and shapes quite different from the standard test bars. Knowledge of the mechanical properties of the typical thin blade sections and their differences compared to standard test bars is necessary to permit proper design of such blades. In addition, the exposure to high temperature during engine operation causes internal metallurgical changes in the commonly used nickel-base superalloys with corresponding changes in mechanical properties. Particularly, ductility may be adversely affected, influencing fatigue, impact and rupture behavior. Thus the effects of engine exposure conditions on the mechanical properties must be known.

The surface and near surface condition of the castings have a more significant effect on thin sections compared to thicker sections. Such conditions include surface roughness, finer surface grains compared to the interior, microstructure and surface reactions with the mold material and during heat treatment. The latter tends to remove elements such as chromium and aluminum and to decrease the amount of carbides near the surface. All such effects should be determined so that they may be avoided or taken into account in design.

The problems of environmental attack - hot corrosion (sulfidation) at temperatures up to about 1000C and oxidation at all temperatures - are greater for thin sections, where a similar depth of attack produces a greater loss in load carrying capacity than for thicker sections. Increasing operating temperatures and longer expected lives of more recent engines, coupled with the poorer, hot corrosion resistance of newer high strength superalloys (lower chromium contents) have necessitated development and usage of protective coatings. Unfortunately, the coatings require heating the blades (which can alter properties) and produce a layer of from 0.005 cm to 0.013 cm in thickness of a material having different properties than the base alloys. This layer represents a greater fraction of the cross section of thin parts and it becomes correspondingly more important to determine the effects of coatings on the mechanical properties for proper design.

Elevated temperature exposure during operation, in addition to the internal metallurgical changes, causes external attack of the coatings and interdiffusion of the coating and base alloy. The changes in composition of the coating and the alloy below it affect properties (by formation of brittle phases, for example), and again the effects are greatest for thinner sections and should be evaluated.

The life limitations of the presently used aluminide protective coatings and the high costs of the newer complex cooled turbine blades make it desirable to be able to re-use the blades after the initial coating has been exhausted. Removal of the coating remains, and the affected material below it followed by application of a new coating would permit further use of turbine blades. The change in thickness and the additional applied heat treatments would be expected to affect properties.

Other microstructural features will differ in thinner sections; grain size (both absolute size and number of grains across the section) and orientation, carbide size and morphology, eutectic γ' nodule size and distribtuion, etc.

The actual physical geometry of the test specimen, in addition to its thickness, will affect the measured properties. For completely rational design and life prediction of cast superalloy turbine blades, all of the preceeding effects should be known.

A review of the expected effects is included along with test results in a prior NASA contract, (NAS3-15557),[1] which together with other work [2] [3] [4] on a range of cast nickel base superalloys have given indications of the magnitudes of some the effects of section thickness of as small as 0.075 cm on mechanical properties. Tensile strengths are slightly lowered (roughly 5-10%) at room temperature to 982C while ductility is not affected noticeably. Mechanical high cycle fatigue properties at room temperature are not changed. However, major losses in stress rupture life (up to 80%) do seem to occur at temperatures from 760C to 982C. Protective coatings may be detrimental or beneficial, depending on the type of coating, the alloy and test condition. Where the surface protection afforded by the coating prevents material attack or loss, it is usually beneficial. If the coating is brittle, then early coating cracking provides fracture initiation sites and less protection and may reduce properties. If the coating thermal treatments are not compatible with the base alloy treatment, some losses may occur.

The effects of elevated temperature exposures on the properties of cast thin sections were investigated. [1] For alloy Rene 80, ignoring material lost due to oxidation, exposure at 982C greatly lowered tensile yield and ultimate strengths and ductility at room temperature, with smaller losses at higher temperatures up to 982C, with no loss at 1093C. Stress rupture life and ductility were lowered up to 982C but not at 1093C. Exposures at 1093C do not decrease tensile properties as much as the lower temperature exposure, but cause a greater loss of stress rupture life up to 1093C. A major benefit of the protective aluminide coating was in reducing or avoiding the significant losses in bare specimen properties due to exposure.

Reported results on the other factors are virtually nonexistant. Even the results mentioned above, while useful in delineating problem areas and approximate magnitudes of effects, are still not sufficient or accurate enough for design purposes.

The general purposes of the present work are to obtain sufficient data (in conjunction with prior data) for two nickel base superalloys; Rene 80 and Rene 120 on the effects of Codep B-1 aluminide coating, exposure and section thickness to permit reliable design and to investigate some of the other factors mentioned previously. The following are the specific objectives of the program:

1. Obtain tensile data at R.T., 871C, 982C and stress rupture data at 760C, 871C, 982C and 1093C with 2 stress levels at the intermediate temperatures for Rene 80 uncoated and Codep B-1 coated for specimen cast thicknesses of 0.038, 0.075 and 0.15 cm with four specimens per test point. A different heat of material to be used than initially used in Ref. 1.

2. Obtain the same test data as in (1) for coated specimens exposed for 1000 hours at 982C cyclically (unstressed). Obtain tensile data and limited stress rupture data for the 0.075 cm thickness similarly exposed at 898C.

3. On the same new heat of Rene 80, determine thermal fatigue behavior of coated and coated + 982C exposed material compared to earlier heat.

4. For two heats of Rene 120, obtain tensile data at R.T., 871C and 982C and stress rupture data at 760C, 871C, 982C and 1093C for uncoated and Codep B-1 coated specimen cast thicknesses of 0.075 and 0.15 cm with four specimens per test point, and limited data for 0.038 cm thickness.
5. Obtain same test data as in (4) for coated specimens exposed for 1000 hours at 982C cyclically (unstressed) for the 0.075 and 0.15 cm thicknesses.
6. Determine R.T. high-cycle mechanical fatigue (bending) strength of uncoated, coated and coated + exposed Rene 120 and thermal fatigue behavior of coated and coated and exposed Rene 120 in the 0.075 cm thickness.
7. For both Rene 80 and Rene 120 with the as-cast surface ground off, ∿ 0.006 cm (0.0025 inch) per side, obtain tensile test data at R.T., 871C and 982C and stress rupture data at 760C, 871C, 982C and 1093C for uncoated, coated and coated + exposed 1000 hours at 982C in the 0.075 cm thickness with four specimens per test point for the Rene 80 and three for Rene 120.
8. For both surface machined, coated Rene 80 and Rene 120 after 1000 hour exposure at 982C, with the coating and sub-coating region ground off, ∿0.010 cm per side, obtain same data as in (7).
9. Investigate grain size in a limited manner. For Rene 80, in the 0.075 cm thickness, on specimens cast with no nucleation coating on the mold, perform tensile tests at R.T. and 871C and stress rupture tests at 871C and 1093C. For Rene 120 in the 0.15 cm thickness, using coarse columar grain castings (tested at right angles

9. (Cont'd)

to the long axis of the grains), perform tensile tests at R.T. and 871C and stress rupture tests at 871C and 1093C. One or two specimens to be tested at each point uncoated and Codep B1 coated.

10. Perform metallurgical analyses, using optical microscopy, x-ray diffraction, electron microprobe, etc. techniques to elucidate structural changes in the base alloys and coatings and their effects on properties.

MATERIALS AND TEST CONDITIONS

Alloys and Coating

For the initial work evaluating the influence of coatings on thin superalloy sections,[1] a representative widely applied cast nickel base alloy, Rene 80, was selected. In the present program, the same alloy is used in order to provide sufficient mechanical property data, in combination with the prior results, to permit reliable usage in design of cast turbine blades. In addition, a newer higher strength alloy, Rene 120, was chosen for similar property determinations. The latter alloy has higher strength and better oxidation resistance than Rene 80, with somewhat poorer hot corrosion resistance, and it has several applications in new engines. Nominal chemical compositions in weight percent for both alloys are as follows:

	Al	Ti	Co	Cr	Mo	W	Ta	Zr	B	C	Ni
Rene 80	3.0	5.0	9.5	14.0	4.0	4.0	-	0.03	0.02	0.17	Bal.
Rene 120	4.3	4.0	10.0	9.0	2.0	7.0	3.8	0.07	0.02	0.17	Bal.

As can be seen, Rene 120 has a greater amount of γ' formers (Al, Ti, Ta) and higher total refractory element content (Mo + W + Ta), compensated by lower Cr. Both alloys are PhaComp controlled to avoid formation of brittle intermetallic phases. Two heats of each alloy were used, and each met all specification chemistry requirements.

In order to have the cast test specimens most closely resemble cast turbine blades, the same casting design and methods were used as previously reported[1] for two thicknesses; 0.075 cm and 0.15 cm. The casting is basically a hollow box (cored) with flat sides and gated along one edge. All mold materials and casting parameters are the same as used for turbine blades. The mechanical property test specimens are cut from the flat sides of the box, and the flat faces are normally used without machining. For the 0.038 thickness, the original mold design was modified to make a shorter, narrower box to aid in filling the extremely thin walls. Rene 80 specimens were successfully cast in the new design, but Rene 120 was so difficult that only a few were produced.

In order to investigate the effects of grain size, in a limited manner, one mold of Rene 80 specimens, 0.075 cm thickness were cast without using the normal nucleation coating on the mold. A coarser grain size resulted from this compared to the usual practice. For Rene 120, some coarse columnar grained specimens 0.15 cm thick were obtained. Normally these would have been rejects for the primary program. The long columnar grains grew from the gating area, and consequently were perpendicular to the gage length of the test specimens.

After casting, the flat sides were cut apart and inspected by normal visual, fluorescent penetrant and x-ray radiographic methods. The grain size (requirement is 0.16 cm diameter maximum) was measured by inspection of the radiographs and were verified by macro-etching and metallographic inspection of several samples. No macro-etched material was used for mechanical testing. Only sound specimens were processed further. Many of the "flats" were bowed. These were flattened during heat treatment by stacking with a weight on top. The heat treatment for each alloy is:

Rene 80	1218C, 2 hours, in vacuum, rapid cool
	1093C, 4 hours, in vacuum, rapid cool
	1052C, 4 hours, in vacuum or in coating process
	843C, 16 hours, in vacuum.
Rene 120	1204C, 2 hours, in vacuum, rapid cool
	1080C, 4 hours, in vacuum or in coating process
	1080C, 15 minutes in vacuum, rapid cool
	927C, 8 hours, in vacuum
	760C, 16 hours in vacuum.

In order to investigate the effect of surface condition on properties, a group of specimens from each alloy were surface ground on each side to remove ~0.06mm per side. The grinding was performed after the first two heat treatment steps for Rene 80 and after the first step for Rene 120, thus removing all the material affected by the high temperature vacuum exposures. The third heat treatment step for Rene 80 and the second step for Rene 120 is the same as the normal thermal cycle for the coating used in the program, Codep B-1. All specimens receive the identical heat treatment whether coated or not.

The Codep B-1 coating, as used in the previous work,(1) is a normal aluminide-type coating; producing a total thickness of ~0.04 mm, the outer half of which is NiAl (with dissolved Cr, Ti and other base alloy elements) and the inner half is a multiple phase region containing NiAl, σ phase and carbides, etc. The Al content of the outer layer is generally near 28%, by weight, and coating thickness is very uniform and reproducible. Initially each specimen had coating thickness measured using a Dermitron coating thickness gage, but this was discontinued, and thickness verified metallographically on several specimens in each coating run and on specimens later tested and cut up for metallography. All coatings met specification requirements.

Exposures

Simulation of engine thermal exposure was done by exposing coated specimens in the products of combustion of JP-5 fuel in the burner rig used previously.(1) Most of the exposure were run at a representative temperature of 982C for approximately 1000 hours. Some Rene 80 was also exposed at 899C. The specimen holder was rotated during exposure, and specimens were cycled to ~94C in under 2 minutes by placing in a compressed air blast every hour during the working day, for an average total of 160 cycles. No coating problems were observed after exposure for any specimen of either alloy.

To evaluate the effects of the coating - alloy interactions during exposure, specimens were surface ground on each side (~0.11mm) after exposure to remove the coating and its sub-diffusion layers, and tested in this condition. Some of the specimens surface machined after exposure were given their standard heat treatment and re-coated to determine if original properties could be regained. Additional specimens were then re-exposed as before to measure subsequent changes in properties.

Mechanical Property Tests

Four types of tests were performed: tensile,stress rupture, high-cycle mechanical fatigue, and thermal fatigue. Tensile tests were performed at room temperature, 871C and 982C, which covers the range of engine operating temperatures for which tensile properties are desirable. Four specimens per test point were used in the following conditions: bare; coated; coated and exposed at 982C; surface machined, bare; surface machined, coated; surface machined, coated and exposed at 982C; surface machined, coated, exposed and re-surface machined; surface machined, coated, exposed, re-surface machined, re-coated; and surface machined, coated, exposed, re-surface machined, re-coated and re-exposed. The first three conditions were used for all thicknesses of one heat of Rene 80 and two heats of Rene 120; the surface machining was used on one thickness, 0.075cm, and one heat of each alloy. For Rene 120, only three specimens per test point for the surface machined conditions were run. Tensile tests on the coarser grained specimens were performed for each alloy, bare and coated at R.T. and 871C in duplicate, except for Rene 120 only at 871C. All stress rupture tests were run using dead weight loading. The following conditions for all thicknesses of bare, coated and coated and exposed Rene 80 (one heat) were tested:

760C, 565 MN/m²

871C, 310 MN/m² and 262 MN/m²

982C, 145 MN/m² and 117 MN/m²

1093C, 34.5 MN/m²

For the surface machined conditions listed under tensile tests, stress rupture testing was performed at the four temperatures above, with only the lower stresses at the two intermediate temperatures. Four specimens per test point were run. The 0.075 cm thick coaser grained specimens were tested in duplicate, bare and coated at 871C (lower stress) and 1093C.

Rene 120 stress rupture tests were conducted on two heats as follows on bare, coated and coated and exposed material:

760C, 634 MN/m² (both heats)

871C, 379MN/m² (first heat), 310 MN/m² (second heat)

982C, 172MN/m² (first heat), 145 MN/m² (second heat)

1093C, 55 MN/m² (both heats)

Four tests per condition were run for the 0.075 cm and 0.15 cm thicknesses, with single specimens bare and coated of 0.038cm thickness at two conditions: 760C and 982C (lower stress). Three tests per condition were made for the same surface machined conditions of the 0.075cm thickness as listed above for Rene 80 (only the lower stress at the two intermediate temperatures). Coarse, columnar grained 0.15 cm thick Rene 120 specimens were tested in duplicate, bare and coated at 871C (lower stress) and 1093C.

Mechanical high-cycle fatigue tests, in reversed bending, at R.T. were performed on 0.075 cm thick Rene 120 specimens bare, coated and coated and exposed. Four specimens in each condition were used. Rene 80 had been tested similarly in the previous program. (1) Thermal fatigue tests on the rounded leading edges of the 0.075 cm castings were run on coated and coated, exposed samples of both alloys. The cycle was the same as used before[1]: 1093C peak temperature, <204C minimum temperature with a 10 second heating time, 50 second hold time at peak and 20 second cooling time. Four specimens of each condition were tested.

TEST RESULTS

Tensile Tests, Rene 80

The tensile data as a function of temperature for the three thicknesses of castings tested are shown in Figs. 1-3 for 0.2% yield strength, Figs. 4-6 for ultimate tensile strength and Figs. 7-9 for elongation. The points plotted are the averages for all specimens tested, in the present and previous[1] program. The 0.038 cm nominal thickness specimens are all from a single heat, B353 and total 12 specimens for each curve (their actual average thickness is 0.045 cm). For the 0.075 cm thickness uncoated or coated, each point represents a total of 3 to 10 specimens from 2 or 3 of the heats of Rene 80; exposure data are 2 to 6 specimens from 1 or 2 heats. The uncoated curves total 16 specimens, the coated total 32 specimens and the coated and exposed at 982C total 20 specimens. The 899C exposed specimens total 5, from a single heat, B322. All data at 1093C is from the earlier work, Heat BV231.[1] The data for the 0.15 cm thickness at R.T. and 982C include two heats with 5 or 6 total specimens per point; at 760C and 871C, single heats with 2 to 4 specimens for a total of 20 uncoated tests, 20 coated tests and 15 coated and exposed tests.

The coating has no significant effect on yield strength at any temperature for the 0.075 cm and 0.15 cm thicknesses. (All coated specimen strengths based on original pre-coating dimensions). For the 0.038 cm thickness, the yield strength is lowered ~10% at R.T. and 760C, but not at 982C. The 1000 hour exposures at 982C lowered R.T. yield strengths 13 to 16% for all thicknesses. However, at 760C and higher for the thicker sections, yield strength is increased by an average of over 10%. The thinnest section still has a reduced yield strength at 871C, but not at 982C. The 899C exposure was more detrimental than the 982C exposure at R.T. and 871C; no difference at 982C.

The effects of coating on R.T. ultimate tensile strength are about the same as on yield strength. Exposures had a more drastic effect on U.T.S. than on Y.S. After the 982C exposure, the losses at R.T. were from 24 to 27% for all sections; at 760C from 13 to 16% (two larger thicknesses); at 871C from 15 to 22% (two smaller thicknesses); at 982C, there was a loss for the 0.038 cm specimens only. The 899C exposure caused a greater loss at R.T., with similar results at 871C and 982C.

Effects on elongation are somewhat more difficult to distinguish because of the occasional failures at the end of the gage lengths. With that in mind, it appears that coating was detrimental for the thin sections up to 982C. For the 0.15 cm thickness, there was no effect up to 871C. Specimen exposure at 899C and 982C reduce the elongation the most, particularly at 871C. Large losses were present at lower temperatures, while there were no losses at 1093C.

The curves of Figs. 1 to 9 are suitable for use as average design values. However, to show the effects of section size more readily, the same data is replotted in Figs. 10 to 12 for Y.S., U.T.S. and elongation, respectively. In these graphs, the actual average thickness for each point is used, and the average standard size bar data is included. For the standard bar data, coated and uncoated are averaged together, since there were no significant differences.

At temperatures up to 871C, for Y.S., Fig. 10, there is a slight tendency towards higher values for the thinnest flat sections, except for the coated thinnest specimens at R.T. Standard bar values are higher than the flat specimens at R.T. and 760C, but lower at 871C. At 982C there is a slight increase in Y.S. with increasing thickness and at 1093C, little difference. Standard bar values at the latter temperatures are higher than the flat specimen values.

Except for the 0.075 cm thickness, at R.T., the values for U.T.S., Fig. 11, show that decreasing thickness results in decreasing strength, at all temperatures coated or uncoated.

Elongations (Fig. 12) increased greatly with increased thickness at 871C and 982C for coated and uncoated specimens, and exceeded standard bar levels for several conditions. At R.T., the 0.075 cm specimens had higher elongations than thicker or thinner specimens or standard bars.

In order to make proper comparisons of properties for the specimens involved in evaluation of the various surface machined conditions, it is necessary to consider the decreased thickness after surface machining (SM). The average thickness after the initial machining was 0.0676 cm, and after the coating, exposure and re-surface machining was 0.0465 cm. To compare all data on the equivalent 0.075 cm thickness basis, the actual test values are "corrected" by the ratios found for the respective properties on Figs. 10 to 12 for the actual thickness ratios. For each SM specimen, the correction factor was obtained by reading the Y.S., U.T.S. or elongation value for the 0.075 cm thickness from the appropriate curve and dividing it by the value for the actual thickness of the SM specimen from the same curve. The same ratio was used for coated and exposed samples as for coated only. Table I contains the average tensile properties for all the SM conditions, together with the average as-cast surface specimens of 0.075 cm thickness for comparison. It should be kept in mind that the "correction" applied accounts only for section size effects and does not include the metallurgical changes and the effects they may have on relative section size values.

At R.T., the SM coated and uncoated specimens had slightly lower Y.S. and slightly higher U.T.S. and elongation than did the as-cast specimens. At 871C strengths were slightly lower for the SM specimens, and at 982C were slightly higher, while elongations were generally equivalent or higher for SM samples. On the average, then, the SM specimens were about equal in strengths and higher in elongation. SM specimen exposure caused smaller relative losses at R.T. and 871C, but greater losses at 982C. After exposure, machining off the coating has affected sub-coating area resulting in increased elongation at all temperatures. The strengths at 982C are back to their pre-exposure level, while at the lower temperatures, the strengths are less. Re-heat treating and coating the specimens after the second SM did increase Y.S. at R.T. and 871C to above values, but lowered Y.S. at 982C. The U.T.S., while raised at R.T. and 871C by the re-treatment and coating, did not attain initial values at any temperature. Elongations at all temperatures are about half the initial levels. The second 1000 hour exposure at 982C of the re-treated specimens caused great losses in all properties at all temperatures and represented the poorest property levels of any condition.

The results of the tensile tests on the coarser grained 0.075 cm specimens are in Table II, together with the averages of the normal specimens, from Figs. 2, 5 and 8. The Y.S. of all coarse grained specimens were higher than the normal specimens, while U.T.S. was lower in all cases except at R.T. for coated samples. Elongations for the coarse grained material were slightly lower at R.T. and slightly higher at 871C than normal material.

Stress Rupture Tests, Rene 80

In order to present the data from the various temperatures and stresses employed on a single plot, the widely known Larson-Miller Parameter was used:

L-M Parameter = T (c + log t)

T = absolute temperature, °R
t = time, hours
c = constant = 20 for Rene 80

The results for uncoated and coated specimens 0.038 cm, 0.075 cm and 0.15 cm thick are shown in Figs. 13, 14 and 15, respectively. Standard size bar data are included in Fig. 13. Average data for each heat is shown separately, and the 1093C, 34.5 MN/m² data from the earlier work [1] is plotted also. The data for the lower temperature tests in the earlier work are not used here, since they were not obtained from dead weight loaded tests, and are not considered as reliable as the new data. These curves may be used as average design values for the appropriate thicknesses. The coated specimen curves are lower than the uncoated for all conditions for the 0.038" thickness. The thicker specimens show the coated curves crossing over the uncoated curves at a parameter value of approx. 52×10^3. Standard bar coated and uncoated data was combined, since no significant differences existed between them.

The effects of section size and of exposure of approx. 1000 hours at 899C and 982C on rupture life are more clearly shown in Figs. 16 to 21 for each of the different test conditions. In these curves, the rupture life is plotted as a percentage of the standard size bar life for the particular heat and test condition as a function of the actual average thickness. For every test condition, the thinner the specimen, the lower the life, without exception. On this scale, the magnitude of the coating effect is more easily seen. The reduction in life due to the coating is greatest at 760C and for the higher stresses at 871C and 982C. At 1093C, the coating is obviously beneficial for the thicker specimens only.

Exposure at 982C was most harmful to the thinner sections at the lower temperatures. The exposures had virtually no effect on the 0.15 cm specimens. The 899C exposure was less detrimental than the 982C exposure.

The variation of average rupture elongation with thickness is shown in Figs. 22-27 for the different test conditions. In all cases, the elongation of the 0.038 cm specimens is lowest. There is little difference between the 0.075 cm and 0.15 cm thicknesses, uncoated, although at 1093C the 0.15 cm elongation seems unusually low (as is the standard size bar elongation). The coating lowers elongation at 760C and for the higher stress at 871C; otherwise the effects are slight. The 982C exposure lowers elongation most for the thinner specimens below 871C, while for the 0.15 cm thickness, the elongation is lowered at 760C and increased at 871C and 982C. Exposure at 899C had lesser effects on elongation.

In evaluating the effects of the various SM conditions on stress rupture properties, the section thickness effect must be taken into account, even more than in tensile properties because of the great variation of life with thickness. The "corrected" properties of the SM specimens are listed in Table III with average AC values for comparison. Adjustment of the rupture lives to a nominal 0.075 cm thickness was made using ratios of lives from the curves of Figs. 16, 18, 20, and 21 and ratios of elongation from Figs. 22, 24, 26, and 27 determined by the same method as used for tensile properties.

At 760C and 871C, both uncoated and coated SM samples had lower lives than the AC samples, while at 982C and 1093C the SM were equal to or better than the AC. Except at 760C, the 982C exposures lowered lives. After exposure, re-SM lives are additionally lowered, even with the correction for thickness. Re-heat treating and coating after the second SM lead to lower properties at 760C, but produced no significant changes at the other temperatures. The second exposure further lowered life at 760C, with little effect at higher temperatures.

Elongations of SM uncoated and coated specimens were higher than the corresponding AC specimens at the three lower temperatures, and slightly lower at 1093C. Exposures lowered elongations except at 1093C. After exposure, SM samples always had equal or higher elongations than exposed AC samples. Re-SM after exposures and re-treatment plus coating after re-SM, did not prove to be beneficial. The second exposure, unlike the initial exposure, did not lower elongation.

The stress rupture properties of the coarse grain (CG) 0.075 cm specimens are compared to the average normal grain 0.075 cm specimen results in Table IV. At 871C, the CG lives are significantly lower, both uncoated and coated, while at 1093C, the CG lives are slightly higher. Elongations for CG samples were similar to normal samples at 871C and lower at 1093C.

Thermal Fatigue Tests, Rène 80

The thermal fatigue tests on 4 specimens each in the as-coated and coated and exposed at 982C conditions produced no cracks or failures in 4000 cycles, at which point the test was terminated. The same results were obtained on the earlier heat.[1]

High Cycle Mechanical Fatigue, Rene 80

For comparison with the Rene 120 results, presented later, the 10^7 cycle reversed bending fatigue strengths at R.T. from the original work[1] are listed below:

As cast, uncoated	372 MN/m^2	(54.0 ksi)
As cast, Codep B-1 coated	359 "	(52.0 ")
As cast, coated, exposed 989 hrs, 982C	331 "	(48.0 ")

Stresses were calculated based on external dimensions in all cases.

Tensile Tests, Rene 120

The data for Rene 120 is presented in the same manner as used for Rene 80. Figs. 28-33 show the variation of Y.S., U.T.S. and elongation with temperature for the 0.075 cm and 0.15 cm thicknesses, uncoated, Codep B-1 coated and coated and exposed ~1000 hours at 982C. They can be used as average design curves for those thicknesses. The points are the averages of all the specimens from both heats tested, and are generally 8 tests per point. Figs. 34-36, for each heat, plot the data as a function of section thickness, and include the standard bar averages.

The coating lowers strengths at almost all conditions at the lower temperatures with the greatest reduction for the 0.075 cm specimens in U.T.S. (about 12%). At 982C, the differences become very small. Elongations are lowered at all test conditions, with a continual decrease in elongation with increasing temperature for the coated specimens. Again, the decreases are greater for the thinner samples. The limited number of 0.038 cm specimens (1 test per condition) have lower properties than the thicker samples, and the coating lowers all properties.

The 982C exposures caused additional losses in Y.S. of 22% and 15% and in U.T.S. of 29% and 23% for the 0.075 cm and 0.15 cm samples, respectively at R.T. At elevated temperature, the effects of exposure were much smaller, especially for Y.S. where the losses almost disappear at 982C.

To evaluate the effects of surface machining (SM) on tensile properties, the thickness was taken into account by the same method used for Rene 80. The average thickness after the first SM was 0.0652 cm with one bare specimen at 982C at 0.0572 cm and after the second SM was 0.0422 for Rene 120. Since only single points at the 0.038 cm thickness were available, the corrections cannot be considered as accurate as for Rene 80, although the actual values were similar in most cases. There were no tests for 0.038 cm Rene 120 at 982C, therefore the same corrections were used as for Rene 80. The values corrected to the 0.075 cm thickness are listed in Table V, together with the average as-cast specimen values.

The SM specimens at R.T. and 871C in the uncoated, coated and coated and exposed conditions had about equivalent strengths to the AC specimens (higher strengths at R.T. after exposure) and equivalent or higher elongations. At 982C, the SM samples had somewhat greater strengths before exposure, and similar strengths after exposure. Elongations were again equivalent or higher for the SM samples. Compared to the exposed values, re-machining after exposure improved U.T.S. and improved (or did not change) elongation at all temperatures. The Y.S. was lowered at R.T. and raised at the elevated temperatures. Re-heat treating and coating improved Y.S. at R.T., but lowered strengths at all other conditions. Elongations were not greatly affected by the re-treatments. As with Rene 80, the second exposures produced drastic reductions in all properties, and resulted in the lowest values of any condition (except for elongation at 982C).

The coarse columnar grain (CCG) 0.15 cm specimen tensile results are in Table VI, with the average properties for the normal grain size specimens for comparison. The room temperature U.T.S. values, coated and uncoated, are 10% lower for the CCG samples. The Y.S. and elongation are lower also. At 871C the CCG samples are all equal to or better than the normal specimens in all properties.

Stress Rupture Tests, Rene 120

The presentation of Rene 120 data is made the same way as for Rene 80. Figs. 37-38 are Larson-Miller Parameter plots of the data for 0.075 cm (with standard bar data) and 0.15 cm specimens, respectively. Averages for each heat are shown, and the curves for the flat specimens may be used for average design values. Coated specimens are noticeably lower than uncoated specimens at the lower parameters, with the difference virtually eliminated by a parameter value of 53 (the 1093C data).

The percent of standard bar life for each heat as a function of thickness is shown in Figs. 39-42 for the four test temperatures. For coated and uncoated samples, the 0.15 cm specimens always had greater lives than the 0.075 cm specimens, with the difference (slope of line) decreasing somewhat at the higher temperatures. The more detrimental effect of the coating at the low temperature, Fig. 39, compared to the higher temperatures is clear.

The ~1000 hour exposures at 982C lowered lives at 760C, 871C and 982C, but not at 1093C. The exposure had a greater effect on the thicker specimens at the 760C and 871C, 379 MN/m^2 conditions. At other conditions the slopes of the exposed specimens were close to those of the coated specimens. Where two stresses were run at the same temperature (871C and 982C), the dependence of life on thickness was greater for the lower stress (longer times).

The few points for the 0.038 cm thicknesses are plotted (single tests), and all indicate much lower lives than the 0.075 cm specimens.

Stress rupture elongations as a function of section thickness are displayed in Figs. 43-45. At 982C, the thicker uncoated samples have higher elongation; at other temperatures there is little difference, or lower elongation for the thicker pieces. Coating lowers elongations, except at 1093C. Standard bar elongations are not much different from the flat section values, possibly lower at 1093C and higher at 871C. The 982C exposures resulted in lower elongations for the 0.15 cm specimens than for the 0.075 cm specimens in all cases. At 760C and 871C, the exposed values were slightly lower than the unexposed, coated values, while at 982C and 1093C, the exposed values were not much different, or possibly slightly higher.

Consideration of the effects of SM on rupture properties must include corrections for thickness. However, the number of specimens of Rene 120 thinner than 0.075 cm are much too few to provide the correction factors. Since these "corrections" are probably largely geometric rather than metallurgical, and since the correction factors for tensile properties were close for Rene 120 and Rene 80, the Rene 80 correction factors were applied to the Rene 120 rupture results. The reliability of the corrected values will be less than for Rene 80, but should be sufficient for reasonable comparisons.

The "corrected" lives and elongations for the SM conditions, together with average AC values are included in Table VII. The SM uncoated lives at R.T. and 871C were lower than corresponding AC lives, and at 982C and 1093C were equal or higher than AC lives. SM coated and coated and exposed specimen lives were lower at R.T. and higher at all other temperatures. Elongations of SM specimens were equivalent or better at all conditions than for AC specimens. Re-machining after exposure permitted some increase in life at 871C and 982C, a slight decrease at 1093C and a noticeable decrease at 760C. Re-treating and coating after the second SM lowered lives at all conditions, and the second exposure at 982C further lowered lives to the lowest values encountered. Elongations for the three conditions after the second SM were always greater than the corresponding elongations after the initial SM, except after coating and exposure for the 760C test temperature.

The coarse columnar grain rupture properties of the 0.15 cm Rene 120 specimens are compared to average normal grain size properties in Table VIII. Results for rupture life are similar to those found for Rene 80: the coarse grain is poorer at 871C and better at 1093C, and coating is detrimental at the lower temperature and beneficial at the higher temperature. Elongations at all conditions are equivalent for both grain sizes.

Thermal Fatigue Tests, Rene 120

As with Rene 80, no thermal fatigue failures occurred up to 4000 cycles for the 4 coated and coated and exposed samples. However, one as-coated specimen had a small longitudinal crack at the end of the sample which appeared after 2000 cycles and did not grow further. This is an unusual location and direction for a thermal fatigue crack.

High Cycle Mechanical Fatigue Tests, Rene 120

Specimens of the 0.075 cm thickness from both heats of Rene 120 were tested in reversed bending fatigue in the uncoated, coated and coated and exposed at 982C conditions, with 4 points at each condition for each heat. The average 10^7 cycle fatigue strengths are:

As cast, uncoated	269 MN/m^2	(39.0 ksi)
As cast, Codep B-1 coated	269 "	(39.0 ")
As cast, coated, exposed 997 or 822 hrs, 982C	286 "	(41.5 ")
Surface machined, coated	331 "	(48.0 ")

Stresses were calculated based on external dimensions in all cases. The coating and exposure were not detrimental, and the surface machined specimens (only 2 points for Heat B415) seemed better than the as cast specimens.

METALLURGICAL ANALYSIS

Microstructure, Rene 80, Uncoated

The structural features of the as-cast and of the heat treated Rene 80 specimens are essentially the same as noted in the previous work.[1] The following lists the important items:

As-cast: 1. Primary MC carbides (M = Ti, W, Mo) are generally chunky and uniformly distributed right up to the outer surface (Fig. 46a).

2. The MC carbide size is greatest in the thickest specimens, the standard size bars, and finest in the thinnest specimens, 0.038 cm thick.

3. Unetched, or with a special carbide etch, only the MC type carbides are visible.

4. After a general structure etch, coring is evident, and the same relative dendrite and grain size exists: larger dendrites and coarser grains in the thicker pieces (Fig. 46b).

5. Eutectic γ' nodules are present; coarse in thick castings and very difficult to find in the thinnest (Fig. 46b).

6. The fine γ' has about the same size and distribution in all castings.

Heat-treated: 1. MC carbides are the same as before, except for a narrow band near the surface, where their concentration is lower, especially in the thinner pieces.

2. A carbide etch now shows fine particles of $M_{23}C_6$ (M = Cr, Mo) in the grain boundaries (Fig. 46c).

3. A fine, recrystallized grain layer exists at the surface (about 0.001 to 0.002 cm deep), with $M_{23}C_6$ in the fine grain boundaries (Fig. 46c).

4. Coring has decreased greatly, and the eutectic γ' has disappeared (Fig. 46d).

5. The fine γ' is somewhat coarser than in the as-cast specimens, and is uniformly distributed. A very thin (~0.0004 cm) de-alloyed surface layer is formed in which no γ' is visible.

Surface machined: 1. The surface profile is smoother than the as-cast surface.

2. MC carbides exist closer to the surface than in the as-cast and heat treated samples. The surface machining is performed after the two initial high temperature treatments.

3. There is no fine recrystallized grain layer or de-alloyed layer at the surface, otherwise the structure is identical to the as-cast specimens.

The coarse grain Rene 80 specimens (no nucleation coating on mold) had about double the average grain size of the normal 0.075 cm specimens. In all other respects their structure and behavior was the same.

After the stress rupture tests at elevated temperatures, the following changes in the structure occur:

1. With increasing time and temperature (up to 982C), the grain boundary $M_{23}C_6$ particles coarsen and become more continuous. At 1093C, they disappear (solution temperature is ~1065C).

2. Surface oxidation increases with increasing time and temperature, as does the depth of a de-alloyed surface layer. The de-alloyed layer grows to 0.006 cm after 70 hours at 1093C (below the remaining metal surface). The total oxidation attack plus de-alloying represent a considerable reduction in the effective cross section of a thin specimen.

3. The γ' is coarsened at the 982C and 1093C test temperatures, and the grain boundaries become wider, possibly with γ' in addition to the $M_{23}C_6$ carbides (Fig. 46g).

Microstructure, Rene 80, Coated

The coating produced by the Codep B-1 process shows an additive layer of NiAl with some embedded αAl_2O_3 particles, and a fingered diffusion zone, Fig. 46e, (fingers are σ phase, with MC carbides and NiAl often present). The relative thickness of each is about the same. For a total coating thickness of 0.005 cm, the additive layer would be 0.0025 cm thick. However, it should be noted that the actual measured increase in thickness for such a coating is only 0.0015-0.002 cm, indicating that some of the "additive" layer is actually formed by diffusion into the substrate material.

Below the diffusion zone, and visible only after etching, is a very narrow band, ~0.0002 cm wide, of finer acicular, or fingered phases. The internal alloy structure is the same coated or uncoated, since the heat treatment cycle is identical. The coating on a surface machined sample appears the same as on an as-cast sample except for the smoother profile.

The following changes in and below the coating occur after elevated temperature stress rupture testing or after long time exposure:

1. Exposure of ~1000 hours at 899C, or tests of about 100 hours at 982C produce equivalent structures: the diffusion zone fingers coarsen, the acicular phase below the diffusion zone grows, and the internal γ' and $M_{23}C_6$ particles and grain boundaries grow somewhat.
2. The 982C exposure results in greater changes of the same type noted in (1). The sub-diffusion zone needles grow up to 0.003 cm deep. In addition, narrow streaks and small patches of the additive layer transform from NiAl to γ' (dark to light color), Fig. 46f, g. The size and amount of MC throughout the section decrease somewhat.

3. Rupture tests at 1093C eliminate the sub-diffusion zone needles and the $M_{23}C_6$. Diffusion zone agglomeration and additive layer transformation increase greatly with time. In the grain boundaries coming down below the coating, coarse γ' particles are formed to a depth of up to 0.006 cm.

The re-surface machining performed on the 0.075 cm, 982C exposed samples was deep enough to remove all the visible microstructural changes under the coating. Re-heat treating and coating these pieces, while producing a new-looking coating, further reduced the size and amount of MC present in the section. The grain boundaries and γ' were restored to "new" appearance. Testing and exposure of the re-treated specimens produced the same changes that the initial tests or exposure did. However, the earlier decreases in MC coupled with the further losses after re-exposure have almost eliminated the MC. The acicular phases extend below the coating, as after the initial exposure, but because of the reduced section thickness now effectively occupy a greater percentage of the area.

Coating Composition, Rene 80

In addition to the microstructural observations, X-ray diffraction on the exterior surface and electron microprobe scans on cross sections of coated samples were performed before and after various tests and exposures.

Diffraction on as-coated specimens showed the normal additive layer NiAl BCC structure with a lattice parameter of 2.887Å, and a small amount of αAl_2O_3 (the embedded particles). No γ or γ' could be detected. The 899C exposure reduced the NiAl parameter to 2.877Å, with no new phases appearing. The 982C exposure lowered the parameter to 2.874Å, decreased the intensity of the NiAl lines, increased the αAl_2O_3, and γ-γ' appeared with a parameter of 3.590Å (matrix parameter = 3.578Å). After stress rupture testing at 1093C (46 to 110 hours), the NiAl parameter dropped to 2.871Å and its intensity was lower than after the 982C exposures. The amount of αAl_2O_3 decreased and the spinel $NiAl_2O_4$ appeared. More γ-γ' (a_o = 3.581Å) appeared than after the 982C exposures

The compositions of the coatings on various thicknesses and on a surface machined specimen and an SM specimen exposed, re-SM and re-coated were very similar, allowing for the slight differences in coating thickness (several coating runs were represented). The principle coating elements and the ranges encountered are:

Aluminum	29-34% (max. values)
Chromium	High at diffusion zone, decreasing to <2% at surface
Titanium	High at diffusion zone, decreasing to <0.5% at surface

Other elements from the Rene 80, Co, Mo and W are present and gradually decrease towards the surface. Nickel is the balance and is rather uniform across the additive layer. The diffusion zone consists of Cr-Mo-W peaks (σ phase), Ti-Mo-W peaks (MC) and increasing Al content.

The 899C exposure reduced the peak Al level to 21% in the additive layer, while raising the Al level in the base alloy to a depth of 0.002 cm below the previously affected zone. With the loss of Al in the additive layer, the other elements increased, with Cr going up to 2-1/2% and Ti to 2% at the surface. Exposure at 982C produced greater effects of the same nature. Al content dropped to about 16%, Cr increased to as much as 5% and Ti to 3%. The depth of the affected zone below the coating was up to 0.004 cm. The diffusion zone phases remained the same, although size and distribution changed.

Fracture Structures, Rene 80

Generally, the fracture surface appearance was the same for each test condition regardless of the specimen condition. Tensile tests up to 982C were mostly intergranular, but many had some portion of the fracture transgranular. Stress rupture tests at 760C and 871C resembled the tensile tests. At 982C and 1093C, the oxidation attack on the fracture made reliable determination difficult, but it seemed that fractures were largely intergranular. Secondary cracks, found frequently at 871C and above, were always intergranular.

The surface of uncoated samples, tensile tested at 760C and 871C had numerous cracks in the boundaries of the fine recrystallized grains. These cracks did not seem to propagate further, although it is possible that the primary fracture initiated at one. The same type of cracks were probably present on the surface of the rupture tests, but the oxidation attack covered them.

Many cracks were formed in the coating of coated samples tested at R.T., Fig. 46f, occasionally at 871C and never at 982C. The coating cracks extended at times to the bottom of the deepest needles, but not into sound base alloy. Stress rupture testing at 760C always produced coating cracks similar to the tensile test cracks, while they were rare at 871C and absent at the higher temperatures. The coating behaves quite ductilly at high temperature and would neck down over base alloy cracks rather than crack. The depth of coating cracks in exposed samples was usually greater than in unexposed samples because of the greater depth of the acicular phase below the coating.

No differences in behavior were noted for the surface machined specimens.

Microstructure, Rene 120, Uncoated

The microstructure of René 120, as-cast, resembles that of Rene 80, with the following differences:

1. The MC, particularly in the thinner sections, tends to have a eutectic appearance frequently, Fig. 47a.
2. There is more, and larger eutectic γ' at each comparable section size.
3. More porosity was found in the Rene 120 castings than in Rene 80.

The heat treatment for Rene 120, unlike Rene 80, does not eliminate the eutectic γ', Fig. 47b, although "diffusion zones" appear to form around them. The grain boundary particles are finer in Rene 120, and the normal γ' rather more irregular in size. Other features are the same as for Rene 80, as is the effects of surface machining.

The changes in structure after stress rupture testing are the same as for Rene 80, except that the depth of oxidation attack is much less. The grain boundary particles which dissolve in Rene 80 do not disappear in Rene 120 at 1093C.

Microstructure, Rene 120, Coated

The structure and behavior of the Codep B-1 coating on Rene 120 are entirely like on Rene 80. The "fingers" in the diffusion zone tend to be somewhat finer in Rene 120 (Fig. 47c).

Testing and exposure produce the same changes as in Rene 80, with a few exceptions:

1. The sub-diffusion zone needles, which dissolve in Rene 80 at 1093C tests, do not dissolve in Rene 120. They do decrease in number and grow in length, Fig. 47d.
2. Interdiffusion effects between the coating and substrate are of the same nature as in Rene 80, but the depth of effect is less for corresponding times/temperatures.
3. The eutectic γ' nodules, which are absent in Rene 80, appear to be affected and may be breaking up after long exposures, particularly for the SM, coated, exposed, re-SM, re-heat treated, re-coated and exposed specimens, Fig. 47e. They do not disappear entirely, even after the most severe conditions seen here.

The depletion of MC and the reduction of sound remaining metal after the exposure and re-exposure at 982C remain as with Rene 80.

Coating Composition, Rene 120

The coating structure, as in Rene 80, is NiAl with entrapped αAl_2O_3 particles. The initial lattice parameter of the NiAl is 2.884Å, and it decreases to 2.872Å after 982C exposure or 1093C testing (66-78 hours). Exposure at 982C reduces NiAl intensities and produces γ-γ' indications, a_0 = 3.589Å, as do the 1093C tests, a_0 = 3.584Å. The base alloy parameter is close to these values: 3.588Å. As with Rene 80, the 1093C tests produced some spinel, $NiAl_2O_4$ and increased αAl_2O_3.

Microprobe analyses indicated similar elemental variations in the as-coated condition. Al was slightly lower, 25-29% maximum levels, Cr decreases to <1% at the surface and Ti and the other elements are relatively as they are in Rene 80. While Al was slightly lower, as-coated, than Rene 80, after 982C exposure it was slightly higher, 17-20% maximum. Cr increased to about 4% and Ti to 1-1/2% (both lower than in Rene 80). The maximum depth of additional inward diffusion of Al and other compositional changes was about 0.0025 cm, also less than in Rene 80.

Fracture Structures Rene 120

In all aspects, the fracture surface appearance and surface and coating cracking behavior was the same as in Rene 80. The large γ' eutectic nodules were cut through by the primary fractures sometimes, but at other times the fracture passed close to the nodules without effect.

The fatigue fractures at R.T. were typically crystallographic for the uncoated and coated samples. The coated and exposed sample, at least at the section inspected, resembled a tensile fracture with probably mixed trans- and inter-granular segments.

DISCUSSION OF RESULTS

Effect of Section Thickness on Tensile Properties

For uncoated Rene 80 and Rene 120 at R.T., there was not a great difference ($\leq$4%) in yield strength between the different thickness specimens. At 871C the 0.038 cm thickness yield strength was highest for Rene 80 while the differences between the other thicknesses were small ($\leq$5%) for both alloys. Thicker specimens had higher Y.S. at 982C. For U.T.S., the effects of thickness are more noticeable. At R.T., the 0.075 cm thickness was strongest, higher than standard bars, while at and over 982C, the thicker specimens were higher, and the standard bars highest. The Y.S. for the 0.038 cm Rene 80 specimens were relatively greater than their U.T.S. at all temperatures. Elongations were highest for the thickest specimens in all cases except at R.T. for Rene 80, where the 0.075 cm thickness had a higher value than the 0.15 cm thickness (and the standard bars). Generally, the elongations of the 0.075 cm and 0.15 cm samples were comparable to the standard bars, while the 0.038 cm samples were considerably lower.

Considering that the thinner specimen has finer grain size, and that fine grain material should have greater strength at low temperatures (below the equicohesive temperature, ~850-950C for these cast alloys), the fact that the thinner specimens did not show a greater Y.S. indicates a marked geometry effect is countering the grain size effect (see Discussion in Ref. 1). Once beyond the Y.S., the specimen's ductility will affect its further load carrying capacity, therefore a low ductility will result in a small difference between Y.S. and U.T.S. In most cases the elongations of the 0.075 cm and 0.15 cm specimens were not greatly different, and their U.T.S. values were close. However, for Rene 80 at R.T., where the 0.075 cm specimens had much greater elongations than the 0.15 cm specimens, the U.T.S. was correspondingly higher. Similarly, the 0.038 cm thickness had much lower elongations than the thicker pieces, and their U.T.S. were low, in spite of closeness of their Y.S. relative values. In fact, the sharp break in elongation between the 0.075 cm and 0.038 cm sizes may mean that a change in failure mechanism has taken place. Once a crack is initiated in the 0.038 cm piece, its propagation is more rapid than in thicker pieces.

Effect of Surface Machining on Tensile Properties

Surface machining eliminates the fine recrystallized surface grains and the MC depleted surface zone. Below 871C this is no advantage, with strengths being similar to as-cast surface specimens. At 982C, the carbides are more important for strength, and the fine surface grains are deleterious; therefore elimination of the fine grains and the de-carbided area by SM removes "weak" material and results in higher strengths.

Effect of Grain Size on Tensile Properties

The few tests on the coarser grain (CG) 0.075 cm thickness for Rene 80 at R.T. and 871C show higher Y.S. for the coarser grain material, but lower U.T.S. in spite of the equivalent elongations. The coarse grains span the entire thickness at many sections, except for the fine recrystallized surface grain layer on each side. Since metallographic evidence indicates cracks initiate in these fine surface grain boundaries (especially at 760C and over), it is likely that both normal and coarse grain specimens become cracked at similar times after yield. The CG specimens, having no longitudinal or angled internal grain boundaries to hinder slip, will then behave weaker than the finer grain specimens with a resulting lower U.T.S.

For Rene 120, the thickness investigated was 0.15 cm. At R.T., all properties were lower for the coarse columnar grain (CCG) specimens, while at 871C the strengths of the CCG specimens were equivalent to normal specimens even though ductility of the CCG specimens was higher. The slight differences in behavior of these specimens compared to the Rene 80 CG specimens may be due to the difference in thickness and to the grain shapes: equiaxed for the Rene 80, long and columnar (transverse to test section) for Rene 120.

Effect of Coating on Tensile Properties

In discussion of coating effects on properties, it should be remembered that the strengths described thus far have been based on the original specimen thickness before coating and that the added coating layer and the diffusion zone are weak and very brittle at low temperature (to about 871C) and weak and ductile at 982C.

On the above mentioned strength basis, the coating lowers Y.S. most at R.T. and less or not at all by 982C and over for both alloys in the as-cast or surface machined condition. However, when the strength is based on the actual unaffected metal below the diffusion zone, there is no loss at R.T., and a benefit at high temperatures. The benefit is due to the oxidation protection of the coating which prevents the intergranular attack - which would otherwise occur. This is of greater aid to Rene 80, which has poorer oxidation than Rene 120. The same observations apply to U.T.S. for both alloys.

The brittleness of the coating and the diffusion zone fingers result in many surface cracks at temperatures under 871C. This ensures that the only load carrying material is the metal below the diffusion zone. Fortunately, the base alloys are not notch sensitive and these coating cracks do not progress any further unless by chance they happen to line up directly with a matrix grain boundary. In that case, some reduction in elongation would occur. For this reason, the average coated specimen elongation for coated Rene 80 is occasionally somewhat lower than for the bare specimens. Rene 120 seems more sensitive to the coating (it has lower basic ductility than Rene 80) and coating always lowers elongation.

Effect of Exposures on Tensile Properties

Since both alloys would normally be used in the coated condition at the exposure temperatures investigated, exposures were performed only on coated specimens.

One heat of Rene 80 at 0.075 cm thickness was exposed at 899C. Long time exposure at this temperature precipitates additional $M_{23}C_6$ in the grain boundaries, becoming nearly continuous. The γ' is coarsened somewhat and the σ needles below the diffusion zone grow and occupy a larger portion of the cross section. Diffusion of Al from the coating down grain boundaries produces thicker and possibly continuous γ' near the surface. All these effects act to lower ductility at lower temperatures. Elongations at R.T. and 760C after the exposure are reduced by over 4:1 and are under 1% in value. The effect of exposure on elongation decreases at 871C, and has no effect at 982C (over the exposure temperature).

The loss produced in Y.S. is 20% at R.T., but the lower Y.S. coupled with much poorer ductility results in a U.T.S. loss of 38%. As the temperature increases, the loss in Y.S. is accompanied by an increase in elongation, which results in smaller relative losses in U.T.S. Once the test temperature exceeds the exposure temperature, no loss in strength occurs. If strength values were based on the "sound" metal under the deepened diffusion zone, the low temperature strength values would be greater.

The ~1000 hour exposures at 982C produce similar metallurgical reactions in Rene 80 as did the 899C exposure, with different relative effects. The grain boundary $M_{23}C_6$ tends to agglomerate into coarser particles than the near continuous film formed at the 899C exposure. This coarser distribution is less harmful to ductility. At the same time, the depth of acicular phases and Al diffusion under the coating increases, the γ' coarsening is greater and MC decreases in amount. The net effect for the 0.038 cm specimens is that the cross-sectional losses and increased brittleness at the surface outweigh the internal grain boundary improvements. Both strength and elongation suffer losses at all test conditions. For the 0.075 cm thickness, where the surface effect is a much smaller contributor, the losses in strength and elongation at R.T. and 760C are less than caused by the 899C exposure. At 871C and higher there are no losses in Y.S., while the loss in elongation is maximum and results in lower U.T.S. as well. As before, when the test temperature is the same, or higher than the exposure temperature, no losses in properties are seen. The thickest specimens had the smallest overall property losses, particularly in elongation at 760C and 871C, confirming that the surface changes do have an effect. Actually, the strengths of the 0.15 cm specimens showed no losses at temperatures above 871C.

The metallurgical changes in Rene 120 caused by the ~1000 hour, 982C exposure are coarsening of the γ', some break-up of the eutectic γ' nodules, a decrease in MC and some grain boundary precipitation (less than in Rene 80), with identical sub-coating changes. The relative losses in elongation were less than encountered in Rene 80 (initial elongations of Rene 120 are lower than Rene 80), which may be related to the lesser amount of grain boundary phase. While the reductions in strengths for Rene 120 were about the same as Rene 80 at R.T., and are decreased as temperature is increased, the strengths did not regain pre-exposure levels at 982C. The major cause for lower strengths should be the coarsening of the γ', as with Rene 80. However, the action of the γ' is more sluggish in Rene 120, due to the presence of Ta and the greater γ' volume percentage. The short time at temperature before and during the tensile test (usually 15 minutes or more) at 982C may have been sufficient to degrade unexposed Rene 80 capability by quickly altering the γ' (and possibly grain boundary phases) to a state closer to that produced by the long time exposure. The slower reactions in Rene 120 may prevent the initial degradation thereby enlarging the apparent effect of the exposure. It should be noted that Rene 120 at temperatures above 871C always has greater strengths than Rene 80, unexposed or exposed.

<u>Effect of Exposure and Re-Treatments on Surface Machined Specimens Tensile Properties</u>

The 982C exposure had identical effects on surface machined specimens as on as-cast surface specimens, except possibly for Rene 80 at 982C, where exposure lowered strengths of surface machined but not as-cast specimens.

Machining-off the coating and exposure-produced acicular phases below the coating should eliminate that part of property changes due to the brittle phase at the surface and the effective reduction in load carrying section. The re-machined specimens are appreciably thinner, which was taken into account by the previously described correction factor based on as-cast, unexposed samples. The fact that the correction is based on rather limited data and on a different condition reduces the accuracy of the "corrected" values. It should be kept in mind as well, that the purpose of "correction" is to permit comparisons of basic property changes, ignoring the geometry factors: on an absolute basis, the thinner re-machined specimens are at lower property levels in all cases for both alloys.

The most common effect of the re-machining was to permit some increase in elongation, which would be expected from the elimination of the brittle coating and acicular phase particles. The internal structural effects of the exposure are not avoided, and the improved ductility is not up to the pre-exposure level (except for the single case of Rene 80 at R.T.). Strength effects of the re-machining are variable. In most cases, for both alloys, strengths were about equal or increased. This should be due to the removal of the weak sub-coating diffusion region. In two instances, Rene 80 Y.S. at 871C and Rene 120 Y.S. at R.T., strength was lower after re-machining. Except for experimental or correctional errors, there is no explanation for the loss.

Re-heat treatment and coating after the re-machining produce several metallurgical changes. The grain boundary phases and γ' are dissolved and re-precipitated close to their original, unexposed morphology by the re-heat treatment. A surface de-alloyed layer and MC free region are produced which were not present after the original or re-surface machining. Specimens were not tested in the re-heat treated condition, but were re-coated before testing. The coating, which is a normal aluminide, again is brittle below 871C and its diffusion zone occupies a greater percentage of the now thinner specimen than it had initially. The internal changes all should act to increase ductility, while the coating acts to reduce it. The net change was beneficial in most cases, and elongation was improved although not to the levels before exposure. For Rene 120 at 871C, a slight drop in elongation occurred, which was less of a percentage loss than found at the same temperature for coating AC or new SM samples. At R.T., the re-coat showed a larger drop for Rene 80, but the re-machined condition had an unusually high elongation compared to the other conditions. The effects of the re-treatment and coating on strength are more difficult to predict. The return of grain boundaries and γ' to original morphology is beneficial; the decrease in MC should be detrimental at 982C; the re-coating itself should be slightly detrimental at low temperature and nearly neutral at 982C, except for the greater relative sub-coating region weakening. The noted changes actually produced were a lowering of strength for both alloys at 982C and for Rene 120 at 871C, while an improvement for Rene 80 occurred at R.T. and 871C. Rene 120 had higher Y.S. with lower U.T.S. at R.T., and is now lower in properties than Rene 80 at R.T. and 871C. The importance of each of the above mentioned factors is seemingly different for the two alloys. The evaluation of metallurgical changes in Rene 120 is complicated by the presence of the eutectic γ', which never disappears, even though the second treatment and coating tend to break it up further.

The effects of the second exposure at 982C were as expected: all properties were lowered, with relatively greater losses in strength than took place in the first exposure. With the additional MC loss and the coating sub-diffusion phase penetration, the remaining "sound" metal occupies only about 2/3 the thickness and that "sound" metal is virtually devoid of MC. The strengths of Rene 80 are lower than the strengths of the 0.038 cm specimens exposed (once) at 982C, especially at the higher temperatures where the MC has more importance (the twice SM specimens are about the same thickness as the 0.038 cm AC specimens).

Practical Evaluation of Effects on Tensile Properties

The areas of turbine blades in which tensile properties are of greatest interest are generally in the dovetails, shanks or roots, where temperatures are relatively low, usually well under 850C, and aluminide coatings not really required. Under these conditions, section thicknesses down to 0.075 cm having several grains per cross section show no strength or ductility problems. Down to 0.038 cm thickness, Y.S. is retained, but ductility and U.T.S. are lowered. No important differences exist between as-cast or machined surfaces down to 0.075 cm. If an aluminide type coating is employed, then some losses in elongation, Y.S. and U.T.S. will occur with thicknesses under 0.075 cm. If operating temperatures are as high as ~850C in these regions, then appreciable decreases in elongation and strengths can be expected at lower temperatures. Although no work here was done on reclamation of properties with exposure temperatures under 982C, re-heat treatment should be of greater benefit than after the 982C exposures. Any decrease in section thickness involved in the reclamation process should be taken into account.

Effect of Section Thickness on Stress Rupture Properties

For all test conditions of both alloys, with uncoated or coated specimens, the thinner the specimen, the lower the rupture life. Rupture elongation was lowest for the thinnest section, 0.038 cm, with little consistent difference between the 0.075 cm and 0.15 cm thicknesses or the standard size bars.

Many of the factors affecting tensile properties are involved in rupture properties. The finer grain size is an advantage at temperatures below the equicohesive temperature (~850C to 950C), while above this temperature, fine grains are not advantageous. At 760C, the rupture lives were relatively highest, averaging 55% of standard bars for 0.075 cm specimens and 90% for 0.15 cm specimens for both alloys uncoated. The fact that life was lower at 760C in spite of grain size benefits indicates that the geometry effect on fracture is of considerable magnitude. It should be kept in mind that a small loss in strength (say 6-10%) is sufficient to cause a large loss in life (50-67%) at 760C for Rene 80.

At temperatures above 760C, the fine grain size is weakening, and the surface attack on bare specimens becomes more important as exposure during test increases in length. The combination of these changes, the ductility variation with test temperature and stress, and possible differences in the geometry effect make prediction of life losses with test conditions difficult. Generally, relative lives decrease in the longer time tests at the same temperature (871C and 982C), which would be expected due to greater surface attack decreasing the load carrying area a larger percentage in thinner pieces. This was most noticeable in Rene 120 0.075 cm specimens. While all 0.075 cm and 0.15 cm samples had lower lives compared to standard bars at 871C and 982C than at 760C, the 0.038 cm samples had much greater relative losses, either indicating the relatively greater surface attack or that there is a mechanism change for this thickness as suggested in the discussion of tensile properties. At 1093C, the 0.038 cm specimens show a much higher relative life than at the intermediate temperatures. Rupture elongation is also higher, and may be the mechanism permitting longer life, or is itself an indicator of a change in the geometry effect. The great amount of surface de-alloying at 1093C may "ductilize" the surface which in turn may be more beneficial to life than the loss of strength in that region is harmful.

Effect of Surface Machining on Stress Rupture Properties

At 760C and 871C, for both Rene 80 and Rene 120, the surface machining lowered lives (more at 760C), while at 982C and 1093C, there was generally an increase in life. The removal of the very fine recrystallized surface grains would be expected to be detrimental at lower temperatures and beneficial at higher temperatures. The removal of the MC depleted layer is also of benefit at higher temperatures. At all temperatures, the SM specimens had equal or greater elongation than the AC specimens. This may be due more to production of a uniform non-tapered gage section than to any metallurgical effects.

Effect of Grain Size on Stress Rupture Properties

The coarser grained specimens of both alloys had lower rupture lives at 871C and higher lives at 1093C, bare or coated, than did the normal grained specimens, just as would be expected with the equicohesive temperature between the two test temperatures. Elongations were different only for Rene 80 at 1093C, where the CG samples had half the value of normal grained samples.

Effect of Coating on Stress Rupture Properties

The effects of coatings were more consistent on rupture properties, where surface protection is needed for the long test times. At low temperatures, where oxidation attack is least, the low strength and brittleness of the coating are harmful, and lives are reduced by ~50% (or more). When the remaining sound metal under the diffusion zone is used for the stress calculation, the extrapolated lives are not reduced at temperatures above 871C. At 760C the loss in life is greater than accounted for by the metal area and may be related more to the brittle, multiple cracking behavior of the coating. Another effect of the coating is to effectively eliminate the fine recrystallized grain layer, which is detrimental at 760C.

The coating is not brittle at high temperatures, and when stress calculations are based on sound metal, there is no loss in life at 871C and 982C. Moreover, the protection of the coating permits increased life at 1093C for both alloys and Rene 120 at 982C (lower stress), even when calculations are based on original metal thickness before coating. The same coating effects are noted on the CG samples of both alloys as on the normal samples.

Coating of SM samples produced the same effects in most cases as coating of AC material. At 760C for Rene 80, there was no apparent decrease in life, however uncoated SM specimens had rather shorter lives than expected, compared to the AC specimens. This is due to the loss of the fine grained surface layer. Rene 120 SM coated samples had a longer life at 871C than the bare ones. Except for this, the SM coated Rene 120 specimens had shorter lives at 760C and 871C than their AC counterparts, and equal or higher lives at 982C and 1093C, just as did the uncoated SM specimens.

The coating lowers rupture elongation in its brittle region, below 871C. The losses in rupture elongation are similar to losses in tensile elongation. The lowering of rupture elongation has a strong effect on rupture life: once a crack is initiated, failure occurs earlier. Diffusion of Al from the coating into grain boundaries near the surface undoubtedly reduces their deformation capability at low temperatures and contributes to the lower ductility. At temperatures above 871C, the coating and the grain boundary phase are ductile and no loss in elongation is seen. Perhaps there is a small increase in elongation due to the surface protection: oxidation and cracking at the outer surface are now blocked.

Effect of Exposures on Stress Rupture Properties

Unlike its reduction of tensile properties, the 899C exposure of Rene 80 does not change rupture lives very much, even at 760C and 871C. The $M_{23}C_6$ distribution produced is not embrittling at the much slower strain rate of the rupture test, nor is the additional sub-coating diffusion sufficient to be effective. The greater microstructural changes caused by the 982C exposures are sufficient to be damaging to Rene 80 at 760C to both rupture life and elongation. The losses are greater for the thinner sections, as would be expected if the greatest part of the damage is surface diffusion related. At temperatures above 982C, the losses are less for the thinner sections and do not affect the 0.15 cm thickness at all. It appears that at rupture test strain rates, the internal structural changes of the 982C exposure (coarser γ', coarser $M_{23}C_6$ at grain boundaries) are towards the direction of increased ductility and that for the thicker sections these changes outweigh the effects of the sub-coating acicular and grain boundary phases.

Rene 120 reacts differently than Rene 80 to the 982C exposure in one major respect: at low temperatures, 760C and 871C, the 0.15 cm thickness experiences greater property losses than does the 0.075 cm thickness specimens. Actually, there is almost no effect on 0.075 cm specimen properties. Only the thicker specimens drop in life and elongation at those temperatures, with no losses at 1093C. It is difficult to give an explanation for the lower values of elongation, and hence life, for the thicker sections after exposure. One possibility is that there must be a change in the fracture mechanism caused by the metallurgical changes which alter the balance of grain interior/grain boundary/surface properties.

Effect of Exposure and Re-Treatments on Surfaced Machined Specimens Stress Rupture Properties

The 982C exposures affected the SM specimens of both alloys the same way they affected the corresponding AC specimens. There is one notable exception, at R.T. for Rene 80, where the rupture life increased while the elongation decreased. It is true that the Al diffusing in from the coating can increase the amount of γ', which is a strengthening factor. However, the depth is small, and the effect was not noted on any of the other tests up to 982C. This point remains an anomaly - test scatter, perhaps.

Removal of the coating and its visible sub-diffusion zone should result in improved ductility at low temperatures. This is found for Rene 80 at 760C and for Rene 120 at 760C and 871C. At the higher temperatures, the coating permits greater elongation, and its removal resulted in lower elongations for both alloys at 982C and 1093C. With removal of the coating, rupture life was lowered for Rene 80 at all test temperatures and for Rene 120 at 760C and 1093C. There was an increase in life at 871C associated with an increase in elongation, but at 982C the life increased while elongation decreased. These values are based on thickness corrections for Rene 80, and they may not fit the true Rene 120 behavior well enough to rely upon too heavily. The general losses in life seem to show that the coating and/or diffusion zone after exposure has some contribution to rupture strength that is lost when it is removed. Without the "correction factor" applied to evaluate the properties at a constant thickness, the absolute values of the rupture lives after re-SM were much lower and for an actual engine part, where no "correction" exists, great losses will occur.

The metallurgical changes produced by the re-heat treatment and coating were mentioned in the tensile property section. Their net effect was to produce a basically weaker cross section at all temperatures and to lower ductility (due to the coating) at low temperature only. Essentially, this represents these results: (1) Lives decreased at all conditions for both alloys except at 871C for Rene 80, where there was no significant change. (2) Elongations dropped at 760C for both alloys and at 871C for Rene 120. (3) At higher temperatures, there were no changes for Rene 80 and some improvement for Rene 120. (4) The elongation effects of the re-coating were very similar to the effects of the original coating.

Re-exposure further lowered lives and elongations at 760C. All specimens failed on loading. This is not surprising, since the U.T.S. of Rene 120 had been reduced to below the rupture test stress, and of Rene 80 close to the test stress. Unfortunately, this was not known early enough to permit lowering of the rupture test stress. Rupture lives at all other temperatures were reduced (except Rene 80 at 871C, no change), while elongations were unchanged at 871C and increased at 982C and 1093C. The further reduction in MC, the greater percentage of load carrying area occupied by diffusion formed phases, and the coarsened γ' all act towards decreasing rupture life. The increases in Rene 120 elongation at the higher temperatures may be due in part to the continuing reduction in eutectic γ' nodules. In terms of absolute lives, the removal of the coating after exposure caused losses for every test condition, since the correction factors for thickness were between 1.4 and 4.0, depending on condition. The life losses, at the actual thickness used here, are great enough to present for serious consideration before "rejuvenation" by this method is applied .

Practical Evaluation of Effects on Stress Rupture Properties

Many turbine blades have thin sections (0.15 cm or under) and operate in the temperature region covered by the present work. Since reductions in rupture life of up to 50:1 may occur in coated thin sections, without exposure, it is clear that coatings and thin sections must be considered properly in design life calculations.

In the original program,[1] a table was made with proposed life percentages for two thicknesses of Rene 80 at various conditions. Much less data was available, and all tests up to 982C were run on lever arm type machines, which were deemed less accurate. Estimations were made to allow for the testing problems. At 1093C, where dead weight loading was used, the lives are close to the values measured in the present work. The effects at 760C were underestimated, and at 982C overestimated due to the lever arm machines used. However, because of the number of variables involved, a simple table is not sufficient for design use. The graphs presented in the present work, either the Larson-Miller plot, or the percent of standard bar curves should be used. The two alloys, while similar in relative properties in the uncoated condition, show a different sensitivity to coating and exposure, and a single presentation to cover the two alloys cannot be made.

It should be emphasized that all the data obtained is on a single shape thin specimen, and that turbine blades do not have long, narrow, thin shapes, run at uniform temperatures. Differing geometrical shapes may be expected to have differing effects, and the particular specimen used here probably has greater losses than would exist in usual blade shapes. Nevertheless, the magnitude of the effects that can happen are presented and make consideration mandatory.

For alloys other than Rene 80 or Rene 120, the percent of standard bar life curves can be used as a guide. The metallurgical structure and initial properties of any other alloy should be compared to these alloys to determine which is closest, or to interpolate between the two.

One of the methods for extending the life of blades which would otherwise be limited by coating degradation is stripping the coating (chemically or mechanically) and re-coating, usually with a re-heat treatment. The present work indicates that rupture lives are usually lowered, depending on exposure and test conditions. If the initial thickness is great enough, perhaps over 0.15 cm, the gain of the new coating life probably outweighs slight rupture life changes. However, with a thinner section, the removal of surface material can shift lives downwards due to the thickness alone. The total effect may reduce the efficacy of the rejuvenation process, and must be a consideration.

Thermal Fatigue Properties

The initial work on Rene 80 showed that the Codep B-1 coating was beneficial in thermal fatigue resistance for the specimen and cycle employed.[1] Uncoated specimens had cracked in 1935 to 2500 cycles, and after exposure had cracked in 1100-1500 cycles. Coated specimens, exposed at 982C (975 hours) or 1093C (483 hours) or unexposed did not fail up to 4000 cycles. The different heat used in the present work was tested coated and coated and exposed at 982C only (uncoated Rene 80 would not be applied at these temperatures). The results duplicated the earlier work, with no failures up to 4000 cycles. A more severe cycle, or greater restraint would be required to cause cracking in fewer cycles, but this would not have permitted direct comparison with the earlier tests.

Rene 120 did not fail up to 4000 cycles in the same test, coated, unexposed and exposed at 982C. There was one unusual crack not related to the thermal fatigue strains. Within the conditions of the test, there is no difference between Rene 80 and Rene 120 and it is likely that the coating is beneficial for Rene 120 as it was for Rene 80.

Mechanical High Cycle Fatigue Properties

The Rene 120 was tested for comparison with the Rene 80 tests performed in the original program.(1) Rene 120 proved to have lower 10^7 cycle fatigue strength at R.T. than Rene 80. This is to be expected, since the Rene 120 has about 6% lower U.T.S. uncoated, and over 10% lower U.T.S. coated and coated and exposed, combined with lower elongation for the thickness tested, 0.075 cm. The Rene 120 did not seem to be affected by the coating and exposure as was Rene 80. Perhaps its proportionately smaller losses in ductility due to coating and exposure are not enough to affect fatigue behavior. The higher value for the SM specimens (from a different heat than the AC specimens) may be due to the heat or to the improved ductility for the SM condition. At elevated temperature, where the strength of Rene 120 is greater than that of Rene 80, the relative fatigue strength of Rene 120 does appear better.

It should be noted that the stresses for the coated samples were based on external dimensions, unlike the tensile and rupture stresses. The calculation of a true metal strength is complicated by the different modulus of elasticity of the coating in addition to its added thickness. This is important because the alternating stress is actually applied as an alternating strain based on load-deflection characteristics of each specimen. Since there is no standard way of applying the data, the simplest method serves as a reasonable method of comparison.

Microstructural Effects

Many of the effects of the microstructural changes occurring during coating, exposure, etc. have been discussed in the preceding sections. A brief review will be presented here.

For Rene 80, after the normal heat treatment, the internal structure consists of fine $M_{23}C_6$ particles at grain boundaries, coarse MC particles relatively randomly distributed and a uniform distribution of γ' (no eutectic nodules). The surface has a very narrow de-alloyed band (loss of Al and Cr) and a somewhat wider band of none, or reduced MC particles. A fine recrystallized grain layer is also present at the surface if any surface working has been done (by grit blasting the as-cast surface, usually). The aluminide coating, when applied, eliminates the de-alloyed band and largely grows through the recrystallized surface grains. The MC-free region usually still exists. The coating itself consists of an "added" layer of NiAl, with small amounts of other base metal elements and entrapped αAl_2O_3 particles, and a diffusion zone with a finger-like structure of σ phase, NiAl, Ni_3Al, carbides, etc. Below the fingered layer is a fine layer of acicular phase (σ) and diffusion of Al into grain boundaries. The "added" layer is actually formed by some inward diffusion of Al into the original metal surface, plus Ni and other elements outward diffusion. The entire coating is very brittle up to temperatures of ~750C to 875C depending on its composition, and contributes little or nothing to the strength of the alloy. The increased Al below the coating does increase the γ', which may be sufficient to improve strength and decrease ductility. Surface machining after the first two heat treatment steps can remove all the de-alloyed band, the MC-free layer and the fine recrystallized grains.

Exposures or tests for successively longer times at higher temperatures coarsens the γ', increases the amount and continuity of grain boundary $M_{23}C_6$ up to a point, then agglomerate it and break its continuity, and eventually dissolve it (temperature over ~1065C). The $M_{23}C_6$ is formed at the expense of MC, which provides the C. The freeing of Ti from the MC permits formation of additional γ'. At the coating, diffusion occurs which lowers the Al in the coating, while the other elements increase, and raises the depth of internal Al penetration. The Al going inwards removes Ni from the matrix to form Ni_3Al, which leaves the remaining matrix unstable and additional σ needles are formed and grow. The σ already present in the diffusion zone grows and tends to agglomerate and lose the finger appearance. All σ dissolves by 1093C. If times are long enough, or temperatures high enough, the loss of Al in the added layer (by diffusion inwards and oxidation externally) will cause transformation of the NiAl to Ni_3Al. The 1000 hours at 982C reaches this condition. Of course, the hot corrosion resistance of the coating is lost when the NiAl disappears, but the coating increases in ductility. In addition to the inwards growth of acicular σ, the Al diffusing in along grain boundaries penetrates deeper and forms coarse γ' in boundaries and may promote formation of other phases as well.

Chemical or mechanical removal of a deteriorated coating after exposure allows the re-application of a new, sound coating. However, the depth of affected base alloy may be too great to remove without reducing the thickness substantially, and thus some of the diffused region will remain. If re-coating is accompanied by a solution treatment and age, the internal structure can be returned to its original state, except for the loss of MC, which is permanent. It should be mentioned that if insufficient material is removed before re-coating, then the coating will be applied on a different base composition than initially and may therefore have a different appearance and different properties.

Rene 120, with its higher Al and refractory element content produces more stable phases, in particular the eutectic γ' is not dissolved by the "solution" heat treatment, nor has it disappeared after the longest exposures run in the program. The sub-coating acicular σ phase is not dissolved entirely at 1093C, as occurs in Rene 80, nor do the internal grain boundary particles dissolve at this temperature. The latter particles, which resemble the $M_{23}C_6$ particles in Rene 80 may be carbides (they are etched by the carbide etch), but their composition is probably different. The lower Cr in Rene 120 and the higher W suggest they may be closer to an M_6C composition.

With the above observations in mind, the behavior of the coating and of the internal structure during surface machining, exposure, re-coating, etc. resembles that described for Rene 80.

One of the common problems of the two alloys is the loss of MC (and really of total amount of carbides) with time and temperature. The MC carbides are not dissolved by any short time heat treatment up to the melting point, and yet they disappear at lower temperatures. In most nickel-base superalloys, the MC is not the thermodynamically stable carbide. The very small amount of C in solution in the matrix is removed by formation of $M_{23}C_6$ or other stable carbide internally and by oxidation at an external surface (during heat treatment). Apparently, the MC reacts with the surrounding material that is below saturation levels of C, and gradually releases C to the matrix. As long as C is being removed from the matrix elsewhere, the loss of MC continues. As the MC is not the stable carbide at operating temperatures, it is not usually re-formed by heat treatments or during use. Some of the carbon appears as MC in the diffusion zone of the coating, which always is richer in MC than the area below. Effectively, this and any C lost to the surroundings are lost as strengtheners, and the alloys do depend on the carbides for strength at higher temperatures.

CONCLUSIONS

Tensile Properties

Average property curves for Rene 80 and Rene 120 are presented as a function of temperature and section thickness. From them, the following general effects are noted:

1. There is little effect of section thickness on 0.2% yield strength. All thin sections have lower values than standard size bars, except at 871C. The beneficial effects of fine grain size of thin sections compared to standard size bars at the lower temperatures are offset by the geometry effect.
2. Ultimate tensile strengths increase with increasing thickness at all elevated temperatures. At. R.T. the 0.075 cm thickness has the highest value.
3. Elongations for 0.075 cm and thicker sections are equivalent at all temperatures; for 0.038 cm sections, elongations are lowest.
4. The aluminide coating lowers the yield and ultimate strengths up to 871C and is beneficial at 982C and above. The strength losses at lower temperatures are greater for thinner sections.
5. The effects of coating on elongation are slight for Rene 80. Rene 120 elongations were lowered at all temperatures.
6. Exposure at 899C lowers tensile properties of 0.075 cm Rene 80 up to 871C; there is no effect at higher temperatures. Exposure at 982C lowers properties of the thinnest sections (Rene 80 and 120) up to 982C. Thicker sections are less affected at R.T., and show losses up to 871C only.
7. Coarse grain Rene 80 (0.075 cm), bare or coated had higher yield strength than normal grain material and near equal or lower ultimate strengths at R.T. and 871C. Coarser grain Rene 120 (0.15 cm), bare or coated, had lower R.T. and higher 871C strengths. Elongation for both alloys was slightly lower at R.T. and higher at 871C for the coarse grain material.

8. Surface machining generally improves elongation at all temperatures. Strengths at R.T. and 871C are not affected much, and are improved for both alloys at 982C. Exposure at 982C produces similar effects in surface machined and as-cast samples.
9. Machining off the coating and diffusion zone after 982C exposure increases elongation, but because of the reduced thickness, results in generally lower strengths. When corrected for thickness, strengths are higher at 982C and nearly equivalent to pre-exposed values at lower temperatures.
10. Re-heat treating and coating after an initial ~1000 hours, 982C exposure and removal of coating/diffusion area generally increases elongation, but lowers strengths, except for Rene 80 at R.T. and 871C. Additional exposure lowers all properties.

Stress Rupture Properties

Larson-Miller parameter plots, and curves of average rupture life and elongation as a function of section thickness are presented. The general effects are as follows:

1. In all cases, coated or uncoated, rupture life decreases with decreasing thickness for both alloys.
2. Uncoated, loss is least at 760C and greatest at 871C to 982C. Losses of as much as 98% of standard bar life occur for the 0.038 cm section. The greatest loss for 0.075 cm samples is 87% for Rene 120 and 77% for Rene 80. The greatest loss for the 0.15 cm samples is 66% for Rene 120 and 48% for Rene 80.
3. Coating lowers rupture lives up to 982C for all thicknesses of both alloys with greatest losses at lower temperatures and with little or no loss for the 0.075 cm and 0.15 cm thicknesses at 1093C. Maximum losses for 0.075 cm sections are 89% for Rene 80 and 93% for Rene 120.

4. Rupture elongation is generally equivalent for 0.075 cm and thicker samples of both alloys. The 0.038 cm samples are always lower. Coating lowers elongation only at 760C and 871C.
5. Exposure for ~1000 hours at 899C for 0.075 cm Rene 80 has little effect on life. Exposure at 982C lowers life up to 982C, with greater losses at lower temperatures for both alloys. There is little effect at 1093C for the thicker samples. Maximum losses are 99% for 0.038 cm section; 97% for Rene 80 and 98% for Rene 120 0.075 cm sections. For Rene 120 at the lowest temperatures, the 0.15 cm samples lost more life than did the 0.075 cm samples.
6. Exposure at 899C for Rene 80 does not affect elongation. Exposure at 982C lowers elongations for Rene 80 at 760C for all thicknesses and at 871C for the thinner specimens only. The exposure affects Rene 120 elongation to a lesser extent.
7. Coarse grain specimens of both alloys, uncoated or coated have lower rupture lives at 871C and higher lives at 1093C than normal grained specimens.
8. Surface machining of 0.075 cm specimens lowers rupture life at 760C and 871C, but not at higher temperatures. Elongations are generally improved. Coating lowers lives and elongations in most cases up to 982C; not at 1093C, as with as-cast surface specimens. Exposure at 982C of surface machined, coated specimens has similar effects as on as-cast surface specimens.
9. Removing the coating and diffusing zone after exposure results in lowering rupture life, in part because of the reduction in thickness. Correcting for thickness still leaves losses in most cases, especially at 760C. Re-treating and coating generally further lowers life. Re-exposure again drops life in most cases.

Thermal Fatigue

The cycle used on Rene 80 and Rene 120 did not cause failure in 4000 cycles with coated and coated/982C exposed specimens.

Mechanical High Cycle Fatigue

Uncoated, coated, and coated/exposed 982C Rene 120 specimens 0.075 cm thick have 10^7 cycle reversed bending fatigue strengths of 269 to 286 MN/m^2. These values are lower than Rene 80, but do not change as Rene 80 does, due to the coating or exposure.

Microstructures

1. Rene 80 contains eutectic γ' nodules, MC carbides and fine γ' as-cast. Heat treatment eliminates the eutectic γ', makes the dispersion of fine γ' more uniform and precipitates a small amount of fine $M_{23}C_6$ at grain boundaries. A fine de-alloyed band, and some MC carbide depletion within 0.002 cm of the surface may occur. A fine recrystallized grain region is formed usually at the surface.
2. Exposures at 899C and 982C coarsen the γ' and produce more $M_{23}C_6$ at grain boundaries at the expense of the MC.
3. Surface machining after initial heat treatment can remove the de-alloyed band, the MC depleted layer and the fine recrystallized grains.
4. Rene 120 contains eutectic γ' nodules, MC and fine γ' as-cast. Unlike Rene 80, the eutectic γ' is not eliminated by heat treatment. The grain boundary particles are of different composition than in Rene 80. The surface of Rene 120 resembles Rene 80, and the effects of exposure are similar.

5. Codep B-1 coating on both alloys forms a uniform "added" layer consisting of αAl_2O_3 particles embedded in an NiAl structure having some Ti, Cr, etc. from the matrix. A "diffusion" zone of equal thickness is formed and consists of finger-like particles of σ with NiAl, MC and matrix. Normal Al content of added layer is near 30%.
6. Exposures, and tests at high temperatures, agglomerate the diffusion zone fingers, grow further σ needles into the substrate, diffuse additional Al into grain boundaries and lower Al content of the added layer to 21% after ~1000 hours at 899C and ~17% after ~1000 hours at 982C. When Al content falls below ~20%, some of the NiAl starts transforming to Ni_3Al. The changes in the coating on Rene 120 appear to take place at a slightly lower rate than in Rene 80.
7. Removal of exposed coating, re-heat treatment and re-coating produces a like-new coating, but with additional MC depletion in the base alloy. Re-exposures after initial exposure plus re-coating produce similar changes to the first exposure. However, the continual decrease in MC may almost eliminate MC in a thin section after the multiple exposures.
8. For all test conditions, tensile and rupture fractures are generally intergranular, with occasional transgranular paths up to 982C.
9. Surface attack during stress rupture testing of uncoated material produces a uniform general oxidation with many fine internal oxides and an alloy depleted layer. Attack is less on Rene 120 than on Rene 80, and at 1093C, less oxide penetration occurs (greater surface scale). No attack is visible on coated specimens during test or exposure.

RECOMMENDATIONS

The present work provides reasonable data for tensile and stress rupture properties of uncoated, Codep B-1 coated, coated and exposed, etc. thin specimens of a single shape for Rene 80 and Rene 120, and by similarity to other cast nickel-base superalloys. Due to the greater variability of test results for the thin sections, even more data is desirable to increase reliability. However, more important aspects that require investigation are the following:

1. The mechanism of the geometry induced property changes should be established. Some types of tests that would be useful are:
    a. Creep tests, especially interrupted creep tests in which the origin and propagation behavior of the failure is followed.
    b. Tests on different geometries of thin sections, such as wider specimens, shorter gage lengths, etc.
2. The effects of other metallurgical variables, including relative and absolute grain size, grain orientation (directional solidification), etc. are useful in themselves and for the light they shed on the mechanisms.
3. The effects of other coatings, particularly diffusion barrier types or other types that produce less coating-substrate diffusion, are of importance to reduce or avoid the effects of ordinary aluminide coatings.
4. Comparison of test specimen performance with the more complicated behavior of actual turbine blades is necessary to insure that the application methods of the data obtained are correct.

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Table I Effect of Surface Machining on Tensile Properties of Rene 80

Specimen Condition	Test Temp. °C	(°F)	0.2% Y.S. MN/m^2	(ksi)	U.T.S. MN/m^2	(ksi)	Elong. %
AC-U	R.T.		800	(116.0)	1075	(155.9)	9.2*
AC-C	"		785	(113.8)	1004	(145.6)	7.8*
AC-C-E	"		680	(98.6)	720	(104.4)	4.9
SM-U	"		789	(114.4)	1086	(157.5)	11.5
SM-C	"		756	(109.7)	1095	(158.8)	8.2*
SM-C-E	"		720	(104.4)	883	(128.0)	5.5*
SM-C-E-SM	"		767	(111.3)	855	(124.0)	10.1
SM-C-E-SM-C	"		803	(116.5)	977	(141.7)	4.9
SM-C-E-SM-C-E	"		613	(88.9)	701	(101.7)	1.5
AC-U	871	(1600)	481	(69.7)	683	(99.0)	10.0*
AC-C	"	"	491	(71.2)	653	(94.7)	8.2*
AC-C-E	"	"	542	(78.6)	560	(81.2)	0.4*
SM-U	"	"	476	(69.0)	645	(93.6)	17.4
SM-C	"	"	471	(68.5)	630	(91.3)	7.7*
SM-C-E	"	"	513	(74.4)	547	(79.4)	1.2*
SM-C-E-SM	"	"	421	(61.0)	538	(78.0)	1.6
SM-C-E-SM-C	"	"	503	(73.0)	612	(88.8)	3.8*
SM-C-E-SM-C-E	"	"	307	(44.5)	351	(50.9)	1.3
AC-U	982	(1800)	224	(32.5)	350	(50.8)	15.2
AC-C	"	"	250	(36.3)	350	(50.8)	9.2
AC-C-E	"	"	275	(39.9)	350	(50.8)	6.0*
SM-U	"	"	259	(37.6)	374	(54.2)	15.7
SM-C	"	"	258	(37.5)	359	(52.1)	12.9*
SM-C-E	"	"	221	(32.0)	317	(46.0)	4.0*
SM-C-E-SM	"	"	280	(40.6)	385	(55.9)	4.6*
SM-C-E-SM-C	"	"	246	(35.7)	312	(45.3)	4.7*
SM-C-E-SM-C-E	"	"	185	(26.8)	233	(33.8)	4.2

See Footnotes on following page.

Footnotes for Table I

All surface machined specimens from Heat B353

Specimen condition: AC = As cast, SM = Surface machined 0.06 mm per side (initial); 0.12 mm per side (second time); U = Uncoated; C = Coated; E = Exposed at 982C, ~1000 hours.

All AC values taken at 0.075 cm thickness from Figs. 2, 5, 8, 10, 11, 12, as required.

All SM values "corrected" to 0.075 cm thickness using ratios taken from Figs. 10-12.

Values are averages of 3 to 4 specimens per condition.

* Some specimens failed at or near end of gage length.

Table II Tensile Properties of Coarse Grain Rene 80

Specimen Condition	Test Temp. °C	(°F)	0.2% Y.S. MN/m^2	(ksi)	U.T.S. MN/m^2	(ksi)	Elong. %
CG-U	R.T.		884	(128.2)	1025	(148.6)	7.1
Normal-U	"		796	(115.4)	1077	(156.2)	9.3
CG-C	"		903	(130.9)	1029	(149.2)	5.8
Normal-C	"		780	(113.1)	1005	(145.8)	7.8
CG-U	871	(1600)	504	(73.1)	653	(94.7)	11.0
Normal-U	"	"	479	(69.5)	685	(99.3)	10.3
CG-C	"	"	506	(73.4)	605	(87.7)	12.3
Normal-C	"	"	487	(70.7)	656	(95.1)	9.0

All CG specimens from Heat B322, 0.075 cm thick.

CG = coarse grain (no nucleation coating on mold), U = uncoated, C = coated

CG results are averages of 2 tests. Normal results from Figs. 2, 5 and 8.

Table III Effect of Surface Machining on Stress Rupture Properties of Rene 80

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m^2	(ksi)	Life hrs	Elong. %
AC-U	760	(1400)	565	(82.0)	67.2	9.2
AC-C	"	"	"	"	14.8	3.9
AC-C-E	"	"	"	"	<2.0	0.4
SM-U	"	"	"	"	6.0	10.3*
SM-C	"	"	"	"	7.1	5.9*
SM-C-E	"	"	"	"	† 22.2	2.9
SM-C-E-SM	"	"	"	"	† 12.1↑	3.4*
SM-C-E-SM-C	"	"	"	"	0.2↑	1.1*
SM-C-E-SM-C-E	"	"	"	"	F.O.L.	-
AC-U	871	(1600)	262	(38.0)	245.2	6.2
AC-C	"	"	"	"	114.6	5.2*
AC-C-E	"	"	"	"	46.1	2.9
SM-U	"	"	"	"	153.4	7.3*
SM-C	"	"	"	"	54.1	5.2*
SM-C-E	"	"	"	"	29.5	2.6*
SM-C-E-SM	"	"	"	"	† 12.5	2.3*
SM-C-E-SM-C	"	"	"	"	14.6	2.5*
SM-C-E-SM-C-E	"	"	"	"	16.1	2.9*
AC-U	982	(1800)	117	(17.0)	93.2	5.4
AC-C	"	"	"	"	72.7	7.7*
AC-C-E	"	"	"	"	52.7	7.3
SM-U	"	"	"	"	110.8	11.9*
SM-C	"	"	"	"	105.6	12.7*
SM-C-E	"	"	"	"	37.4	8.2*
SM-C-E-SM	"	"	"	"	32.7	3.2*
SM-C-E-SM-C	"	"	"	"	10.0	3.4*
SM-C-E-SM-C-E	"	"	"	"	5.4	4.6*

Table III (Cont.)

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m^2	(ksi)	Life hrs	Elong. %
AC-U	1093	(2000)	34.5	(5.0)	48.9	16.5
AC-C	"	"	"	"	55.7	12.2
AC-C-E	"	"	"	"	53.9	5.6
SM-U	"	"	"	"	72.7	12.5*
SM-C	"	"	"	"	61.1	11.0*
SM-C-E	"	"	"	"	43.5	10.4*
SM-C-E-SM	"	"	"	"	19.0†	5.6*
SM-C-E-SM-C	"	"	"	"	14.5	4.5*
SM-C-E-SM-C-E	"	"	"	"	16.8	6.0*

All surface machined specimens from Heat B322

Specimen condition: AC = As cast, SM = Surface machined 0.06 mm per side (initial); 0.12 mm per side (second time); U = Uncoated; C = Coated; E = Exposed at 982C, ~1000 hours.

All AC values are weighted log averages of actual 0.075 cm specimen test results.

All SM values "corrected" to 0.075 cm thickness using ratios calculated from Figs. 16, 18, 20, 21.

Values are log or † log-linear averages of 3 to 4 specimens per condition.

* Some specimens failed at or near end of gage length.

† Specimen(s) failed in grip. Life should be higher.

Table IV Stress Rupture Properties of Coarse Grain Rene 80

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m^2	(ksi)	Life hrs	Elong. %
CG-U	871	(1600)	262	(38.0)	158.1	6.3
Normal-U	"	"	"	"	245.3	6.2
CG-C	"	"	"	"	69.9	4.7*
Normal-C	"	"	"	"	114.4	5.2*
CG-U	1093	(2000)	34.5	(5.0)	61.1	7.0
Normal-U	"	"	"	"	50.0	16.5*
CG-C	"	"	"	"	66.0	6.1
Normal-C	"	"	"	"	56.7	12.2

All CG specimens from Heat B322, 0.075 cm thick

CG = coarse grain (no nucleation coating on mold), U = uncoated, C = coated.

CG results are log averages of 2 tests. Normal results are weighted log averages of both heats.

* Specimen(s) failed at or near end of gage length.

Table V Effect of Surface Machining on Tensile Properties of Rene 120

Specimen Condition	Test Temp. °C (°F)	0.2% Y.S. MN/m^2 (ksi)	U.T.S. MN/m^2 (ksi)	Elong. %
AC-U	R.T.	851 (123.4)	1019 (147.8)	7.1
AC-C	"	805 (116.8)	881 (127.8)	4.0
AC-C-E	"	622 (90.2)	628 (91.1)	1.7
SM-U	"	823 (119.3)	1018 (147.7)	8.0
SM-C	"	816 (118.4)	887 (128.7)	3.1*
SM-C-E	"	713 (103.4)	716 (103.8)	1.8
SM-C-E-SM	"	650 (94.3)	867 (125.7)	1.8
SM-C-E-SM-C	"	718 (104.1)	747 (108.3)	2.2
SM-C-E-SM-C-E	"	554 (80.3)	566 (82.1)	0.8
AC-U	871 (1600)	660 (95.7)	723 (104.9)	3.6
AC-C	" "	583 (84.6)	676 (98.1)	2.7*
AC-C-E	" "	540 (78.3)	572 (83.0)	1.0
SM-U	" "	639 (92.7)	714 (103.6)	7.8
SM-C	" "	592 (85.9)	700 (101.5)	3.6*
SM-C-E	" "	524 (76.0)	558 (80.9)	1.1*
SM-C-E-SM	" "	590 (85.6)	597 (86.6)	2.5*
SM-C-E-SM-C	" "	472 (68.5)	525 (76.2)	2.0
SM-C-E-SM-C-E	" "	280 (40.6)	311 (45.1)	1.0
AC-U	982 (1800)	359 (52.1)	433 (62.8)	4.1
AC-C	" "	382 (55.4)	433 (62.8)	1.9*
AC-C-E	" "	340 (49.3)	385 (55.8)	1.6
SM-U	" "	400 (58.0)	490 (71.0)	4.6
SM-C	" "	370 (53.6)	464 (67.3)	3.4
SM-C-E	" "	334 (48.5)	356 (51.7)	2.2*
SM-C-E-SM	" "	348 (50.4)	478 (69.3)	3.3*
SM-C-E-SM-C	" "	316 (45.9)	402 (58.3)	3.8*
SM-C-E-SM-C-E	" "	262 (38.0)	307 (44.5)	3.2

Table VI Tensile Properties of Coarse Columnar Grain Rene 120

Specimen Condition	Test Temp. °C	(°F)	0.2% Y.S. MN/m²	(ksi)	U.T.S. MN/m²	(ksi)	Elong. %
CCG-U	R.T.		828	(120.1)	915	(132.7)	5.1
Normal-U	"		885	(128.4)	1016	(147.3)	7.2*
CCG-C	"		820	(119.0)	901	(130.7)	4.1
Normal-U	"		828	(120.1)	991	(143.7)	6.8*
CCG-U	871	(1600)	638	(92.6)	738	(107.0)	6.9
Normal-U	"	"	630	(91.4)	738	(107.0)	4.5
CCG-C	"	"	681	(98.8)	769	(111.5)	5.4
Normal-C	"	"	620	(89.9)	729	(105.7)	3.2*

All CCG specimens from Heat B415, 0.15 cm thick.

CCG = coarse columnar grain, U = uncoated, C = coated.

CCG results are averages of 2 tests at R.T., single test at 871C. Normal results from Figs. 29, 31, 33.

* Specimen(s) failed at or near end of gage length.

Table VII Effect of Surface Machining on Stress Rupture Properties of Rene 120

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m^2	(ksi)	Life hrs	Elong. %
AC-U	760	(1400)	634	(92.0)	58.4	3.0
AC-C	"	"	"	"	7.2	1.5
AC-C-E	"	"	"	"	† 6.2	1.3*
SM-U	"	"	"	"	22.3	3.4
SM-C	"	"	"	"	4.8	3.3
SM-C-E	"	"	"	"	† 4.1	2.5
SM-C-E-SM	"	"	"	"	0.2↑	3.6*
SM-C-E-SM-C	"	"	"	"	F.O.L.	2.1*
SM-C-E-SM-C-E	"	"	"	"	F.O.L.	0.9*
AC-U	871	(1600)	310	(45.0)	† 37.0	2.9
AC-C	"	"	"	"	† 24.2	1.8*
AC-C-E	"	"	"	"	7.4	1.6
SM-U	"	"	"	"	† 17.6	2.7
SM-C	"	"	"	"	35.6	1.5*
SM-C-E	"	"	"	"	21.4	2.8*
SM-C-E-SM	"	"	"	"	† 45.5↑	7.7
SM-C-E-SM-C	"	"	"	"	15.3	3.7*
SM-C-E-SM-C-E	"	"	"	"	1.6↑	3.4*
AC-U	982	(1800)	145	(21.0)	63.9	2.5
AC-C	"	"	"	"	43.6	3.4*
AC-C-E	"	"	"	"	23.5	4.4
SM-U	"	"	"	"	104.2	3.3
SM-C	"	"	"	"	43.5	4.1*
SM-C-E	"	"	"	"	36.2	10.6
SM-C-E-SM	"	"	"	"	69.8	5.6*
SM-C-E-SM-C	"	"	"	"	23.8	10.9*
SM-C-E-SM-C-E	"	"	"	"	6.5	?28.7

Table VII (Cont.)

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m²	(ksi)	Life hrs	Elong. %
AC-U	1093	(2000)	55	(8.0)	39.5	3.7
AC-C	"	"	"	"	38.8	4.9
AC-C-E	"	"	"	"	39.4	4.3
SM-U	"	"	"	"	34.7	4.1
SM-C	"	"	"	"	39.7	6.6
SM-C-E	"	"	"	"	53.0	6.2*
SM-C-E-SM	"	"	"	"	27.1	4.3*
SM-C-E-SM-C	"	"	"	"	22.7	11.6
SM-C-E-SM-C-E	"	"	"	"	13.1	14.3

All surface machined specimens from Heat B415

Specimen condition: AC = As cast, SM = Surface machined 0.06 mm per side (initial); 0.12 mm per side (second time); U = Uncoated; C = Coated; E = Exposed at 982C, ~1000 hours.

All AC values are weighted log averages of actual 0.075 cm specimen test results.

All SM values "corrected" to 0.075 cm thickness using ratios calculated from Figs. 39-42.

Values are log or † log-linear averages of 3 specimens per condition.

* Some specimens failed at or near end of gage length.

† Specimen(s) failed in grip. Life should be higher.

Table VIII Stress Rupture Properties of Coarse Columnar Grain Rene 120

Specimen Condition	Test Temp. °C	(°F)	Test Stress MN/m^2	(ksi)	Life hrs	Elong. %
CCG-U	871	(1600)	310	(45.0)	160.7	2.1
Normal-U	"	"	"	"	199.0	2.1
CCG-C	"	"	"	"	112.1	2.4
Normal-C	"	"	"	"	121.2	2.1
CCG-U	1093	(2000)	55	(8.0)	83.4	3.9*
Normal-U	"	"	"	"	70.6	3.6
CCG-C	"	"	"	"	106.7	4.0
Normal-C	"	"	"	"	56.8	3.4

All CCG specimens from Heat B415, 0.15 cm thick

CCG = coarse columnar grain, U = uncoated, C = coated.

CCG results are log average of 2 tests. Normal results are weighted log averages of both heats at 1093C, and of Heat B415 at 871C.

* Specimen(s) failed at or near end of gage length.

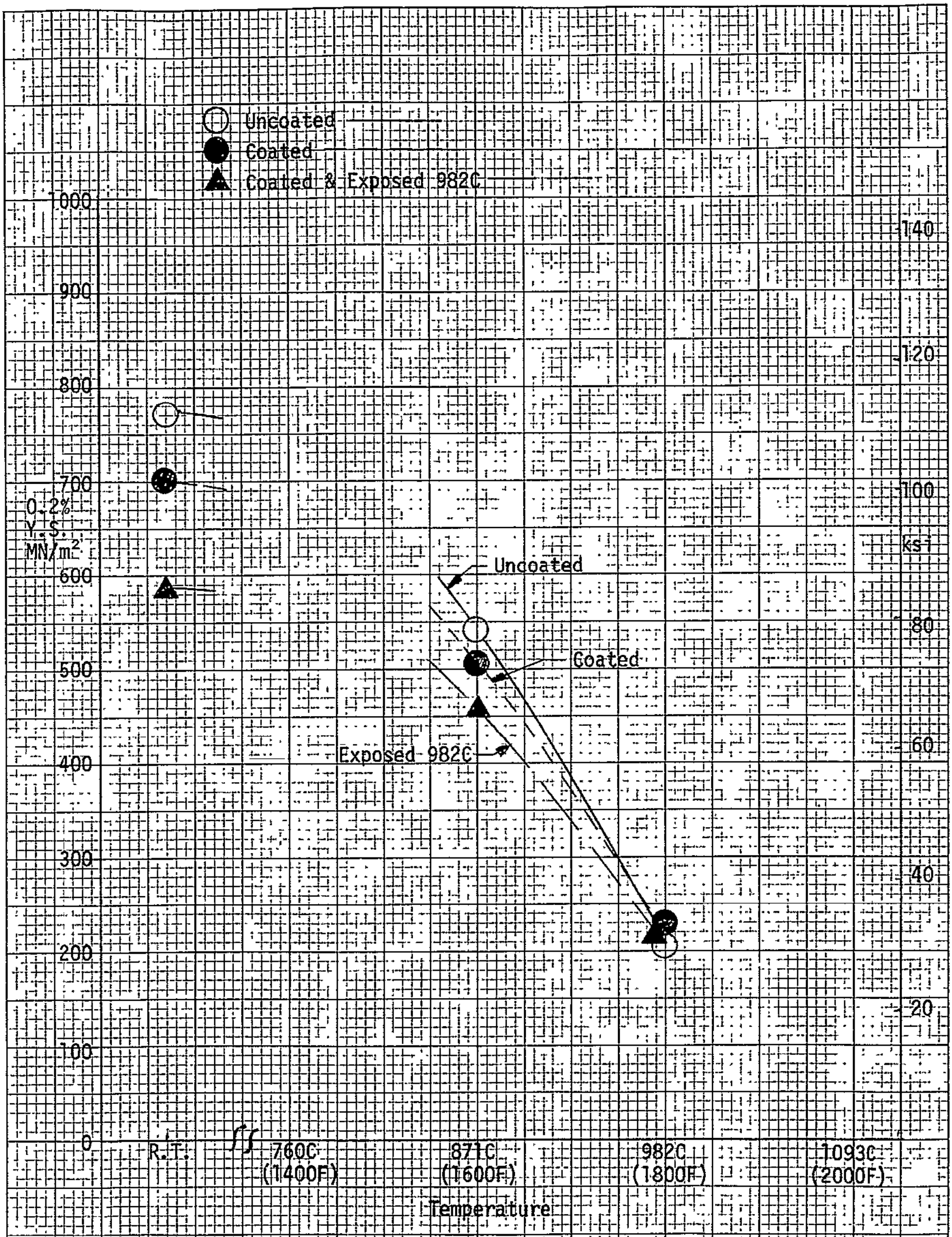


Fig. 1 0.2% Yield Strength, Rene 80, 0.038 cm Thick

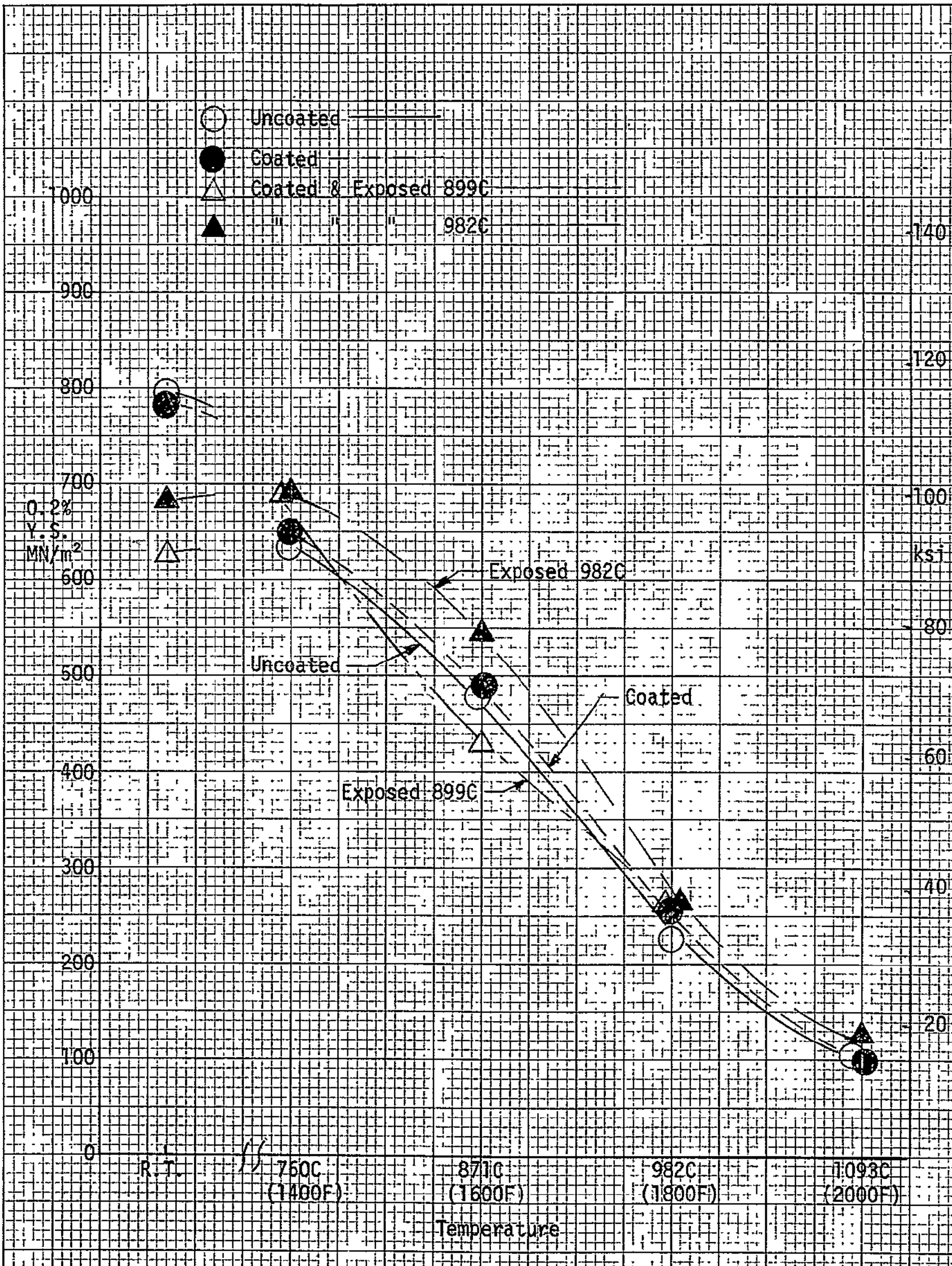


Fig. 2 0.2% Yield Strength, Rene 80, 0.075 cm Thick

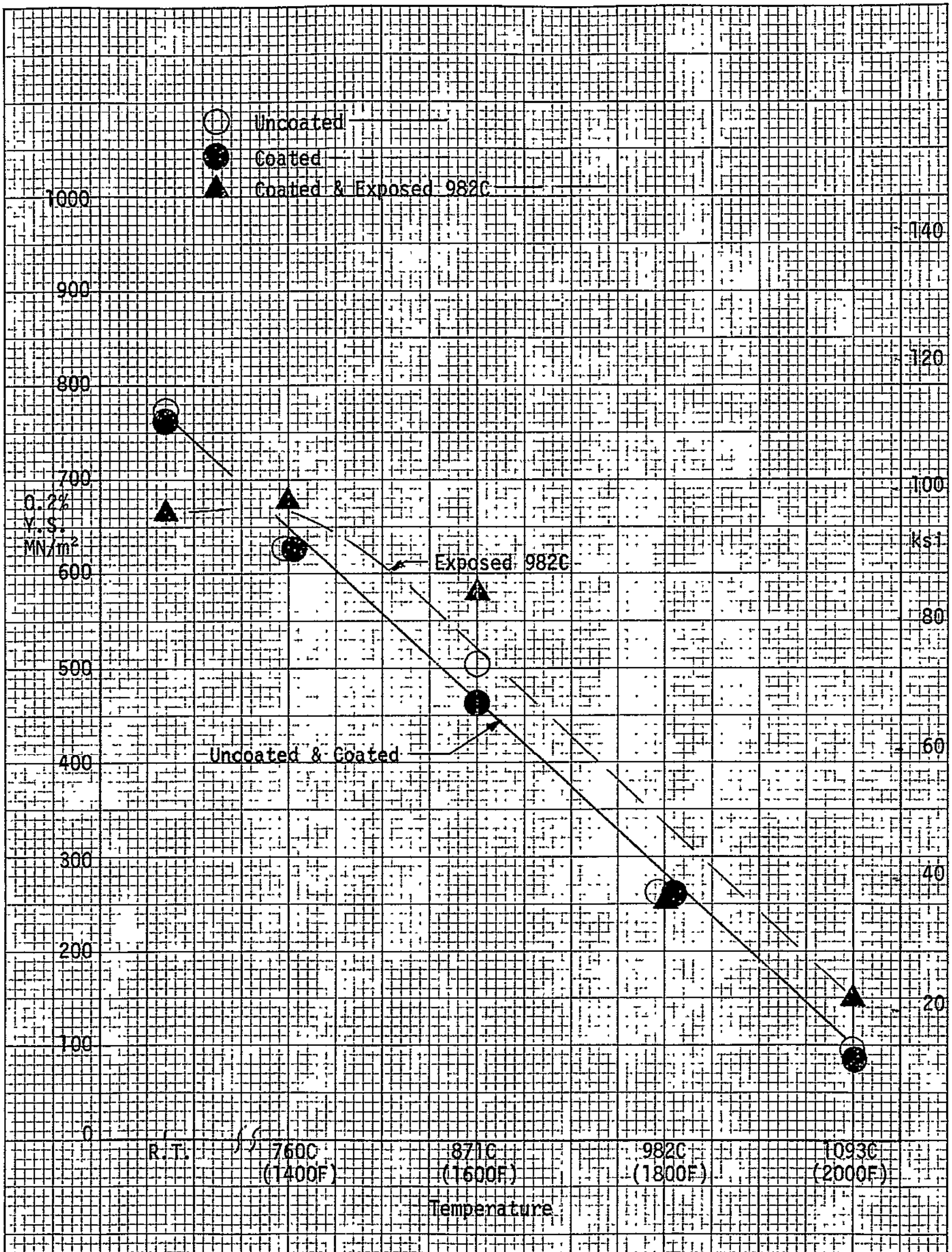


Fig. 3 0.2% Yield Strength, Rene 80, 0.15 cm Thick

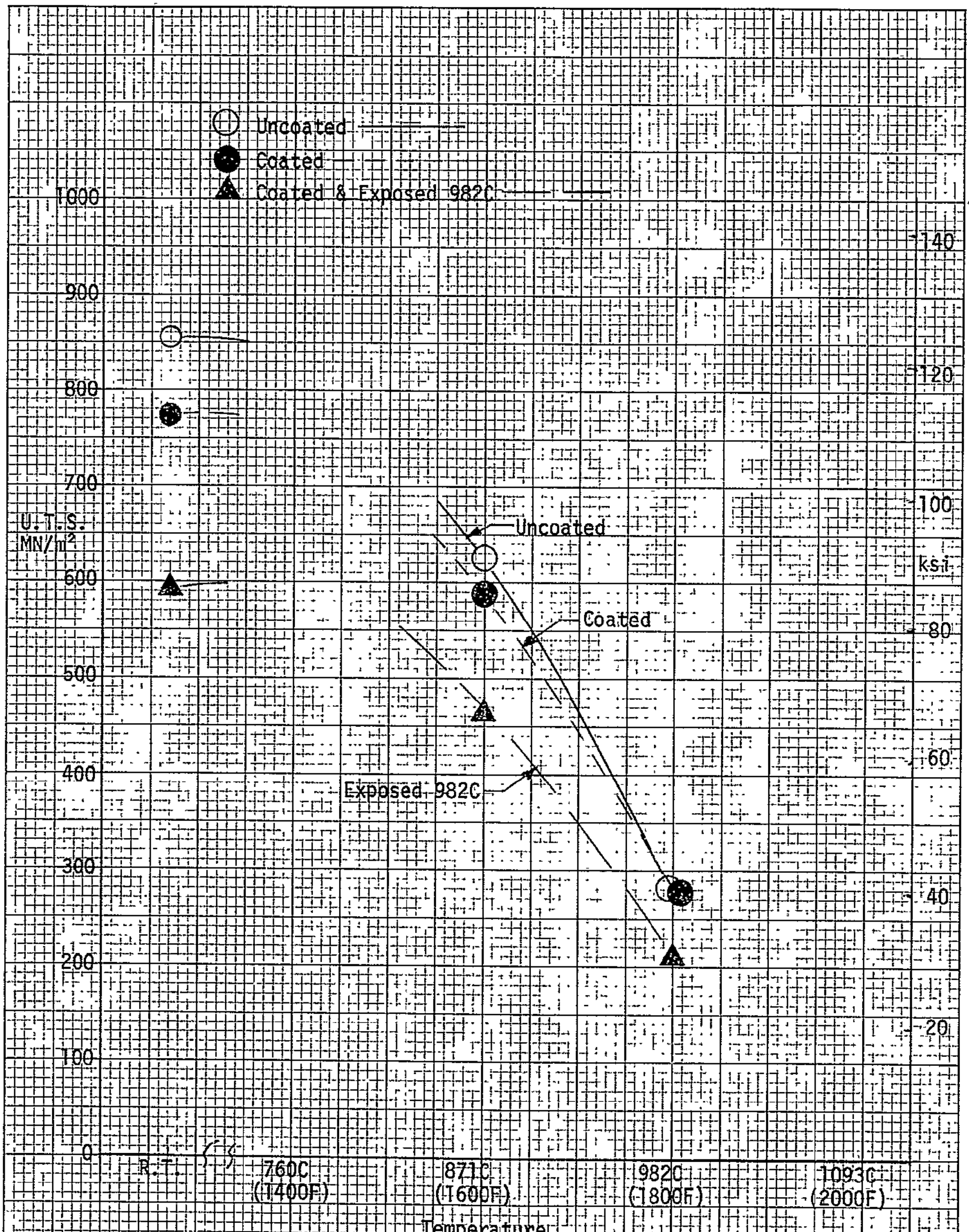


Fig. 4 Ultimate Tensile Strength, Rene 80, 0.038 cm Thick

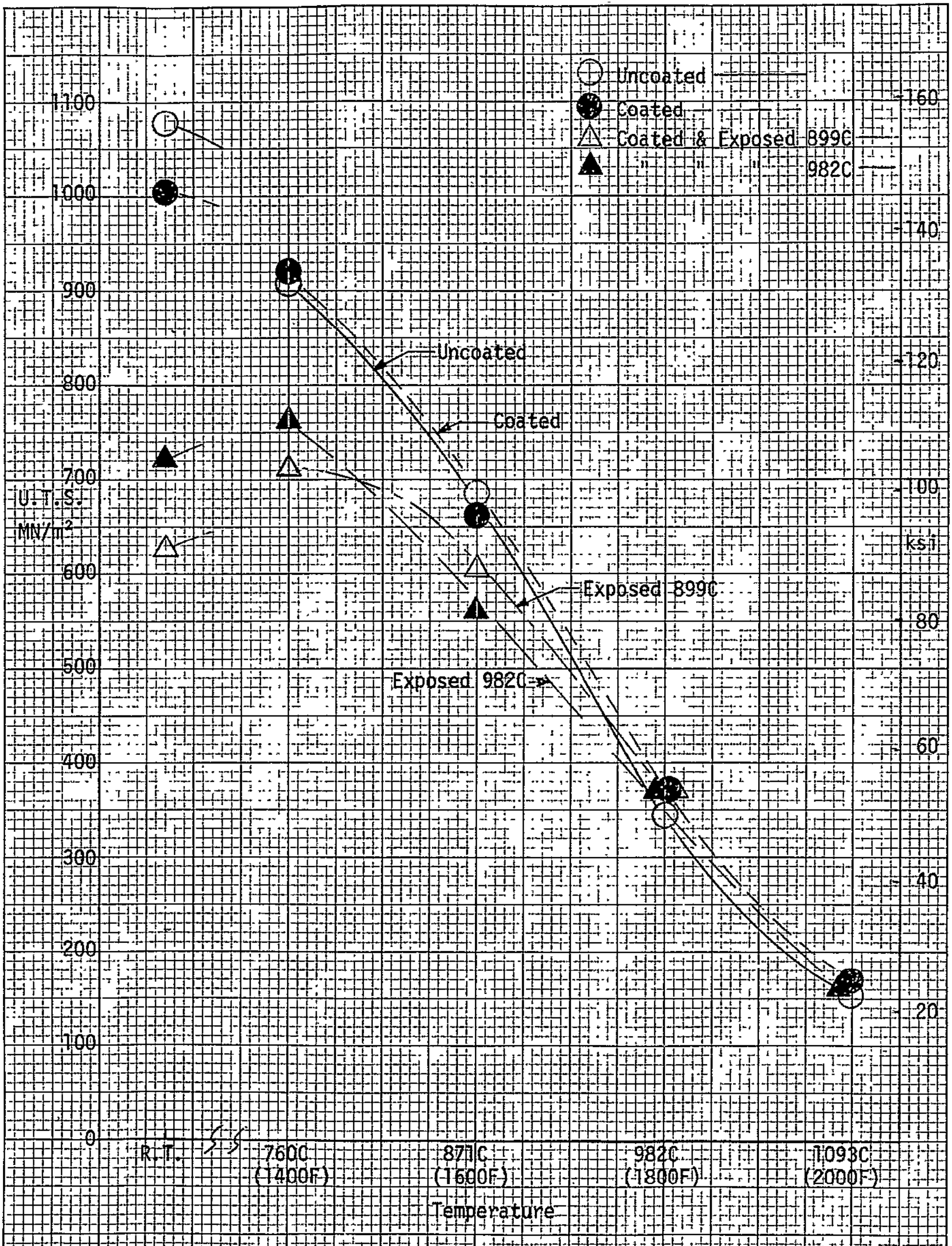


Fig. 5 Ultimate Tensile Strength, Rene 80, 0.075 cm Thick

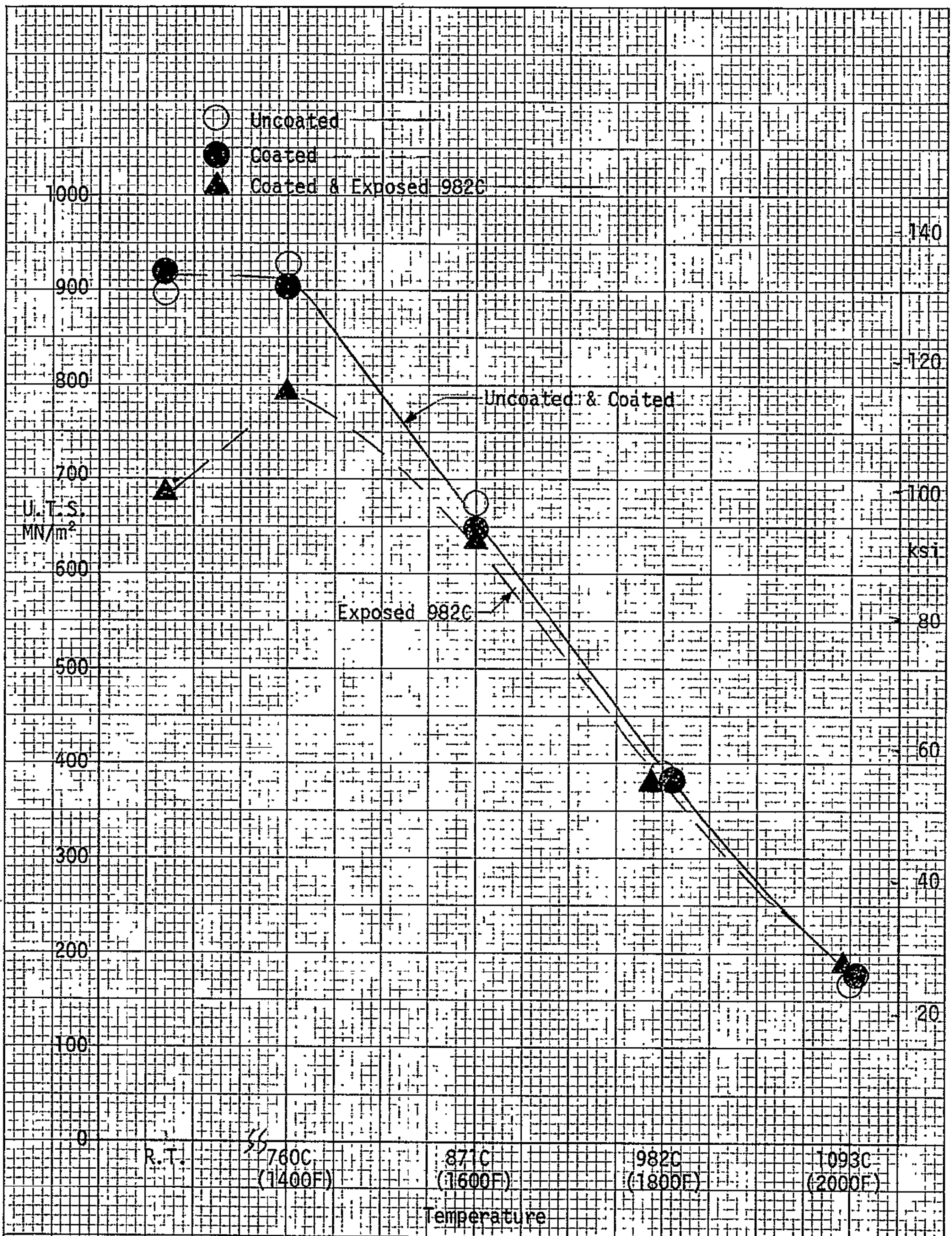


Fig. 6 Ultimate Tensile Strength, Rene 80, 0.15 cm Thick

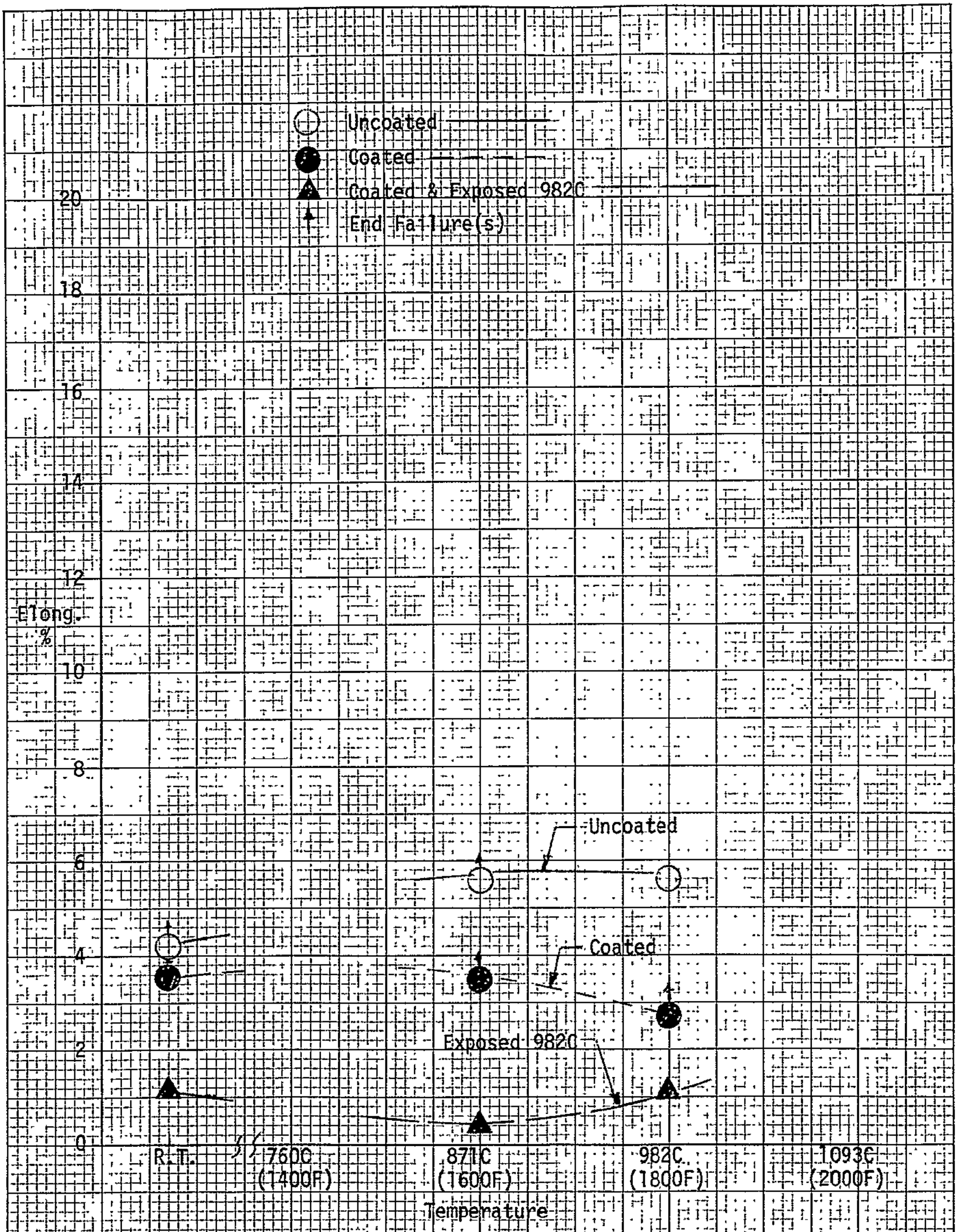


Fig. 7 Tensile Elongation, Rene 80, 0.038 cm Thick

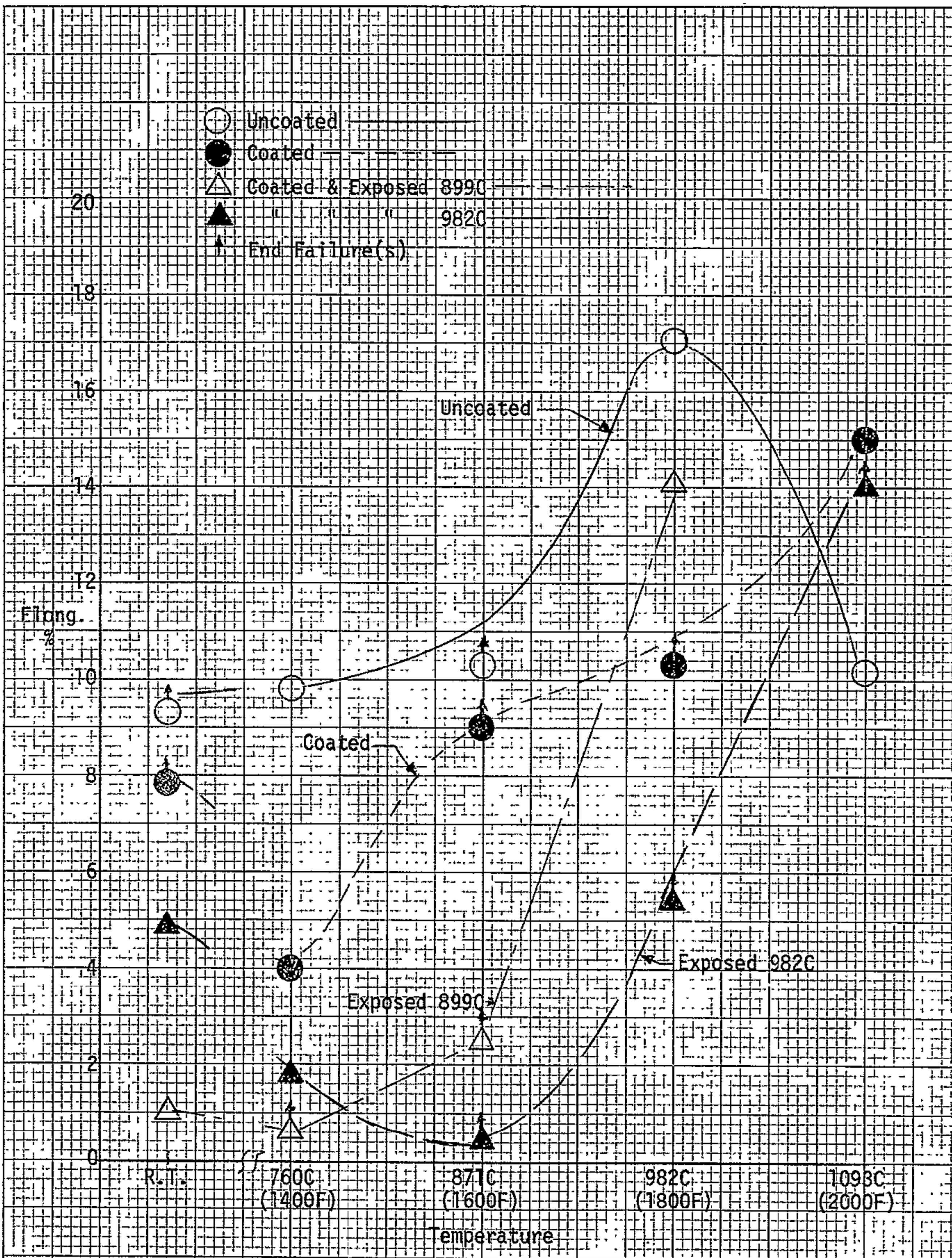


Fig. 8 Tensile Elongation, Rene 80, 0.075 cm Thick

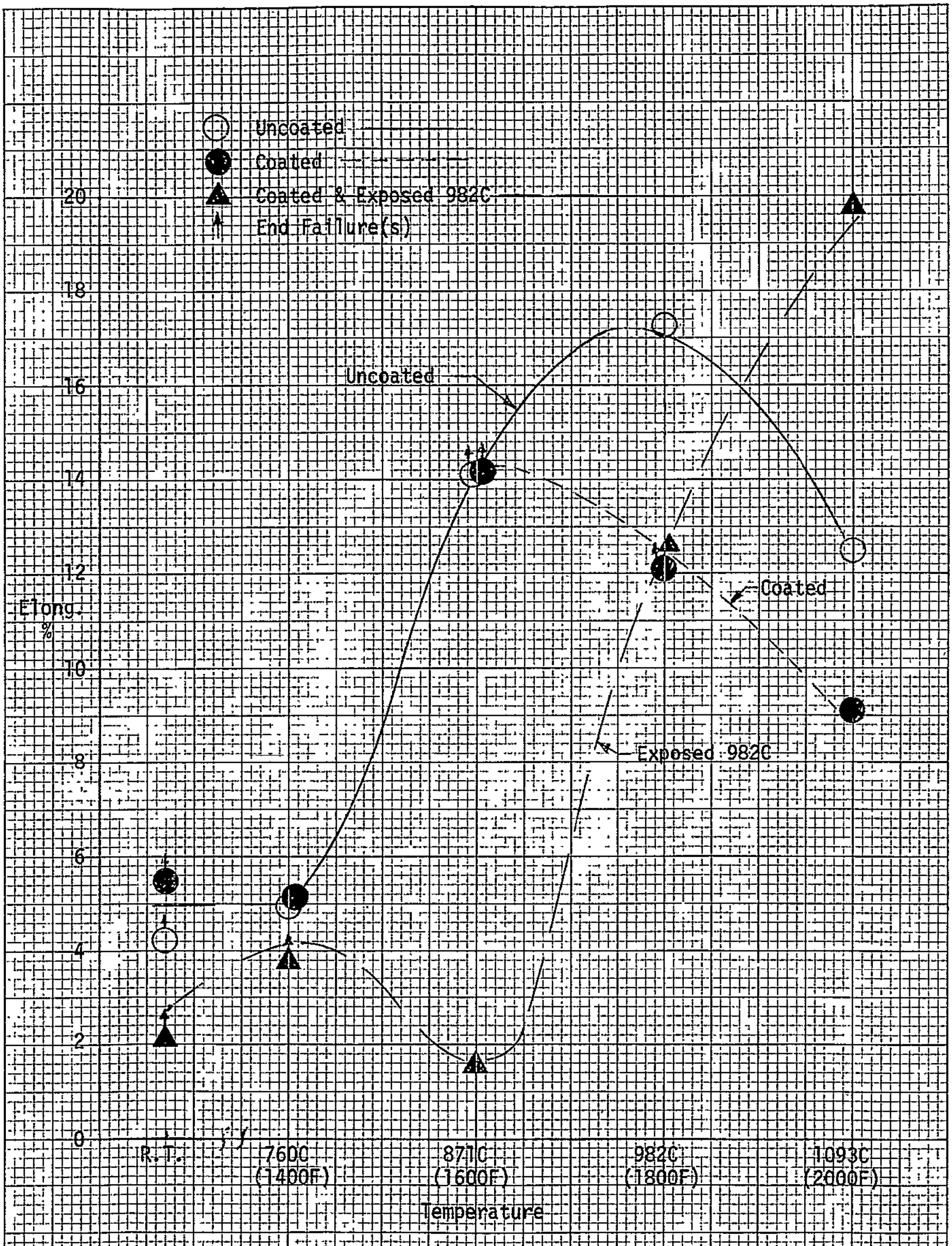


Fig. 9 Tensile Elongation, Rene 80, 0.15 cm Thick

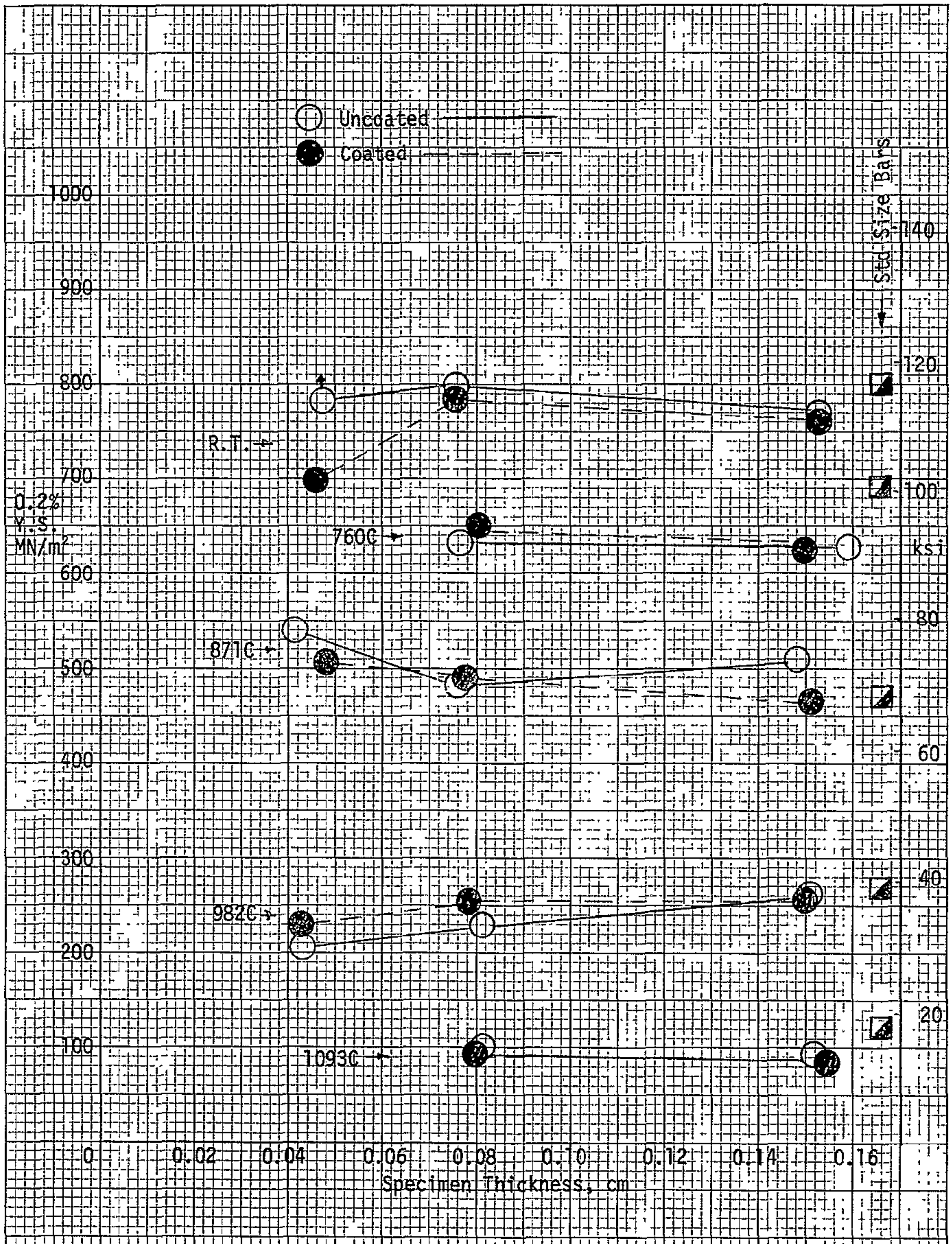


Fig. 10 Effect of Section Thickness on 0.2% Yield Strength, Rene 80

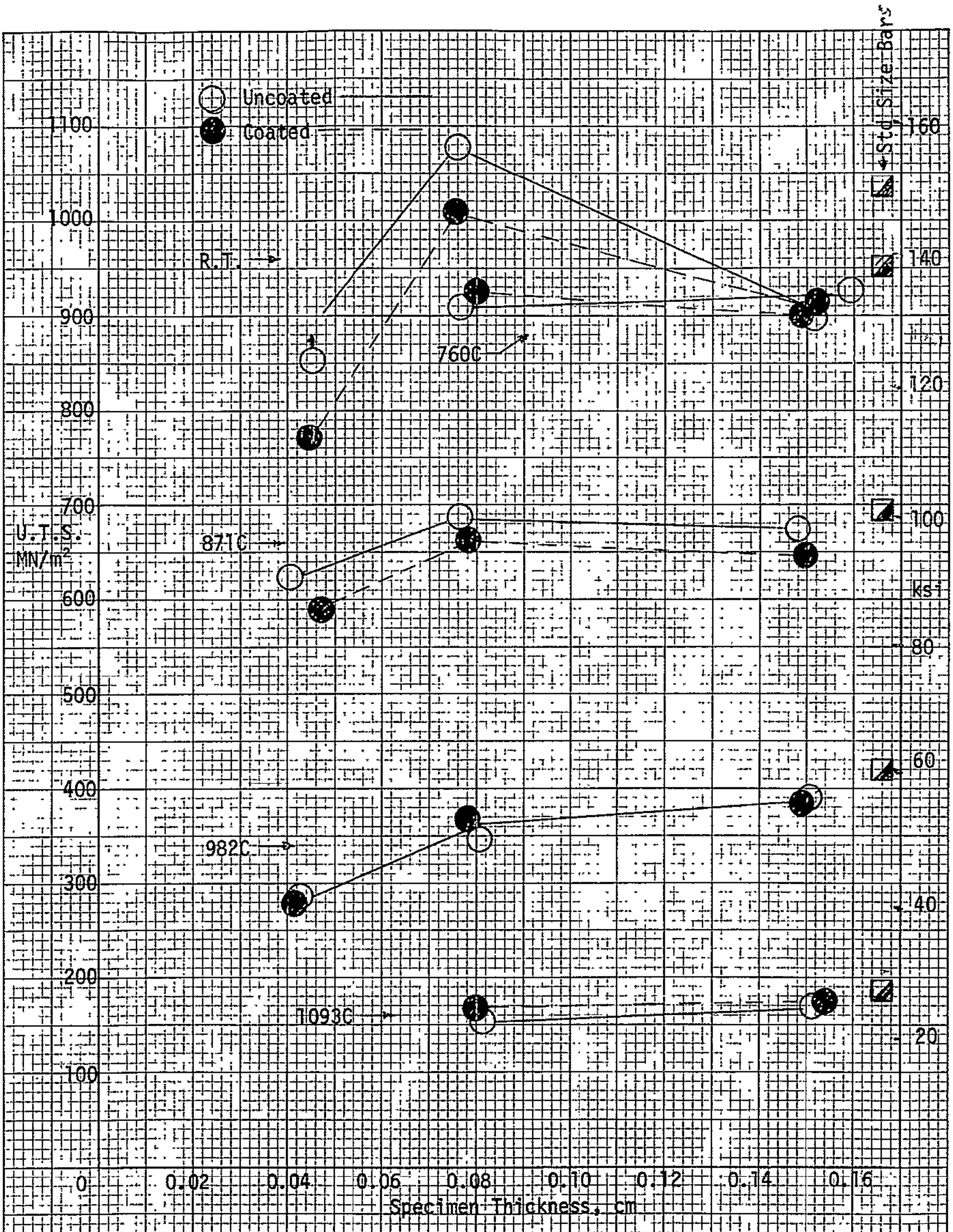


Fig. 11 Effect of Section Thickness on Ultimate Tensile Strength, Rene 80

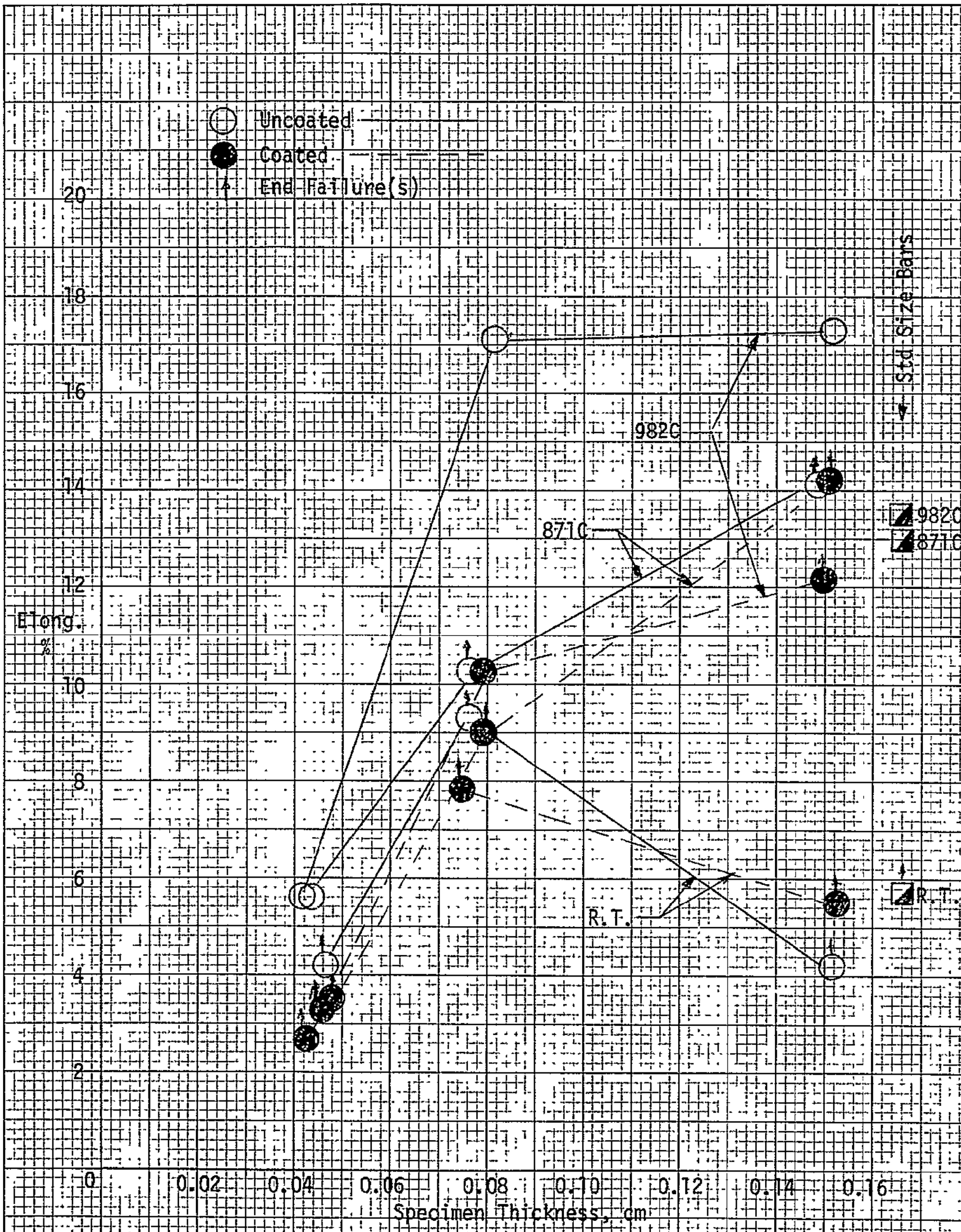


Fig. 12 Effect of Section Thickness on Tensile Elongation, Rene 80

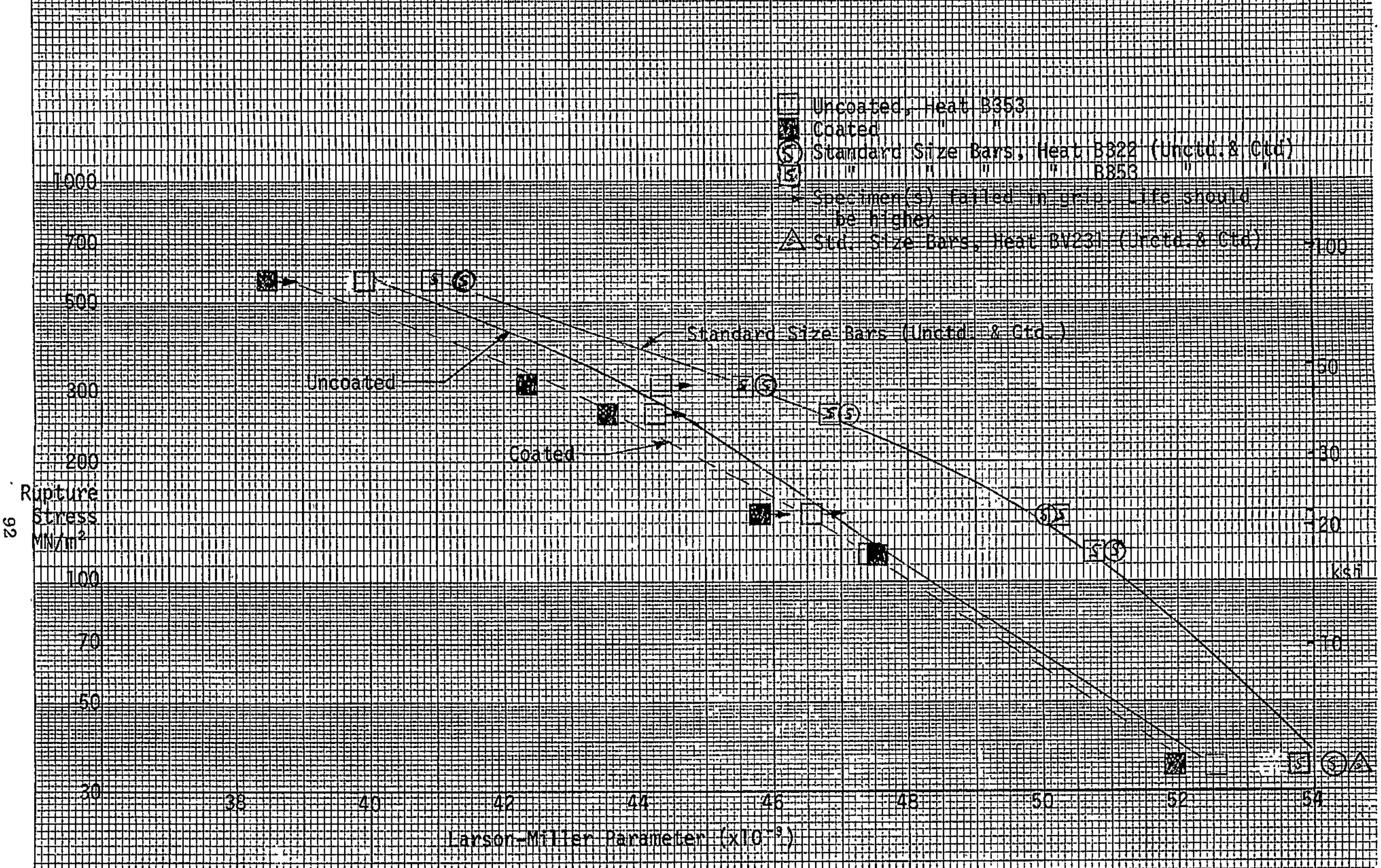


Fig. 13 Stress Rupture Strength of 0.038 cm and Standard Bars, Rene 80

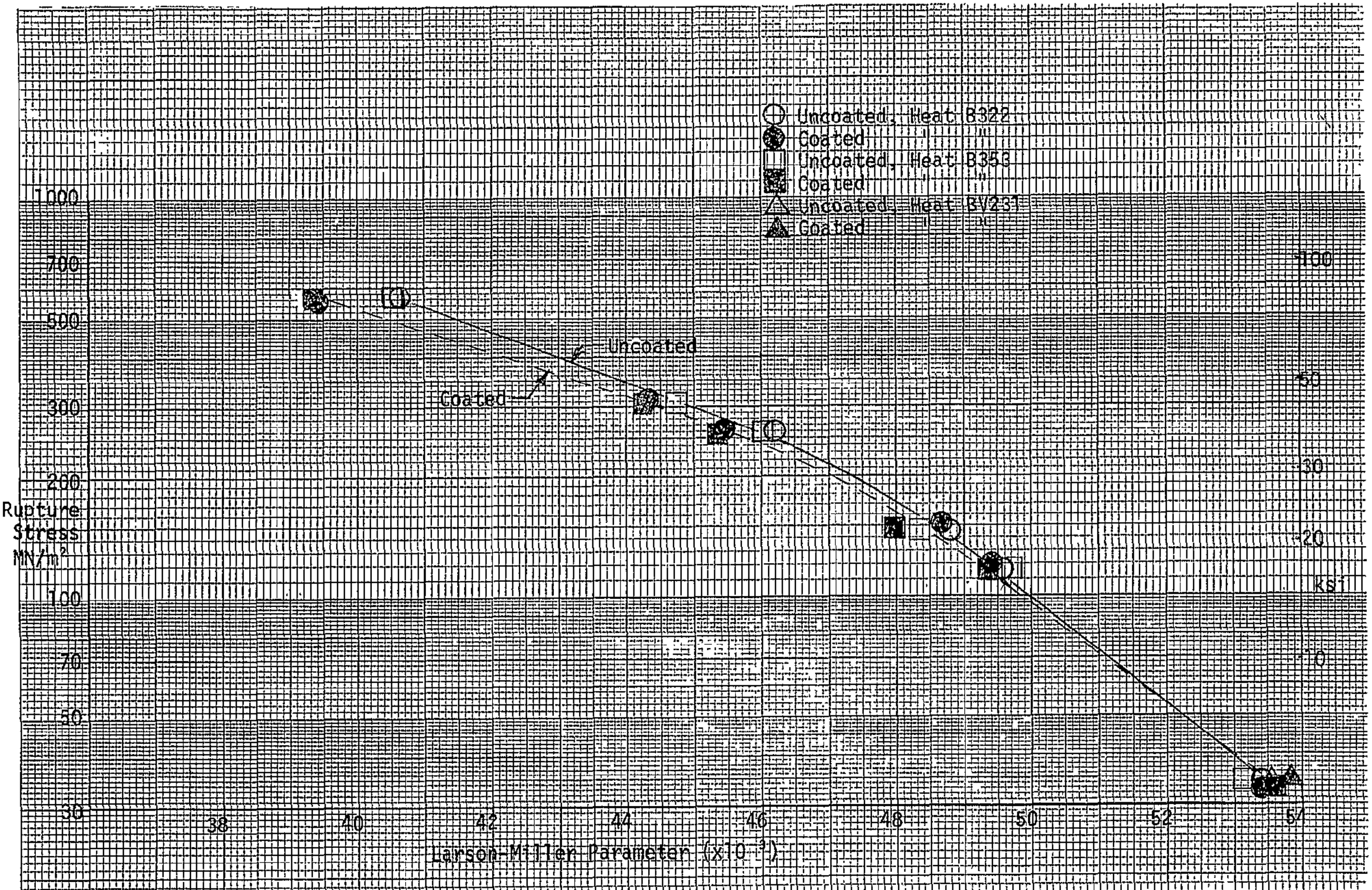


Fig. 14 Stress Rupture Strength of 0.075 cm Thick Rene 80

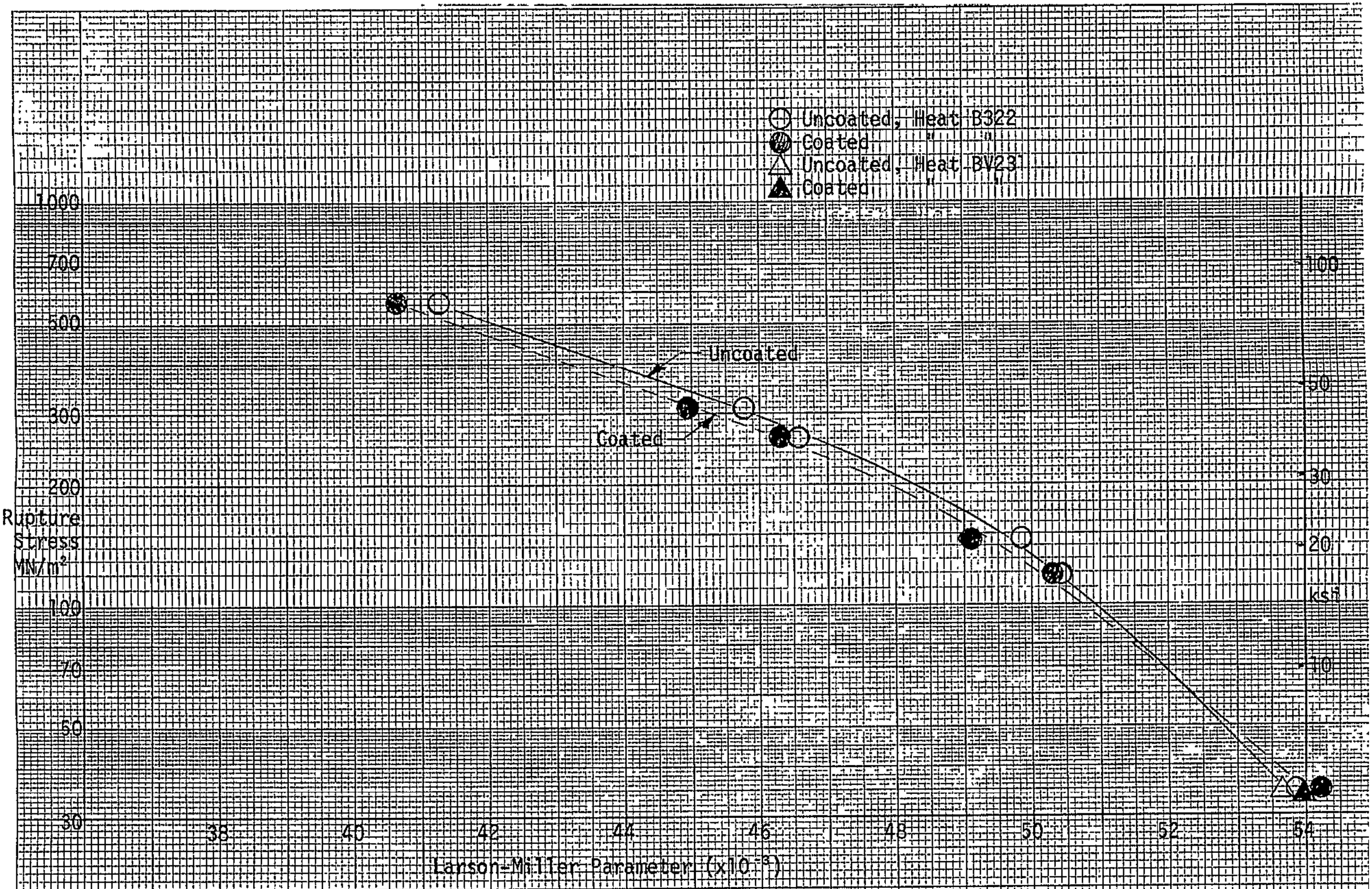


Fig. 15 Stress Rupture Strength of 0.15 cm Thick Rene 80

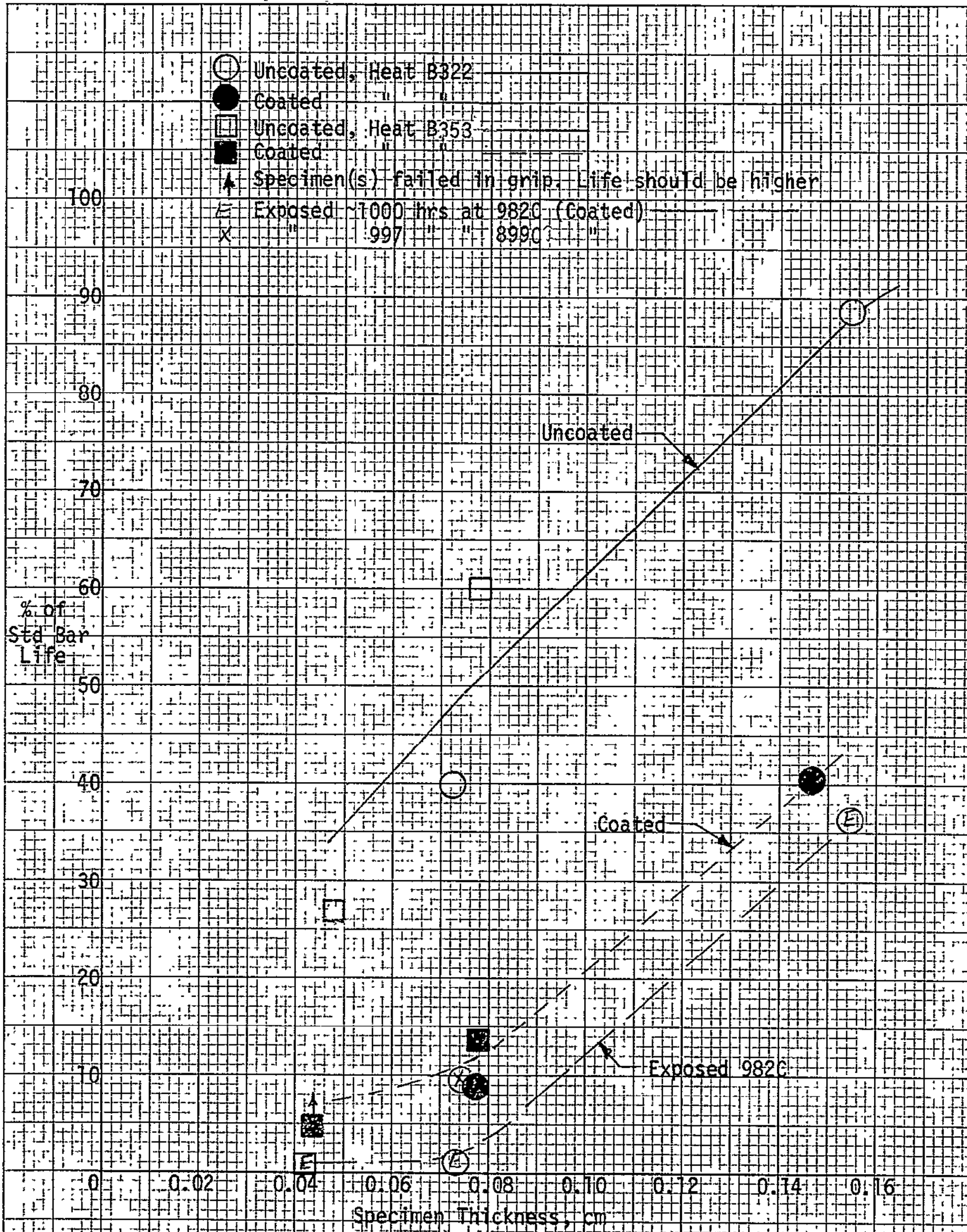


Fig. 16 Percent of Standard Bar Life, Rene 80, 760C, 565 MN/m²

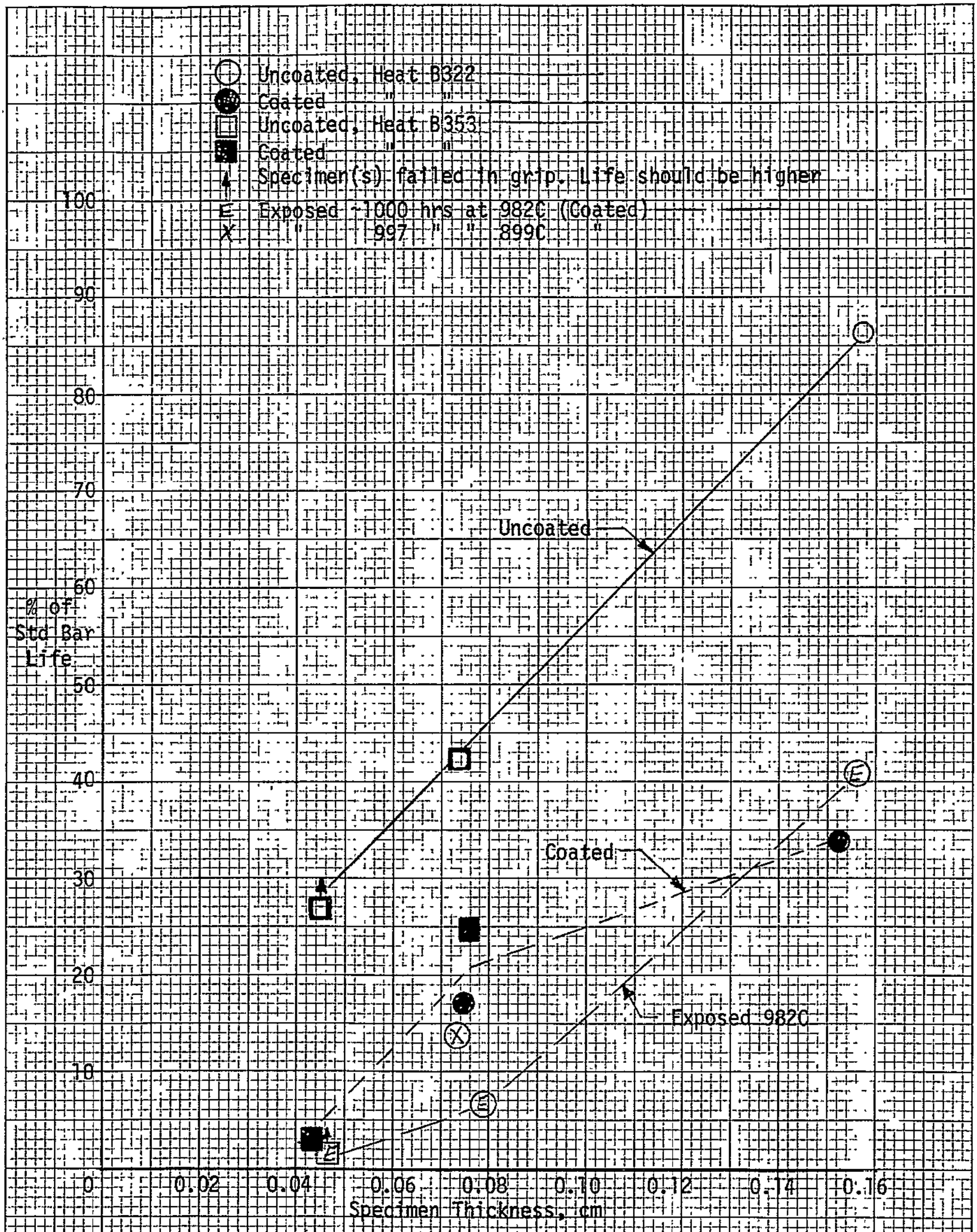


Fig. 17 Percent of Standard Bar Life, Rene 80, 871C, 310 MN/m²

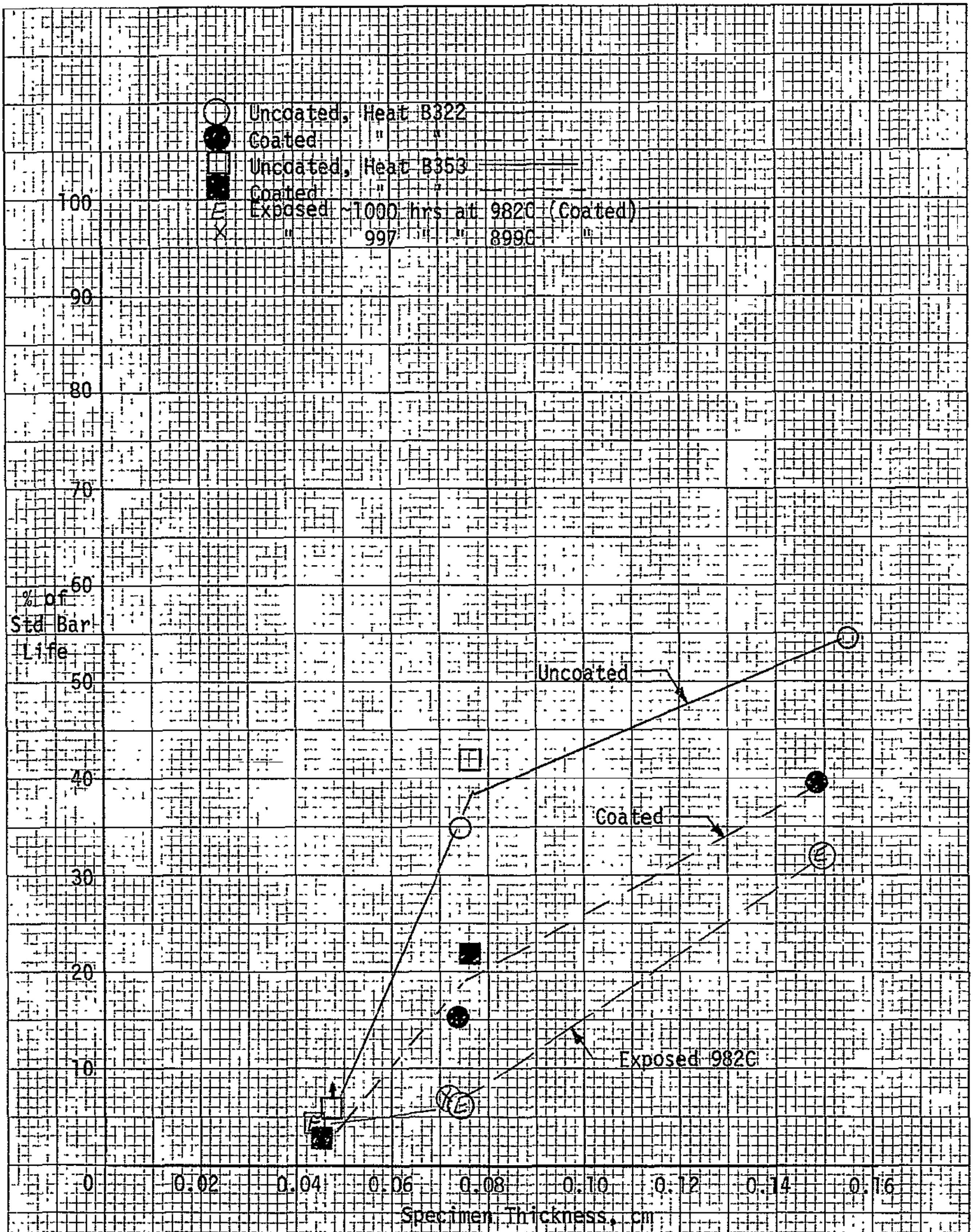


Fig. 18 Percent of Standard Bar Life, Rene 80, 871C, 262 MN/m²

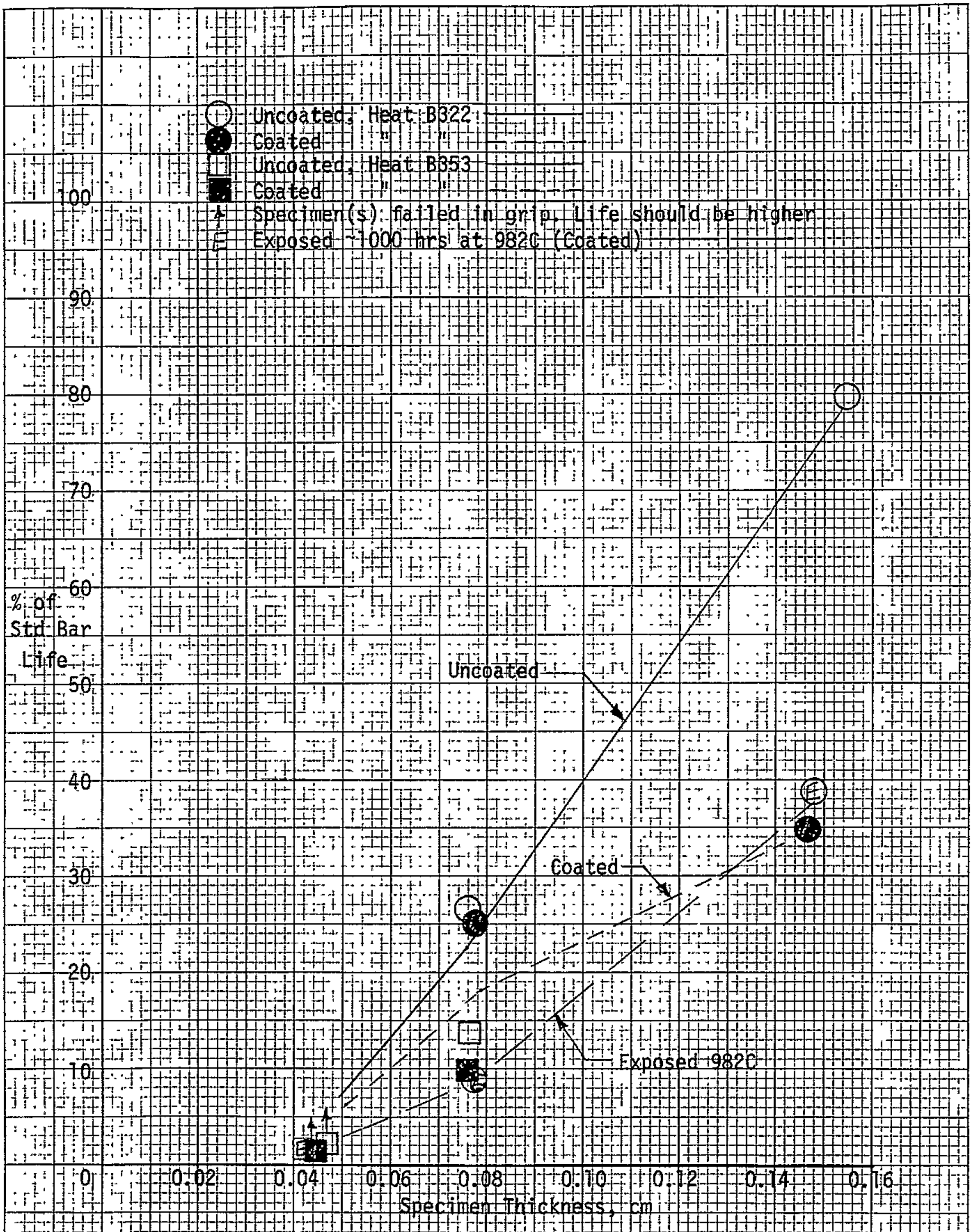


Fig. 19 Percent of Standard Bar Life, Rene 80, 982C, 144 MN/m²

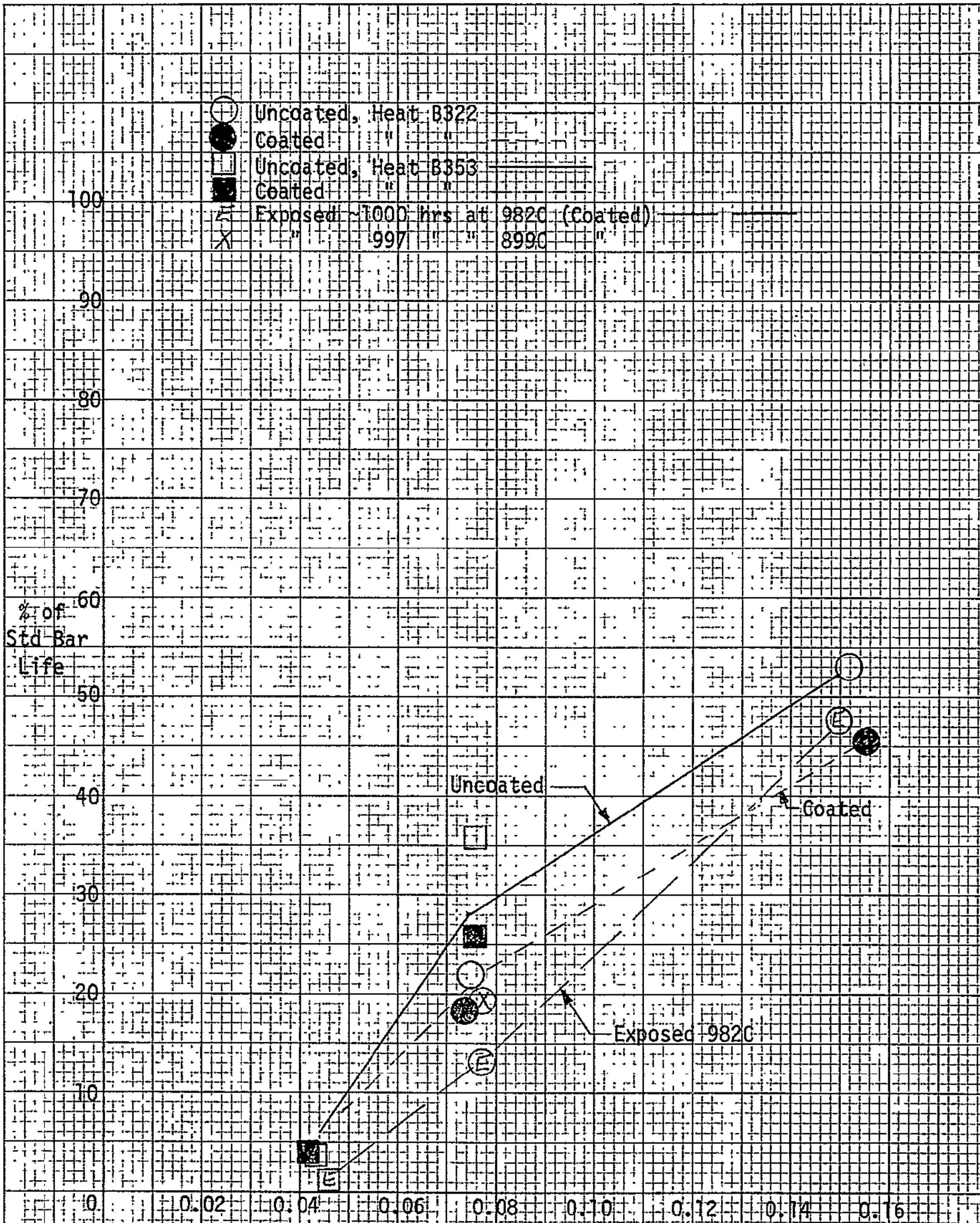


Fig. 20 Percent of Standard Bar Life, Rene 80, 982C, 117 MN/m^2

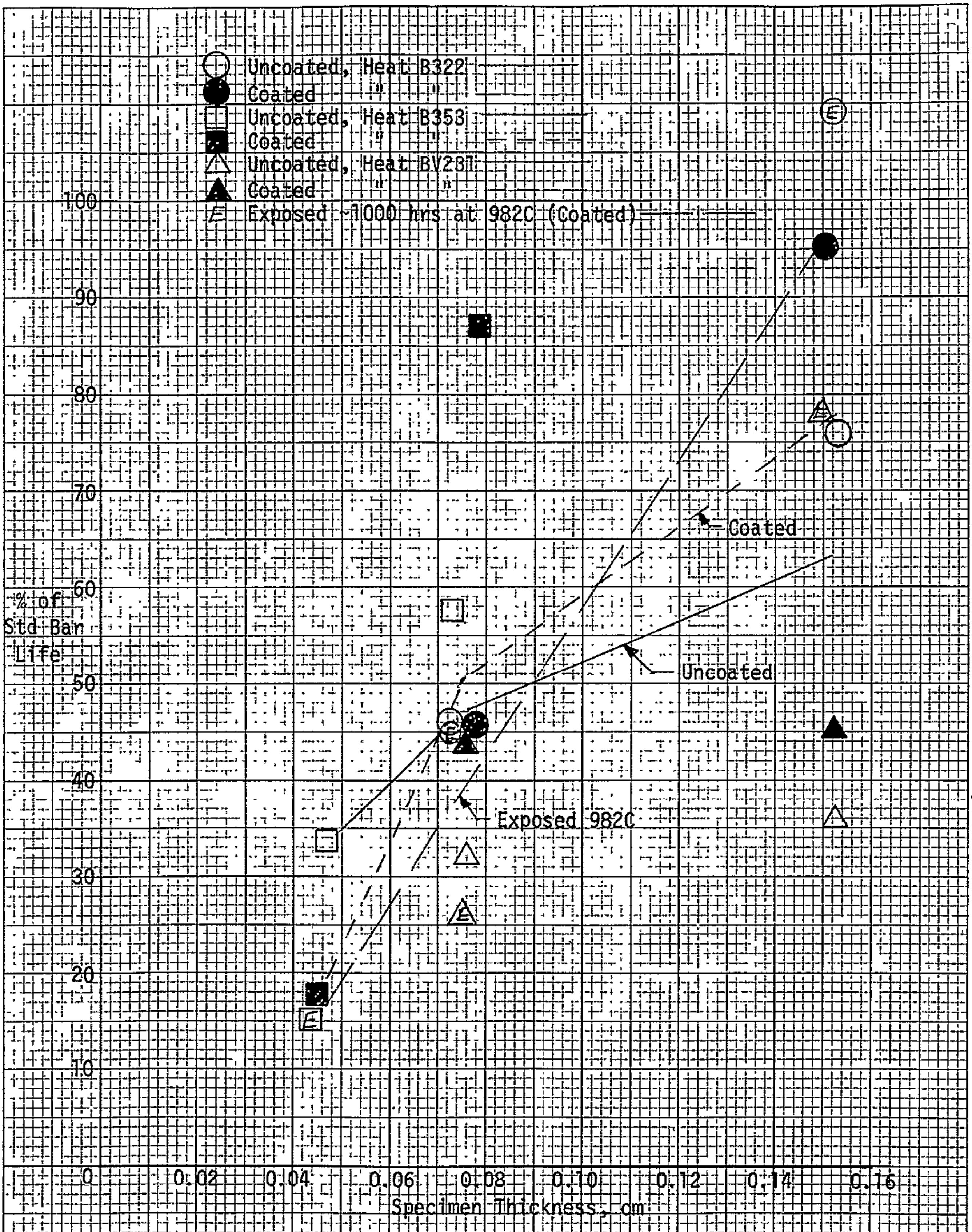


Fig. 21 Percent of Standard Bar Life, Rene 80, 1093C, 34.5 MN/m²

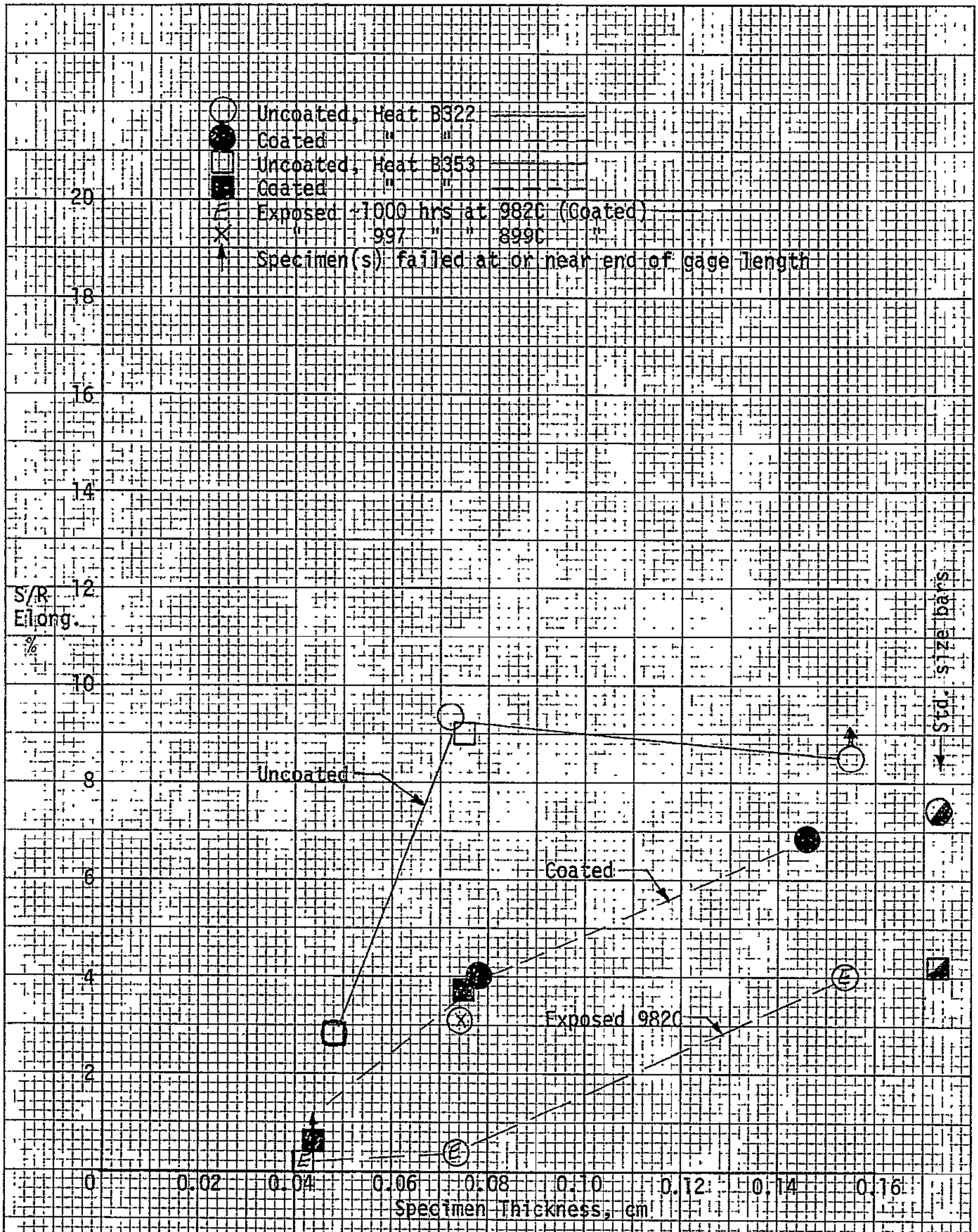


Fig. 22 Stress Rupture Elongation, Rene 80, 760C, 565 MN/m²

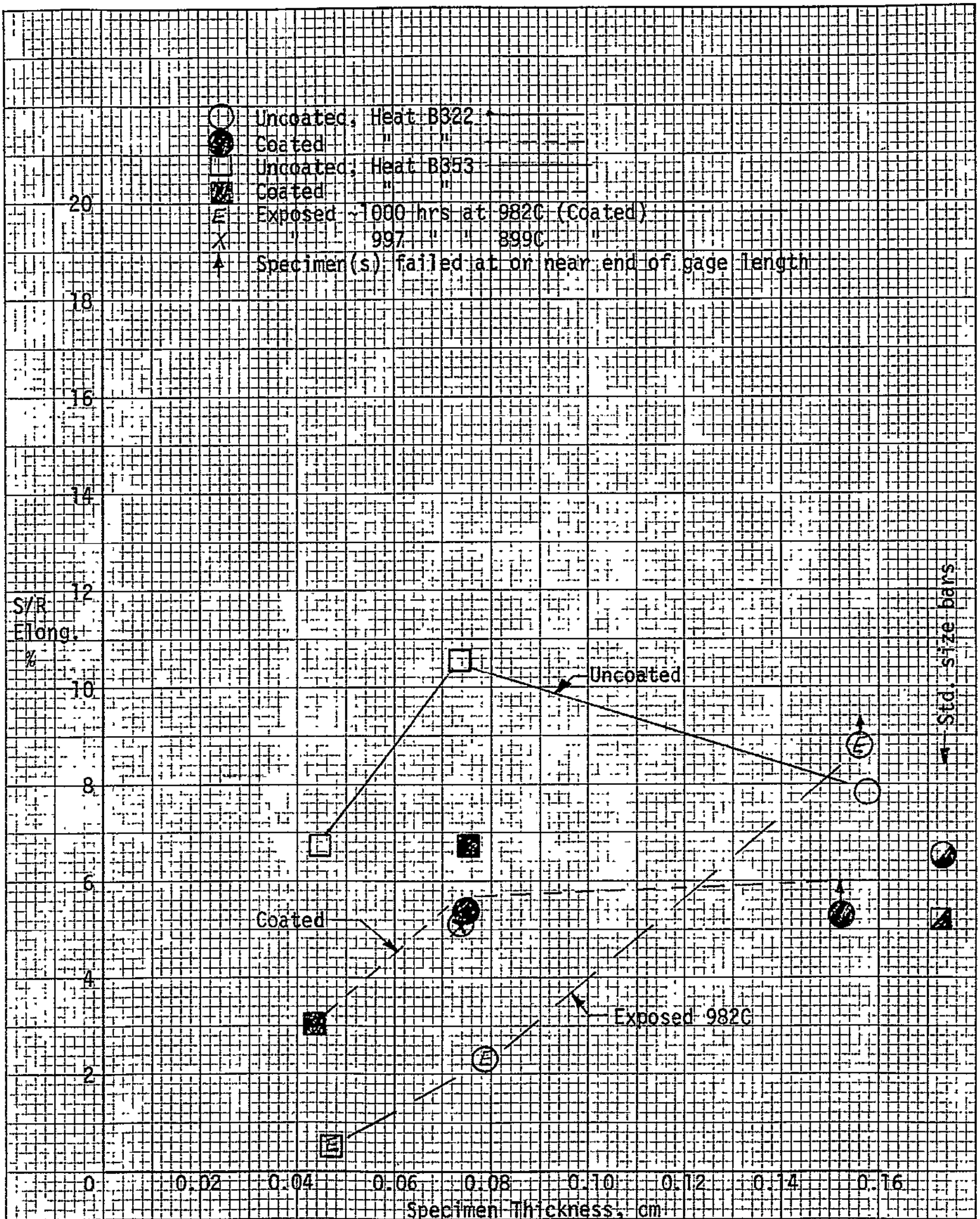


Fig. 23 Stress Rupture Elongation, Rene 80, 871C, 310 MN/m²

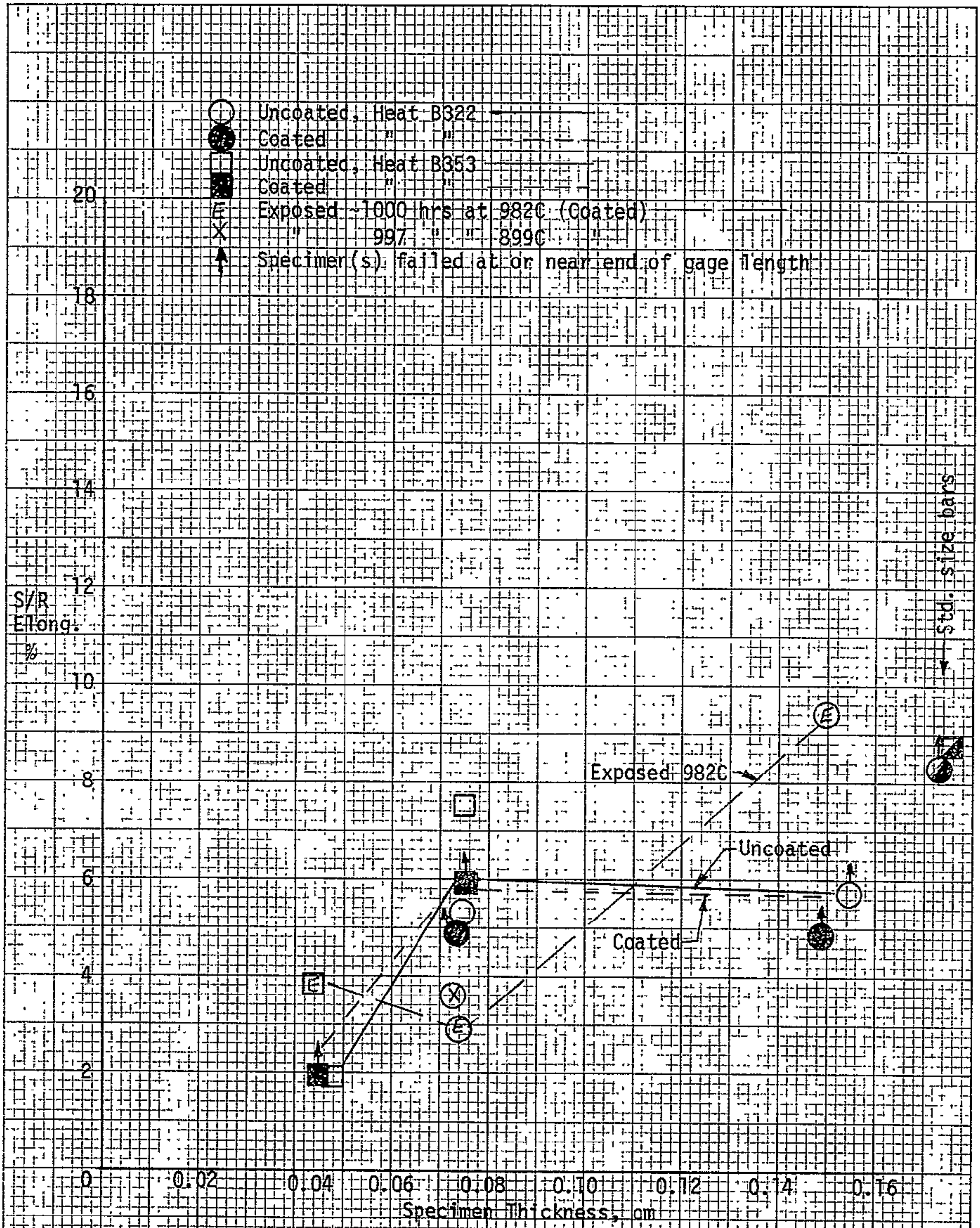


Fig. 24 Stress Rupture Elongation, Rene 80, 871C, 262 MN/m²

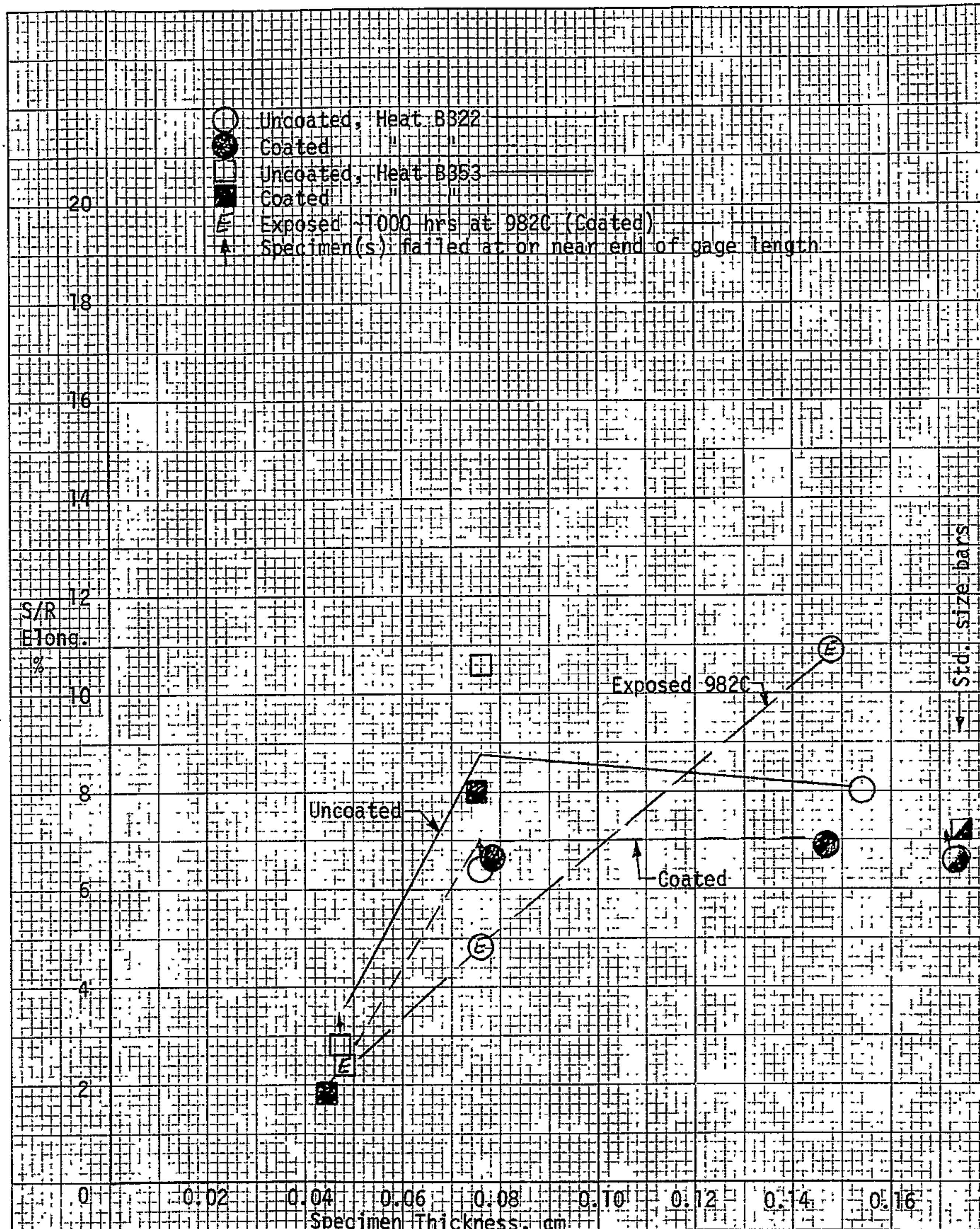


Fig. 25 Stress Rupture Elongation, Rene 80, 982C, 144 MN/m²

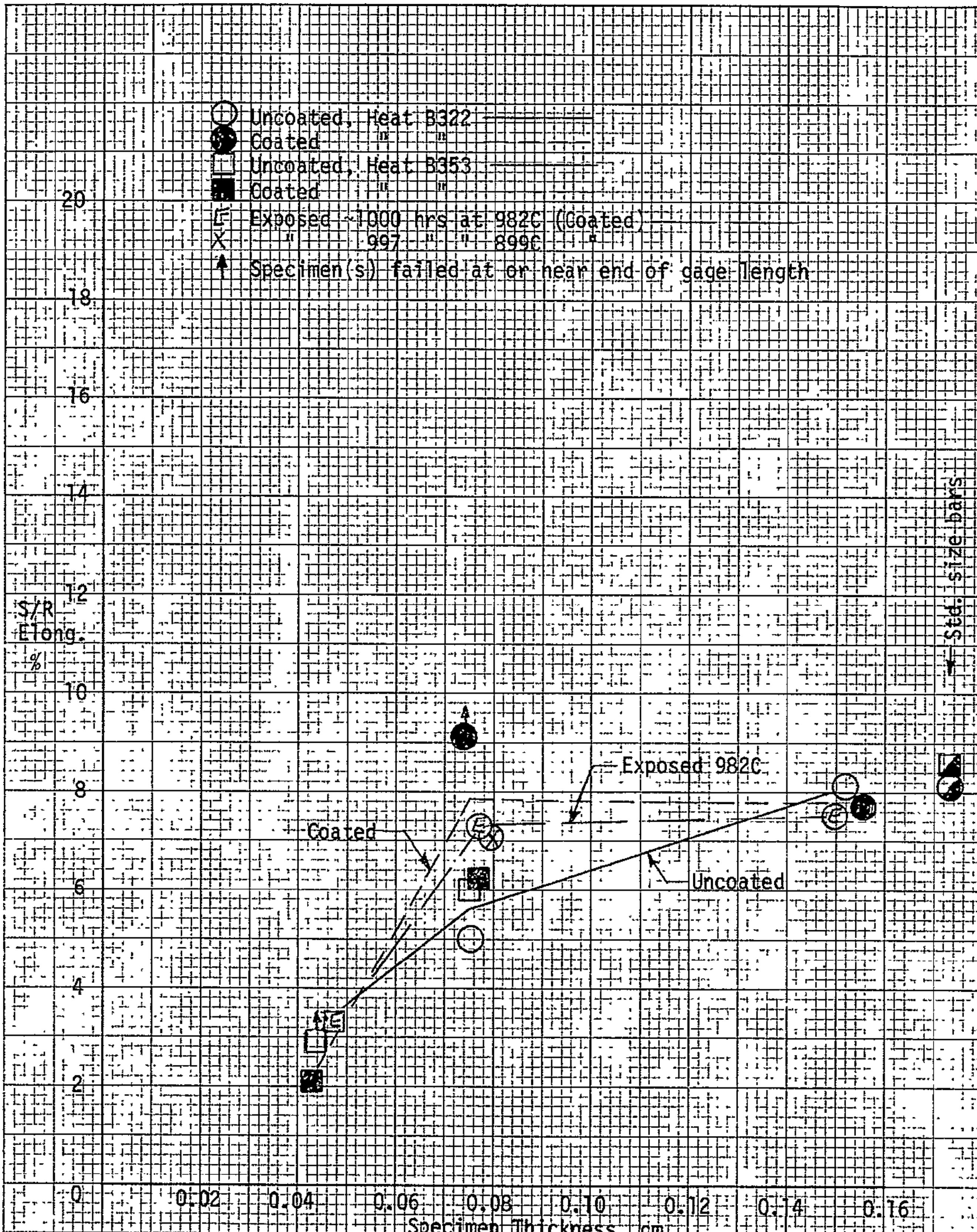


Fig. 26 Stress Rupture Elongation, Rene 80, 982C, 117 MN/m²

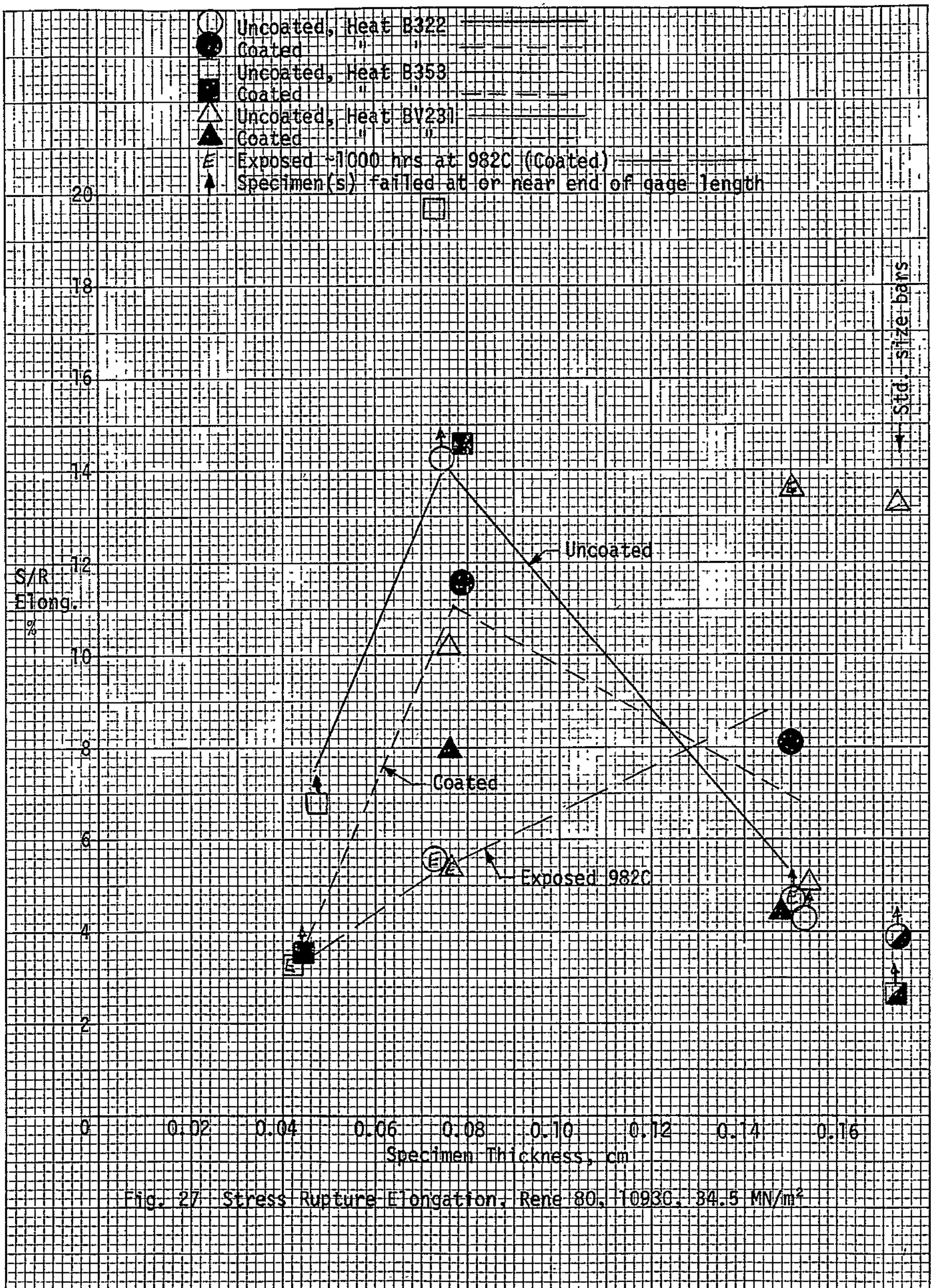


Fig. 27 Stress Rupture Elongation, Rene 80, 1093C, 34.5 MN/m²

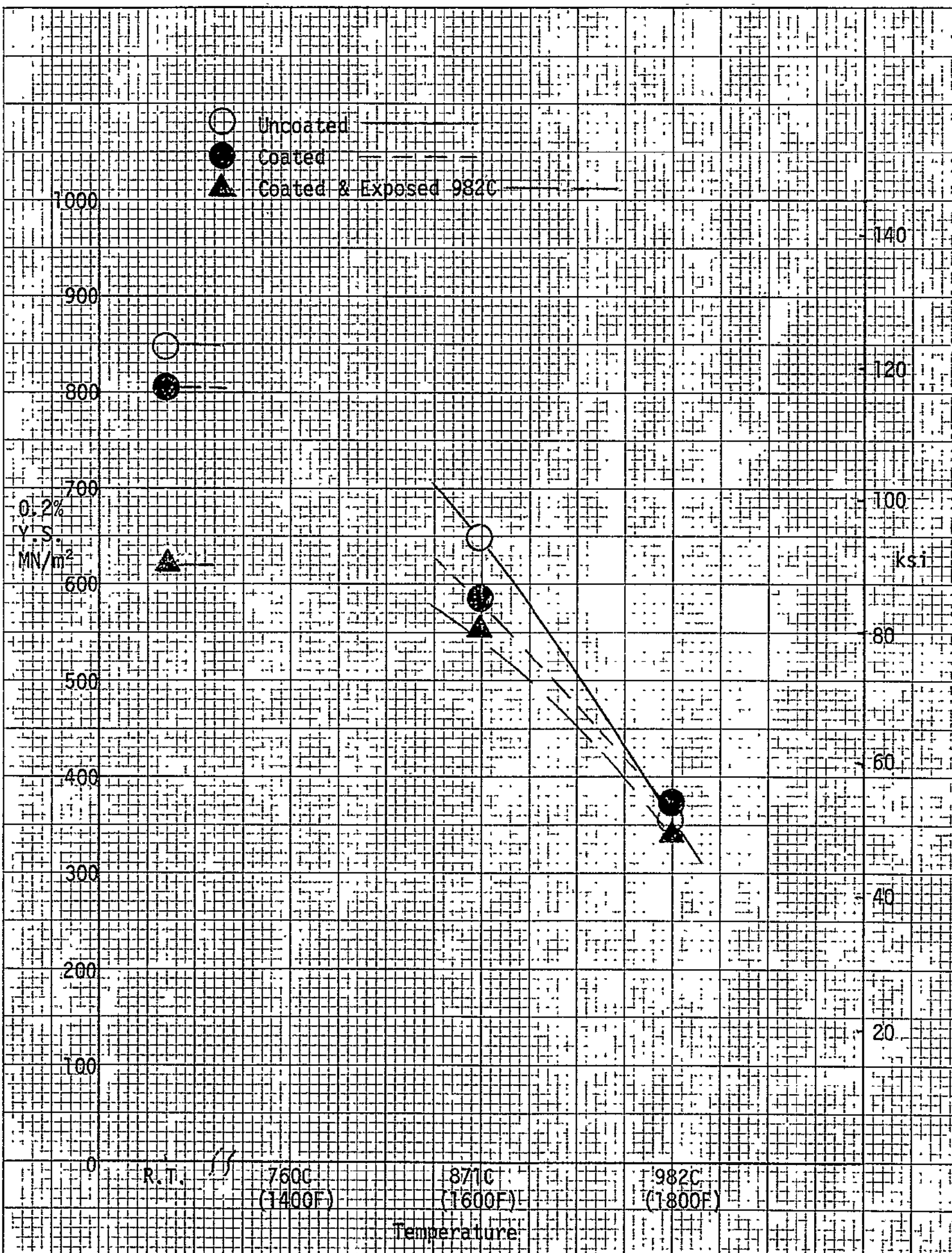


Fig. 28 0.2% Yield Strength, Rene 120, 0.075 cm Thick

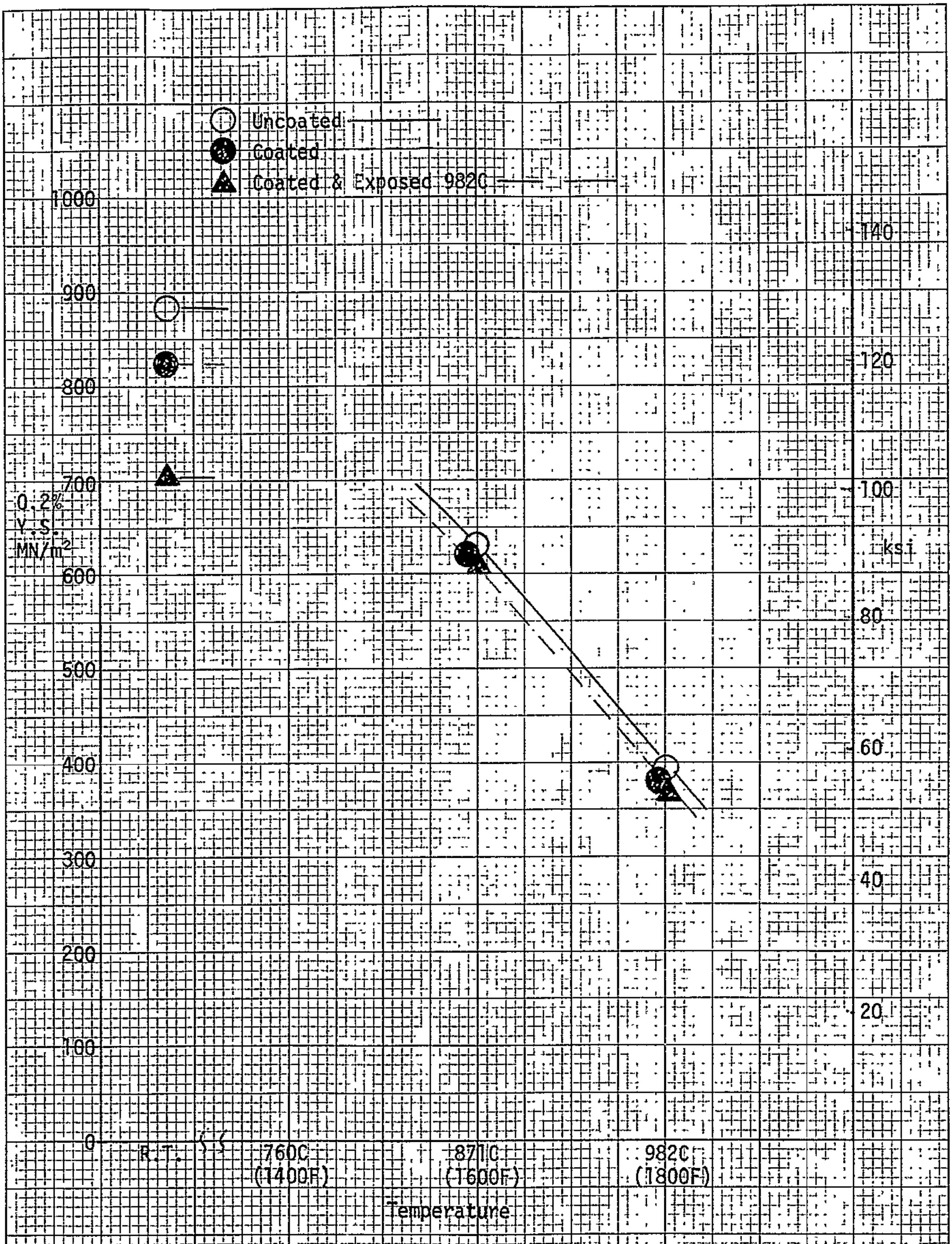


Fig. 29 0.2% Yield Strength, Rene 120, 0.15 cm Thick

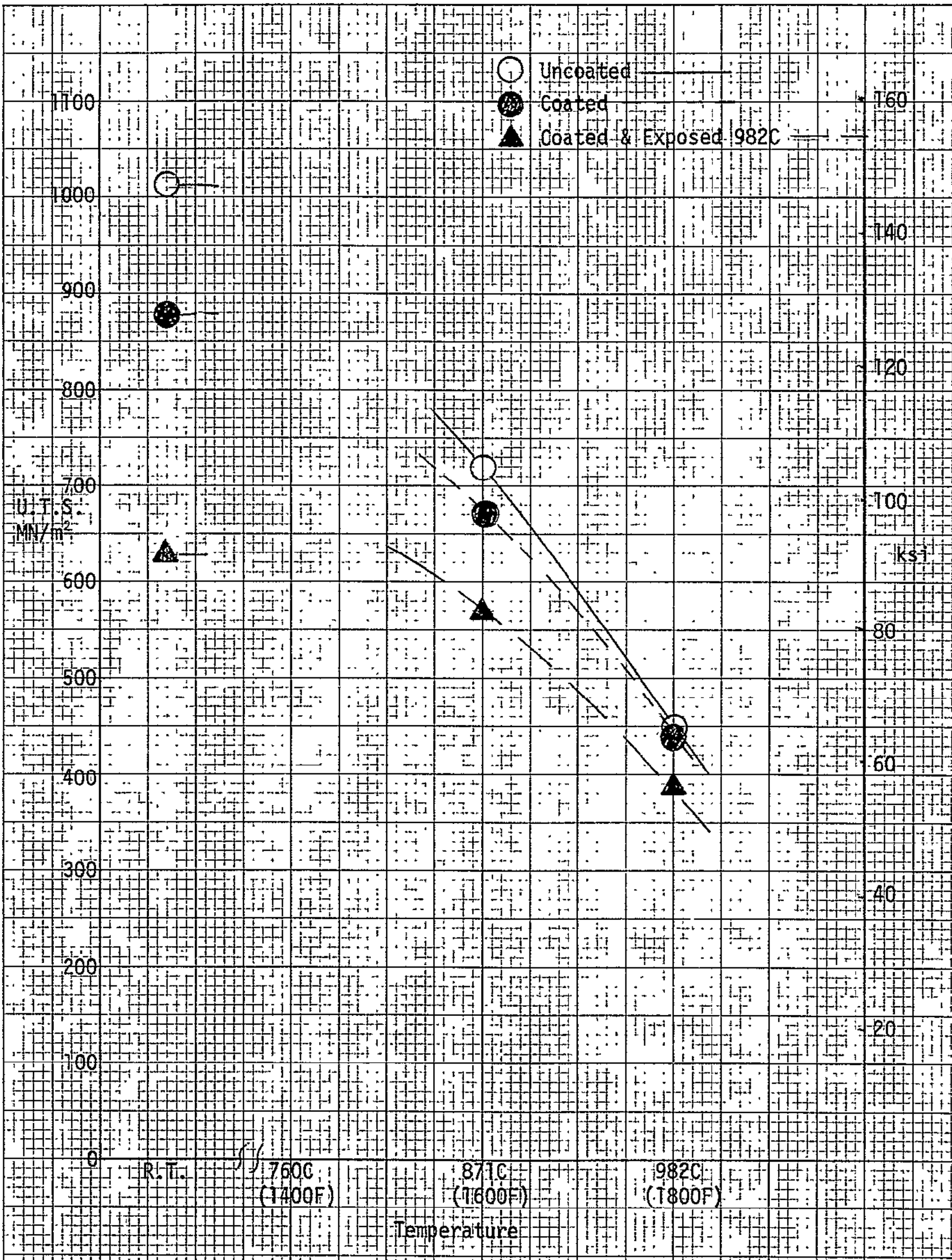


Fig. 30 Ultimate Tensile Strength, Rene 120, 0.075cm Thick

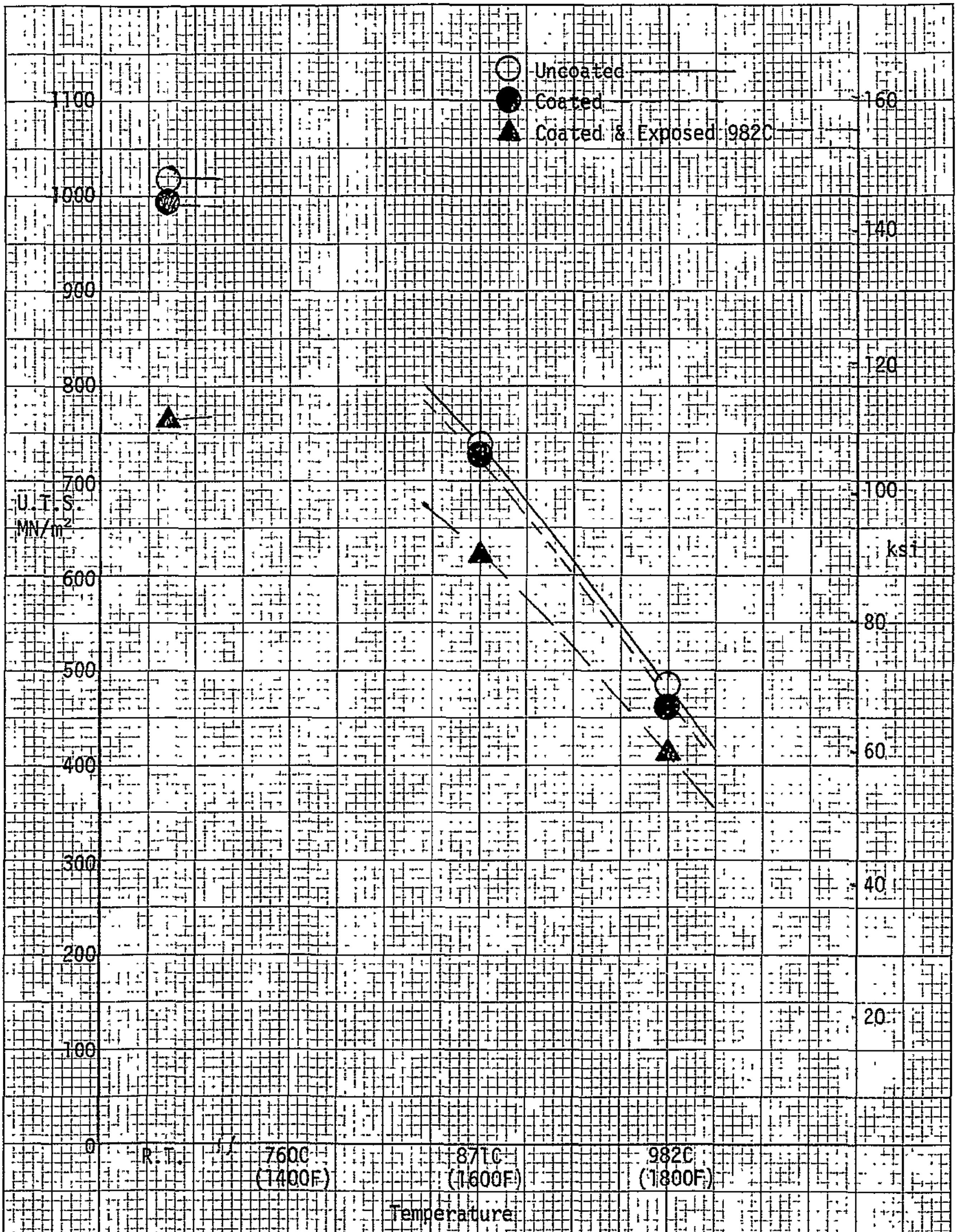


Fig. 31 Ultimate Tensile Strength, Rene 120, 0.15 cm Thick

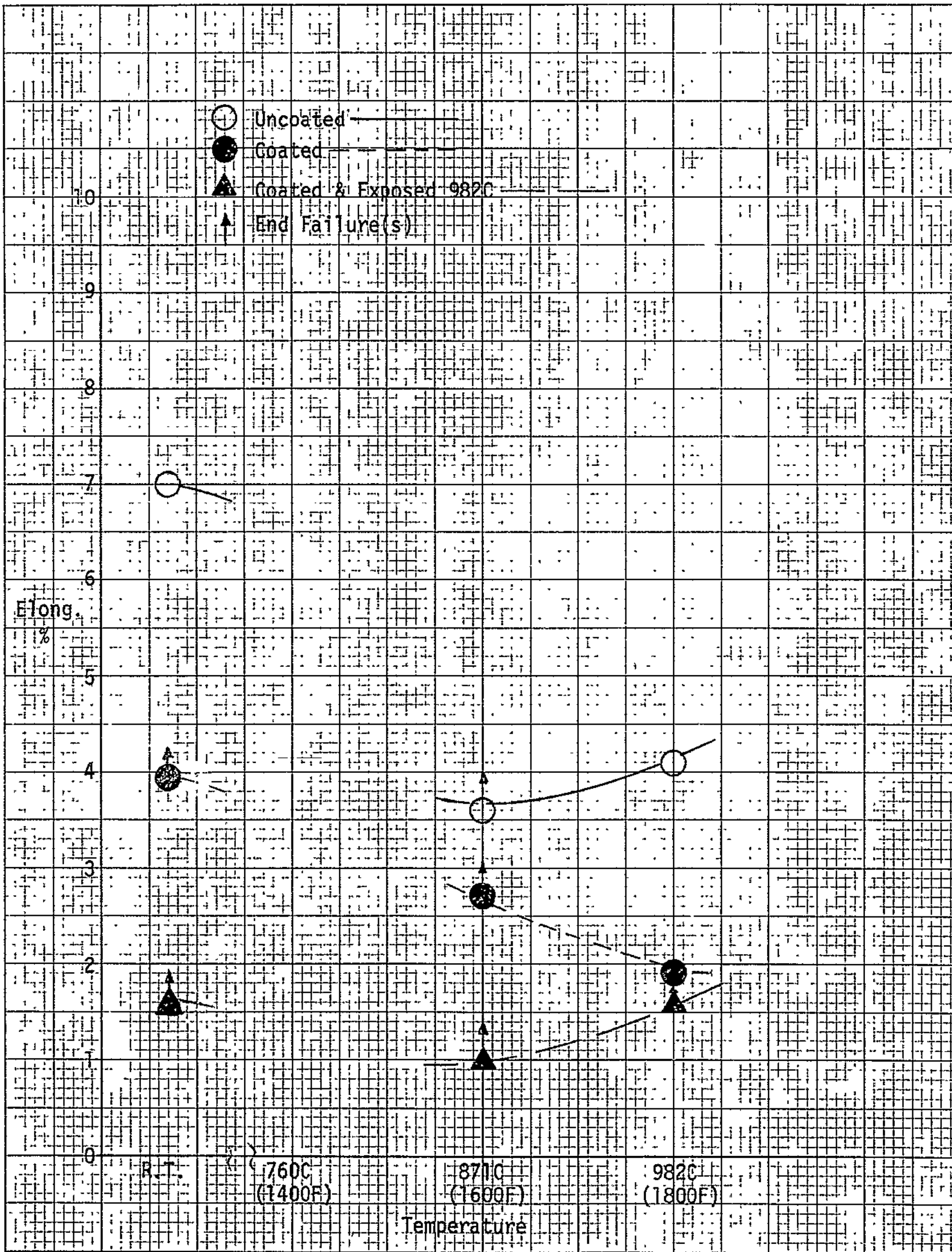


Fig. 32 Tensile Elongation, Rene 120, 0.075 cm Thick

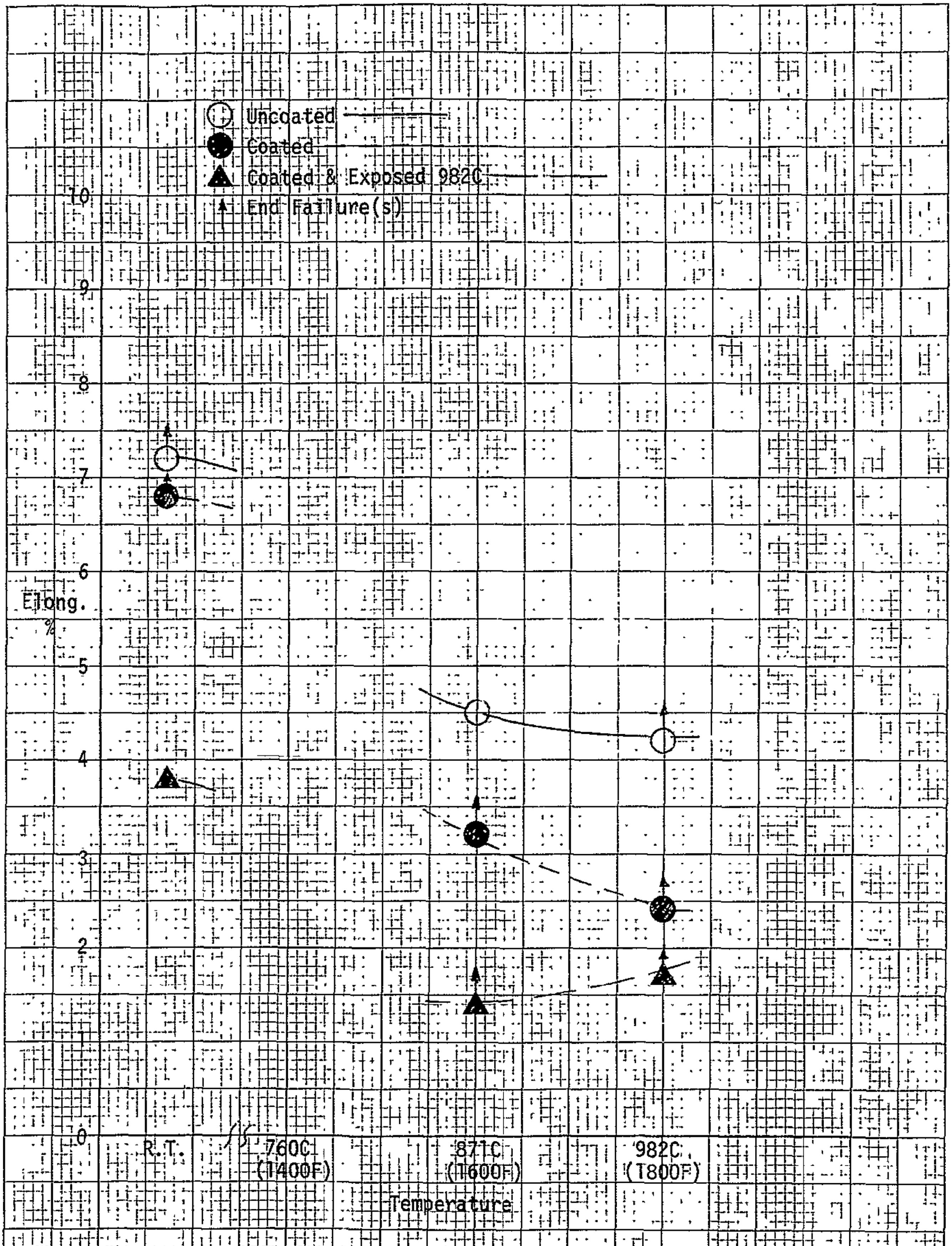


Fig. 33 Tensile Elongation, Rene 120, 0.15 cm Thick

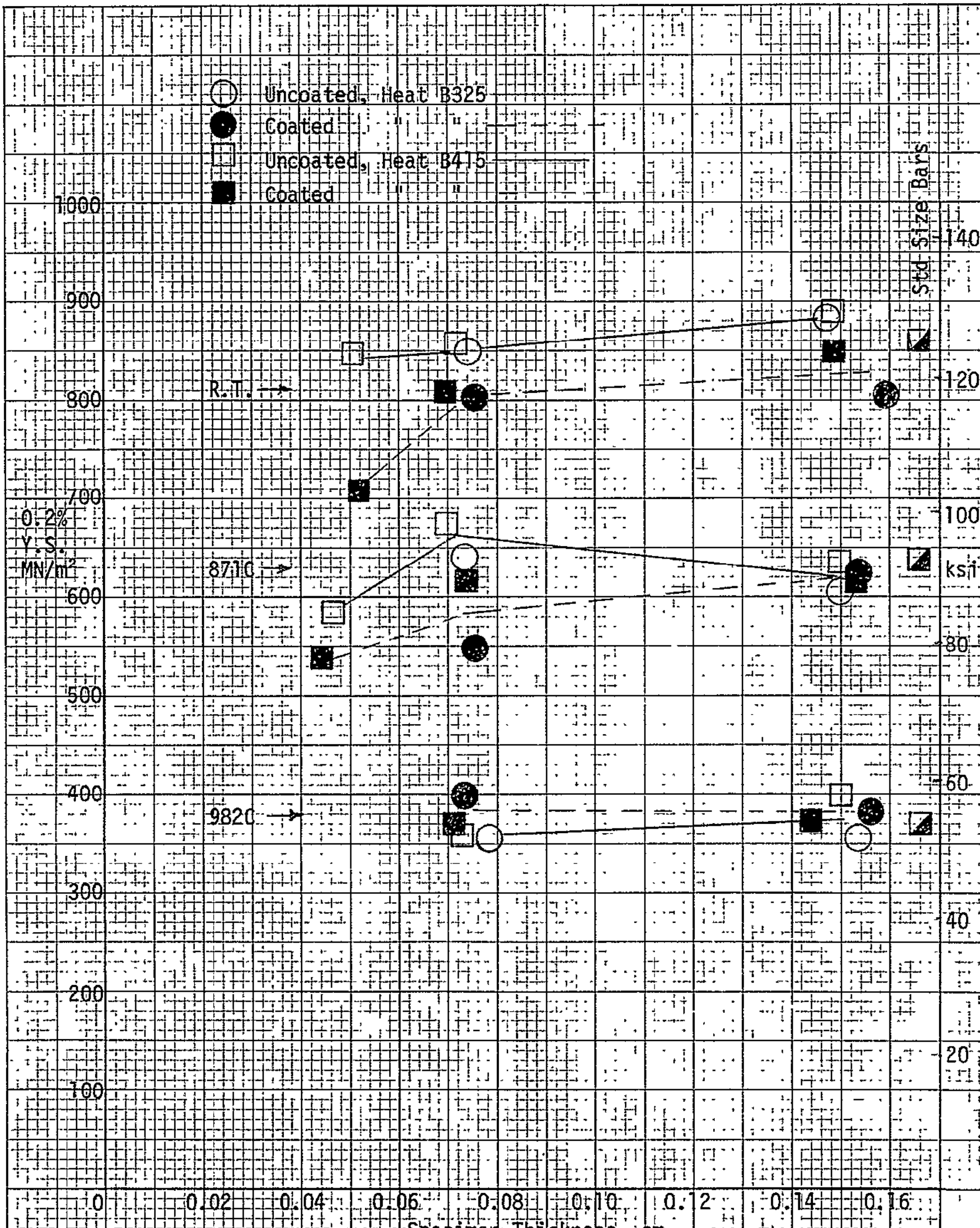


Fig.34 Effect of Section Thickness on 0.2% Yield Strength, Rene 120

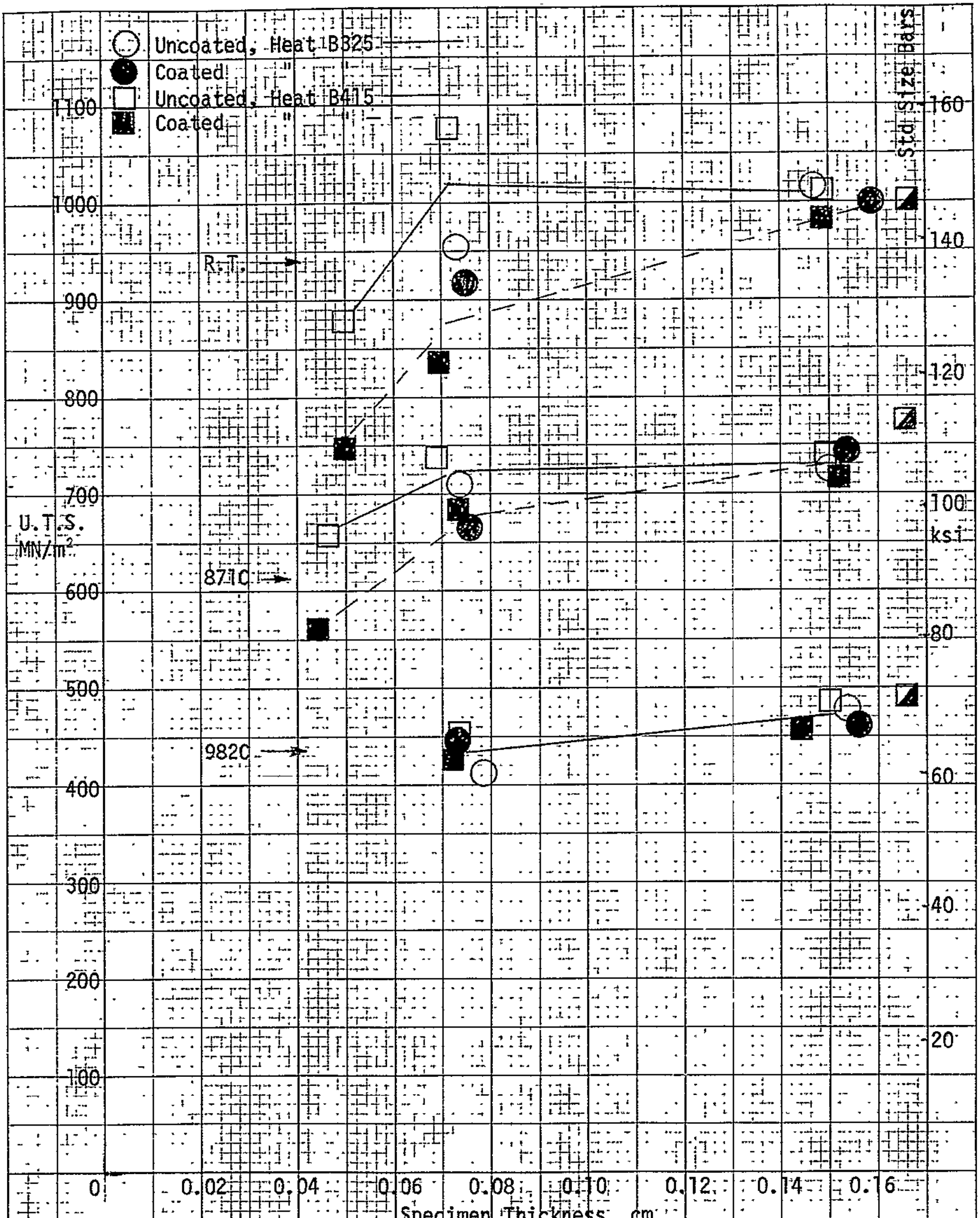


Fig. 35 Effect of Section Thickness on Ultimate Tensile Strength, Rene 120

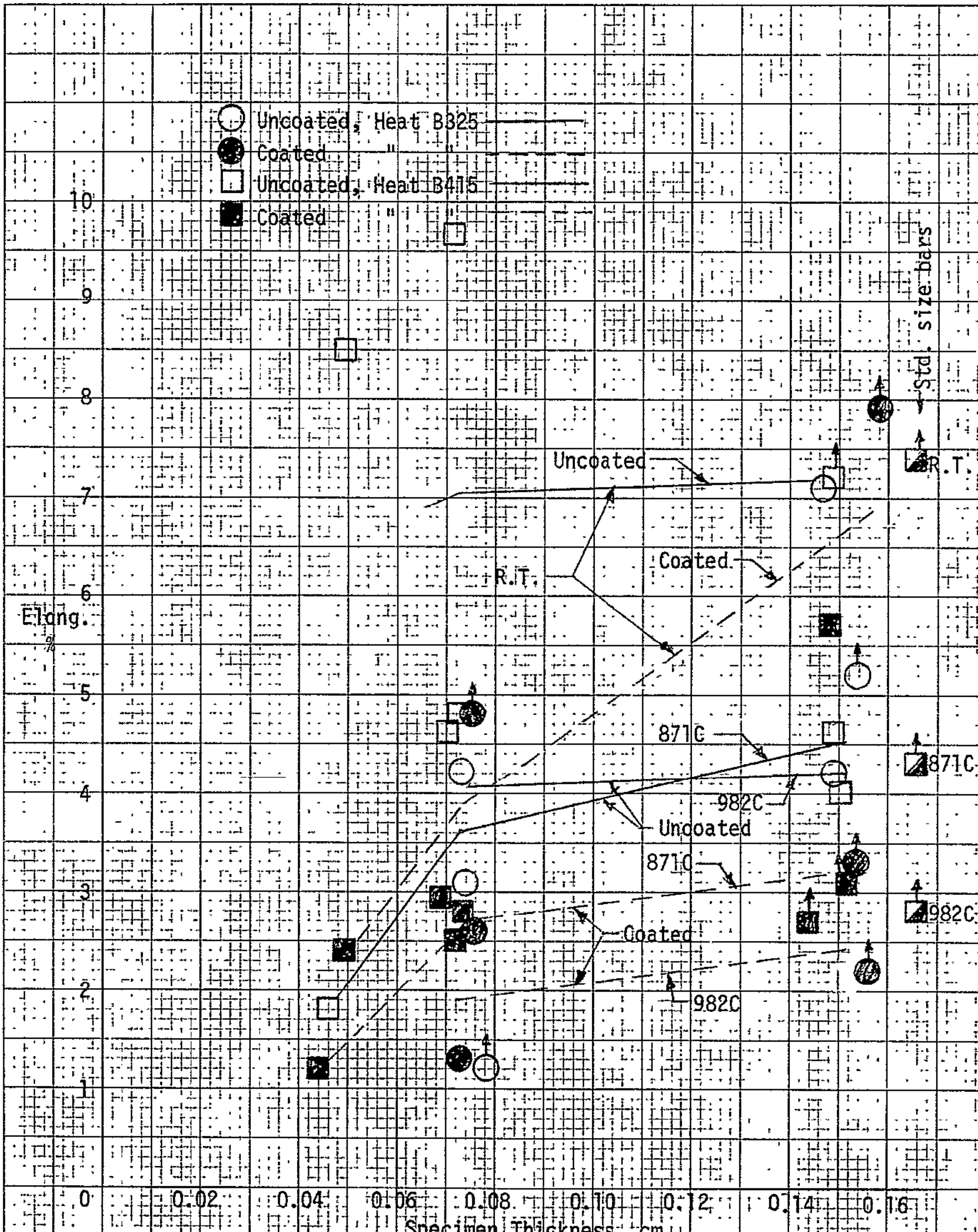


Fig. 36 Effect of Section Thickness on Tensile Elongation, Rene 120

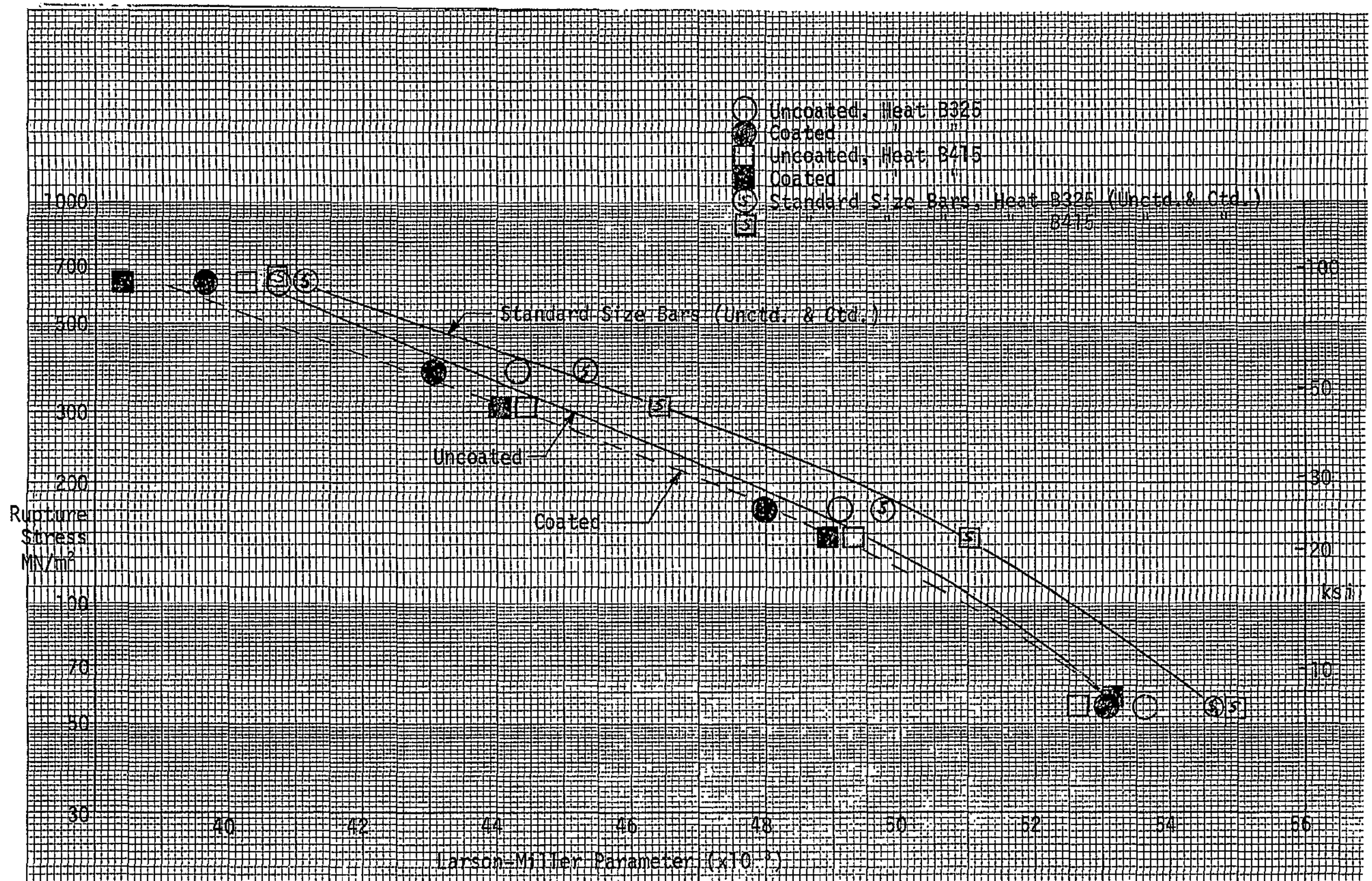


Fig. 37 Stress Rupture Strength of 0.075 cm and Standard Bars, Rene 120

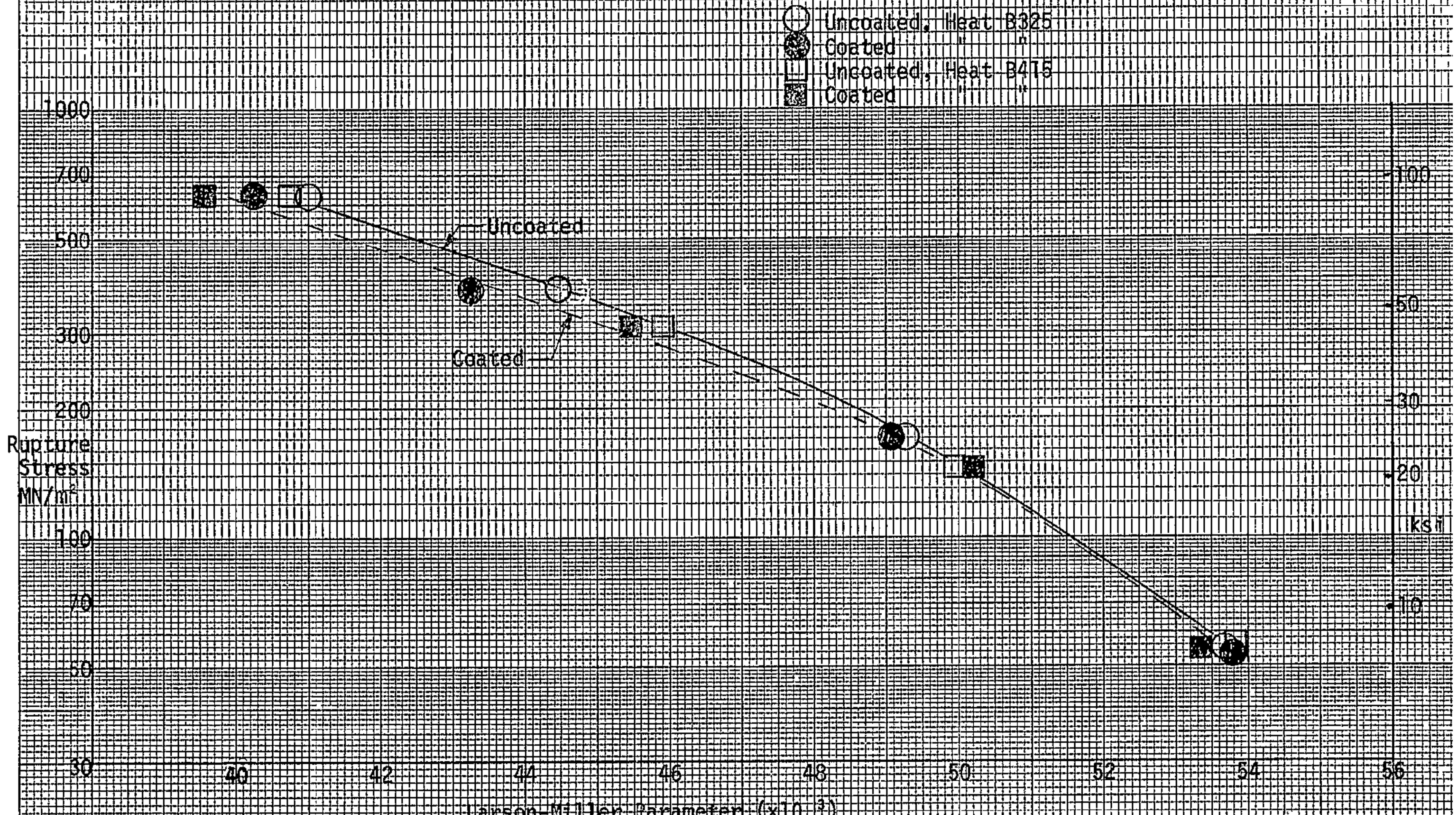


Fig. 38 Stress Rupture Strength of 0.75 cm Thick Rene 120

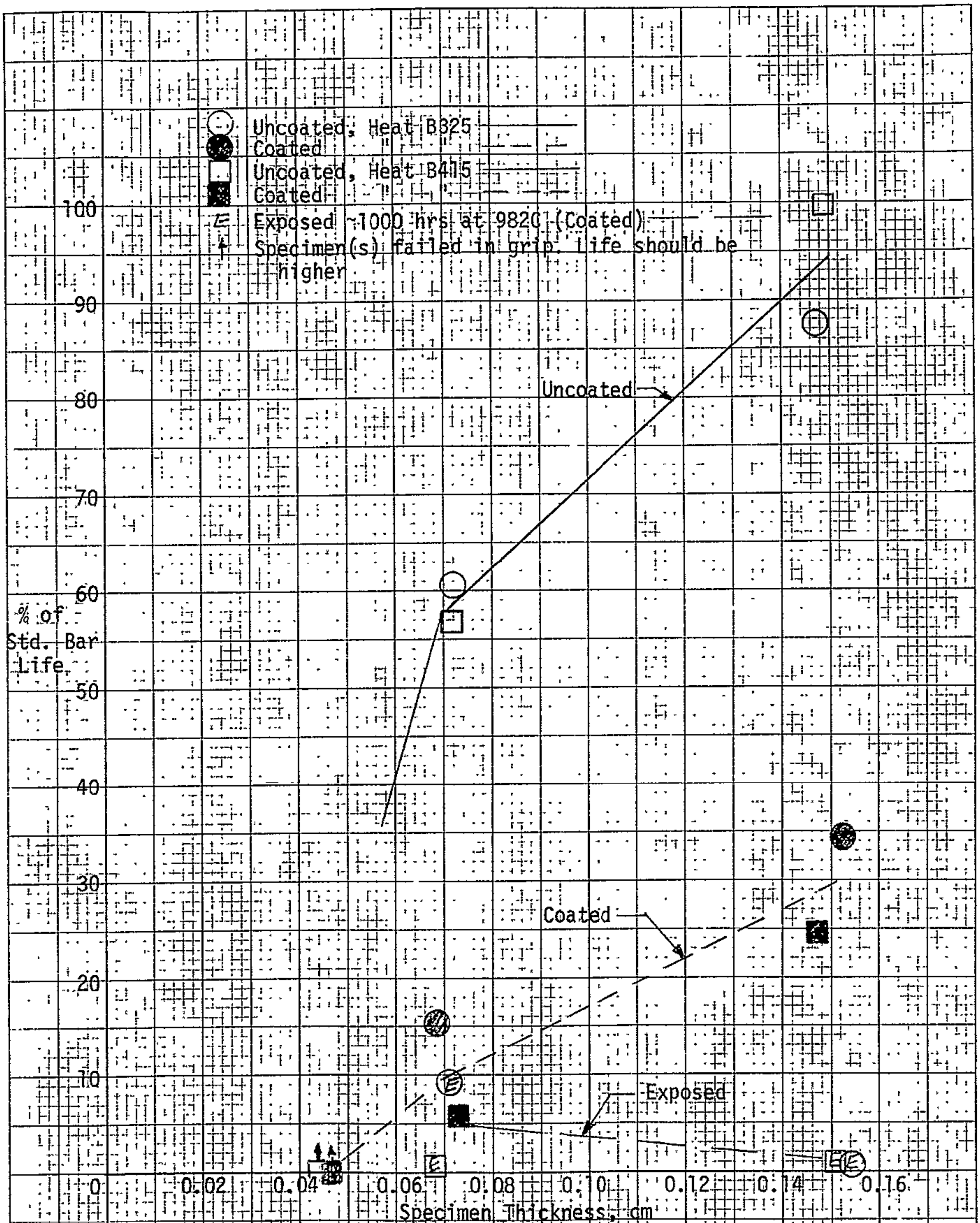


Fig. 39 Percent of Standard Bar Life, Rene 120, 760C, 634 MN/m²

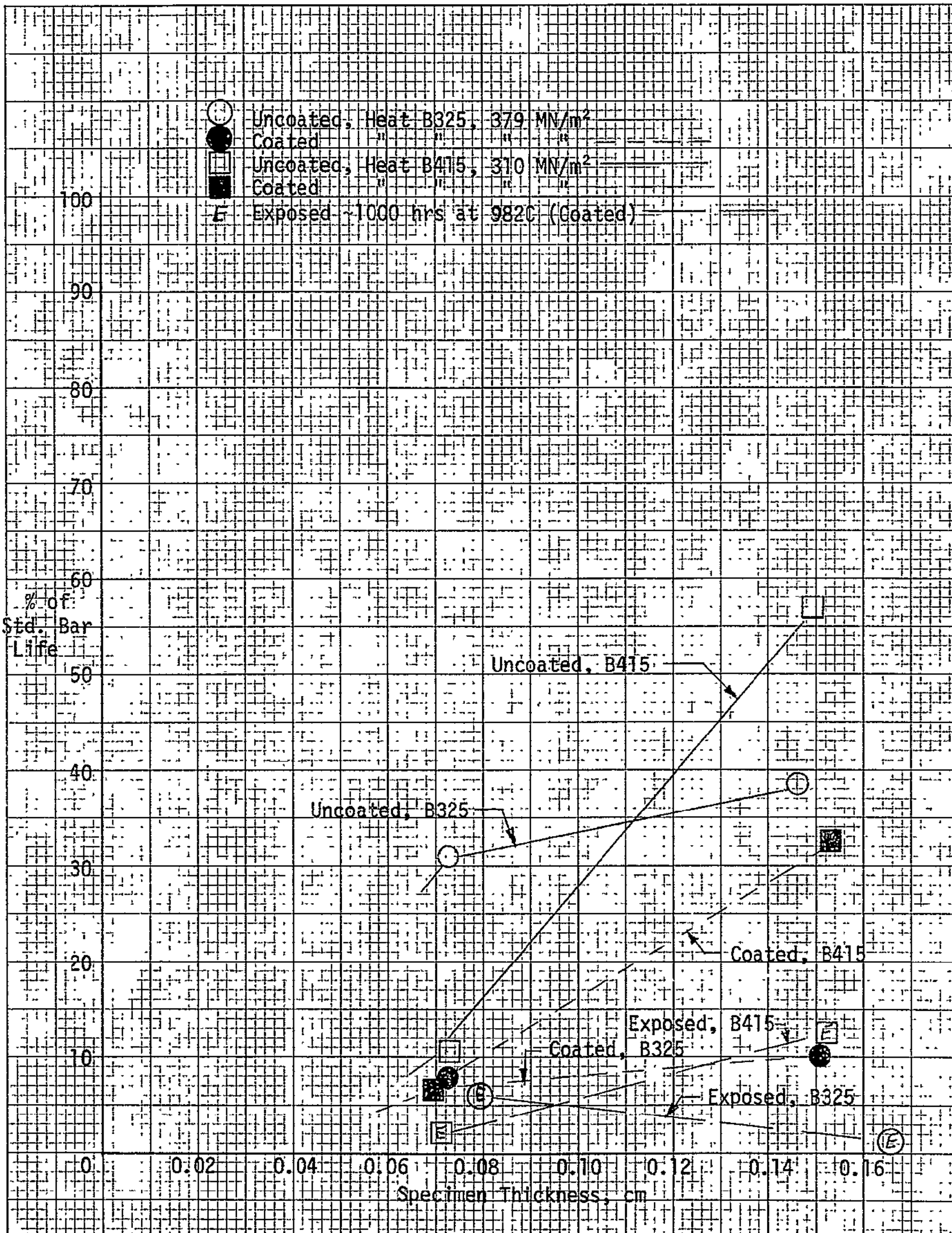


Fig. 40 Percent of Standard Bar Life, Rene 120, 871C

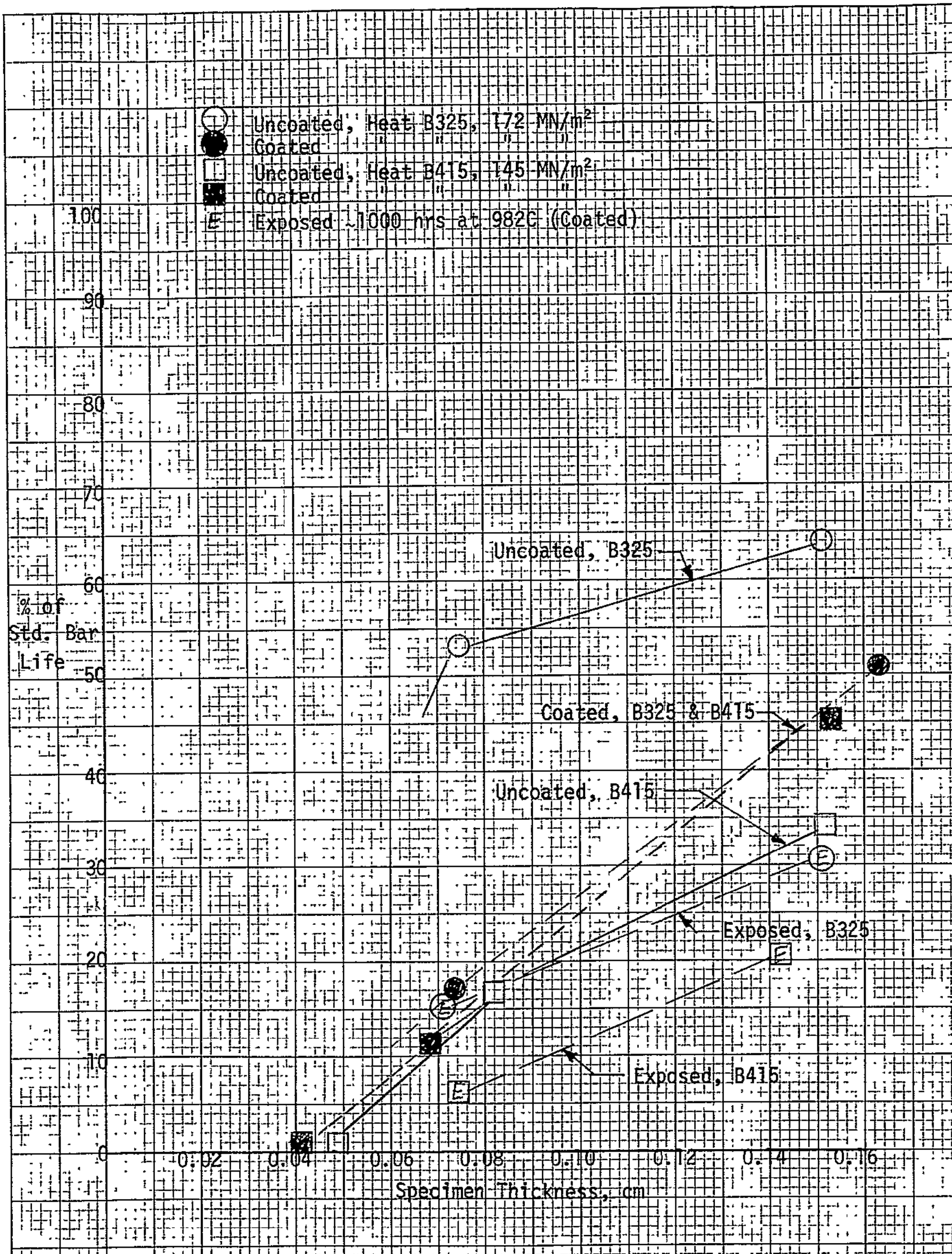


Fig. 41 Percent of Standard Bar Life, Rene 120, 982C

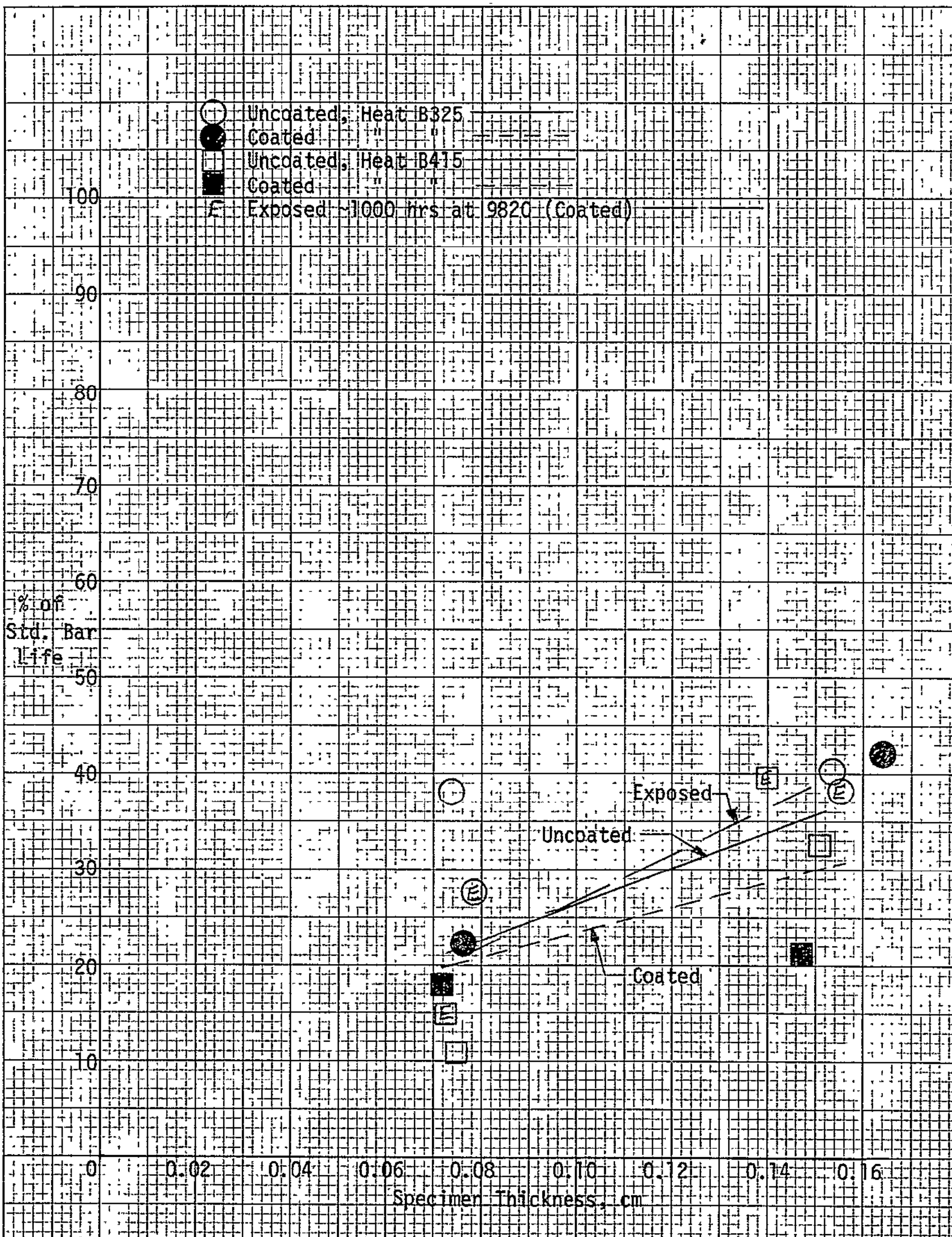


Fig. 42 Percent of Standard Bar Life, Rene 120, 1093C, 55 MN/m²

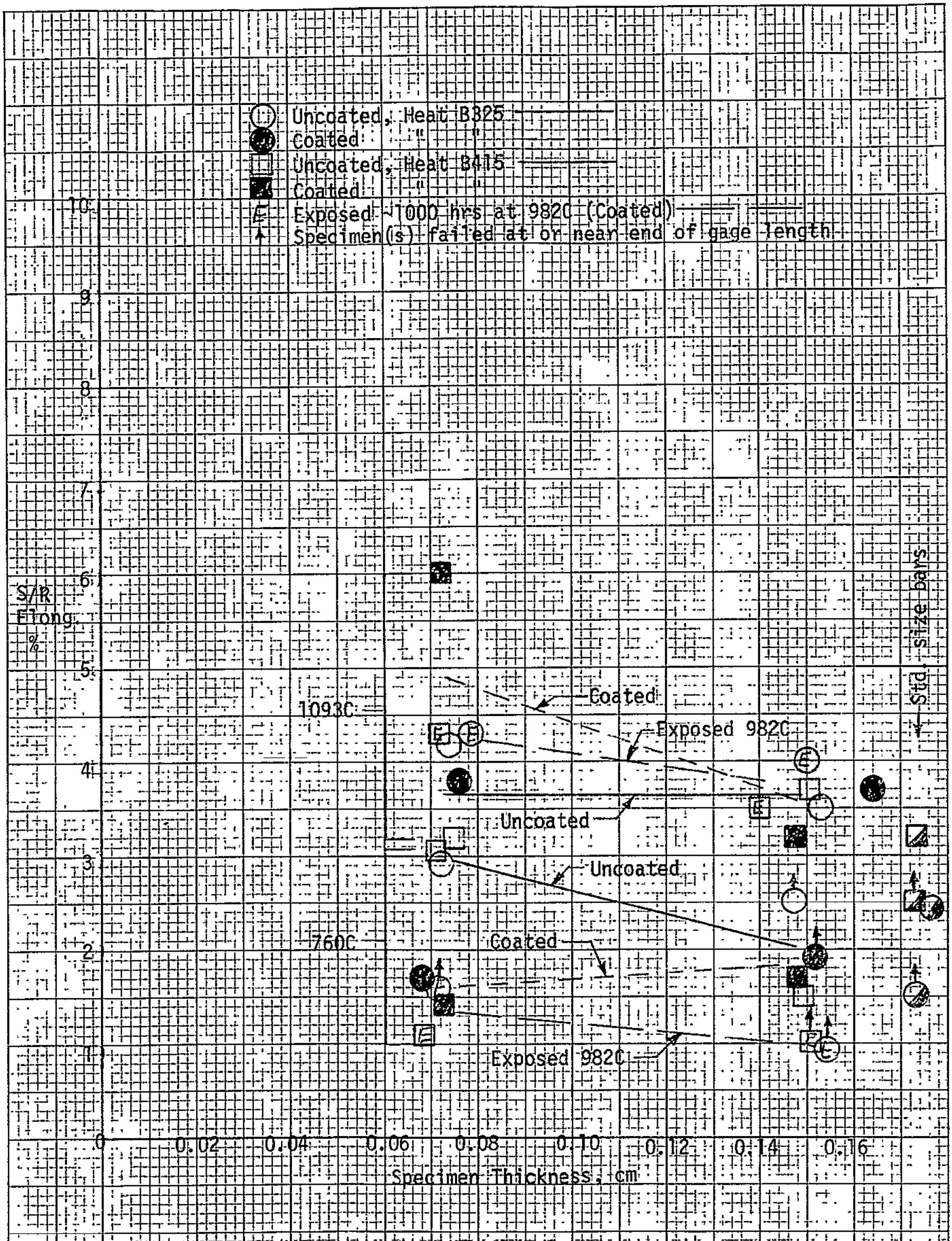


Fig. 43 Stress Rupture Elongation, Rene 120, 760C, 634 MN/m^2 and 1093C, 55 MN/m^2

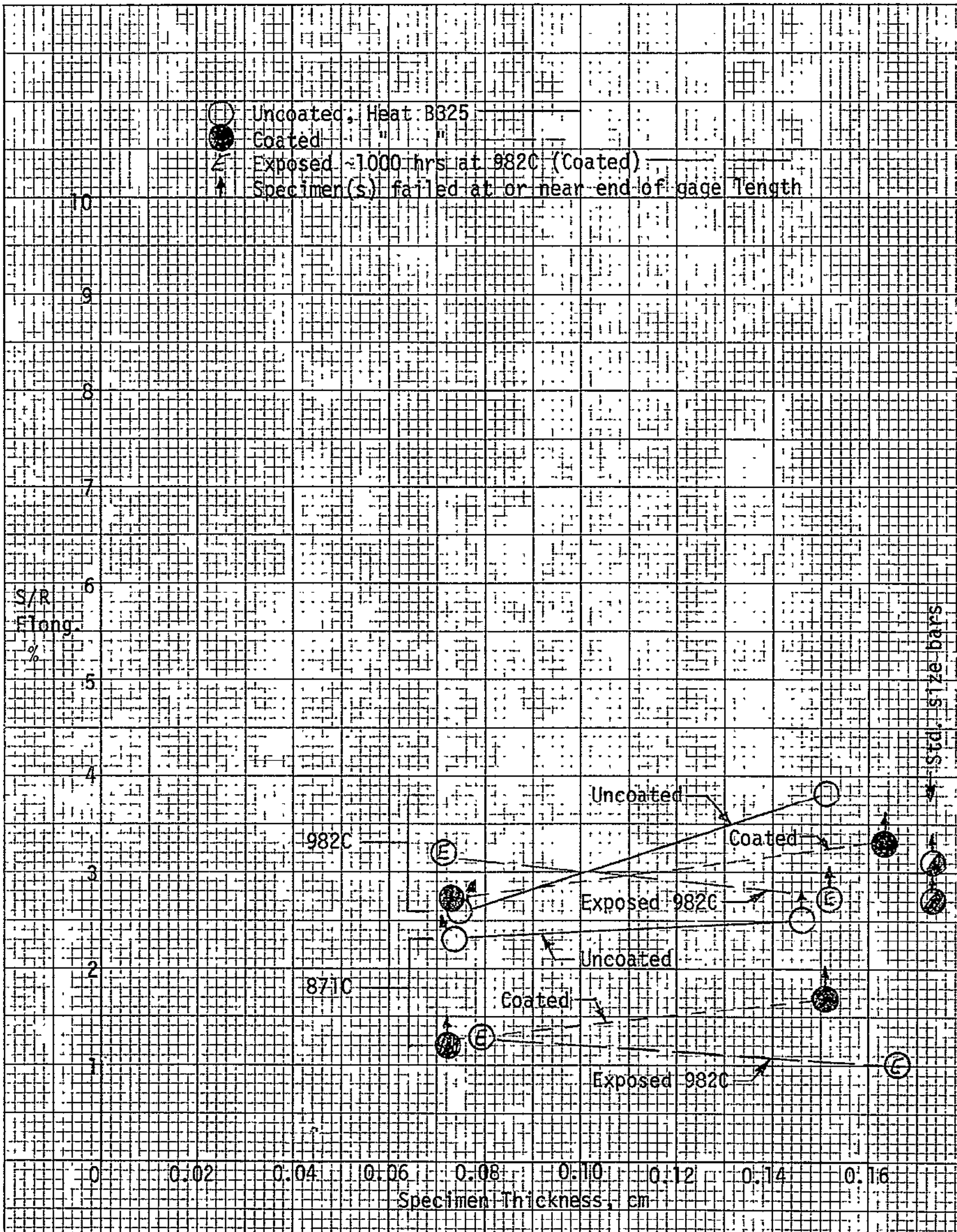


Fig. 44 Stress Rupture Elongation, Rene 120, 871C, 379 MN/m² and 982C, 172 MN/m²

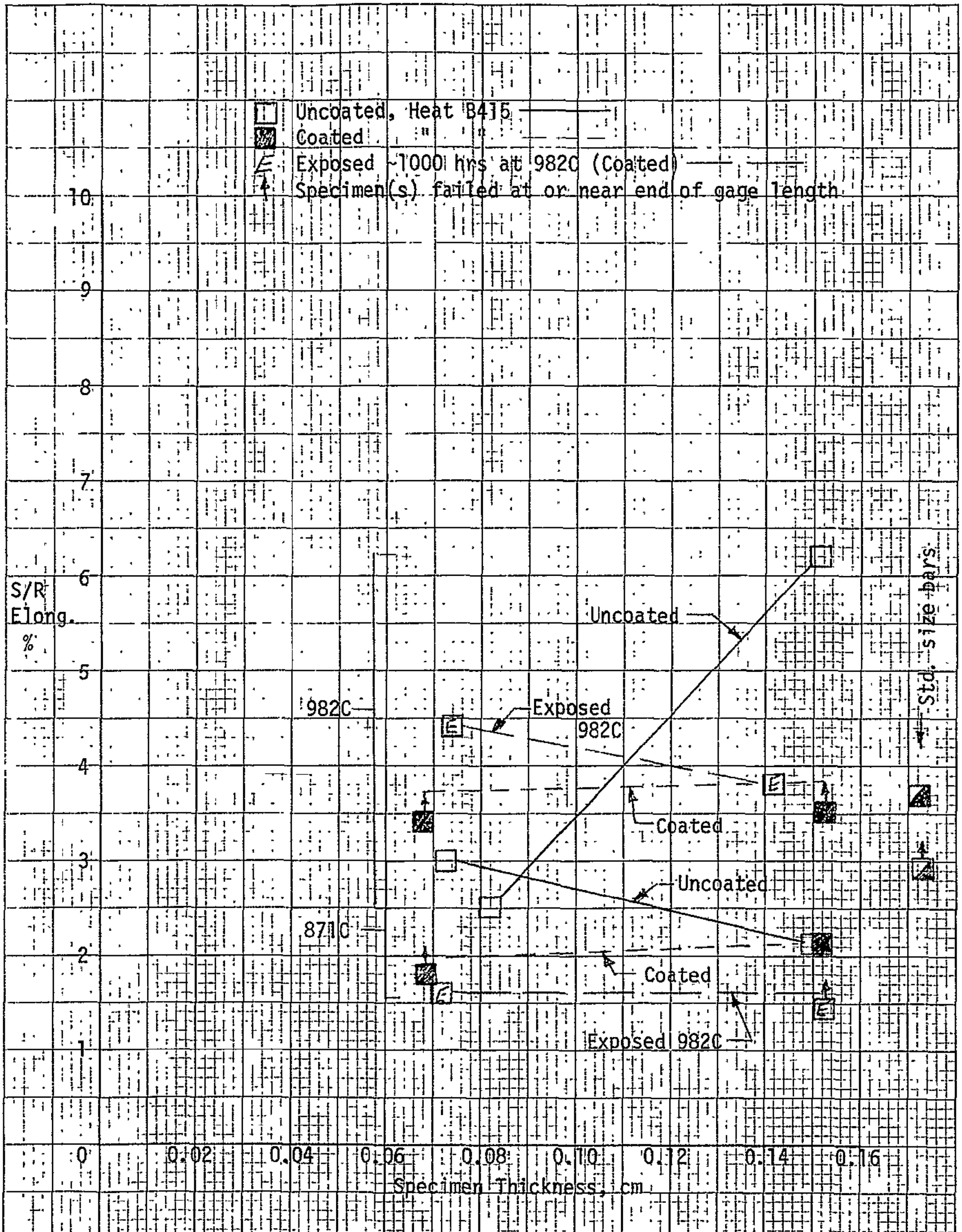
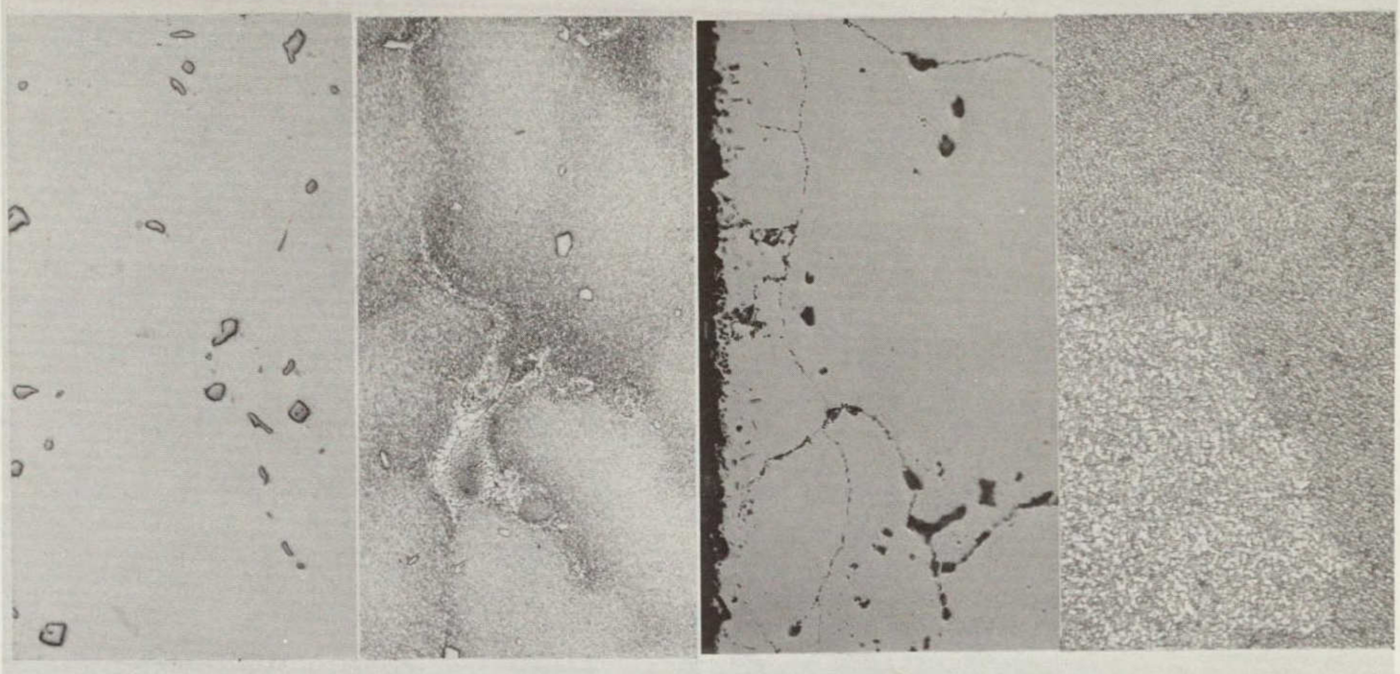


Fig. 45 Stress Rupture Elongation, Rene 120, 871C, 310 MN/m^2 and 982C, 145 MN/m^2

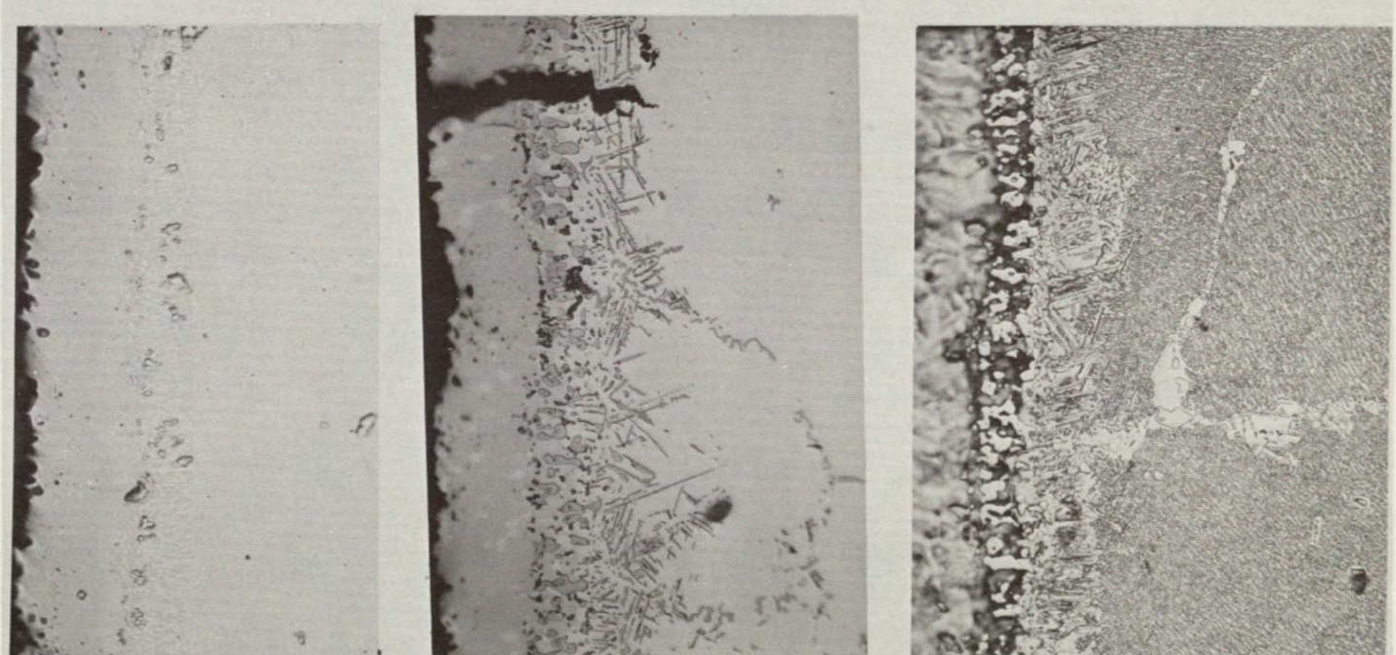


a. As-cast, unetched, MC carbides

b. As-cast, 92-5-3 etched, fine and eutectic γ'

c. Heat treated, carbide etch, recrystallized grains, $M_{23}C_6$ carbides

d. Heat treated 92-5-3 etch normal γ'



e. As-coated, unetched added layer, fingered diffusion zone

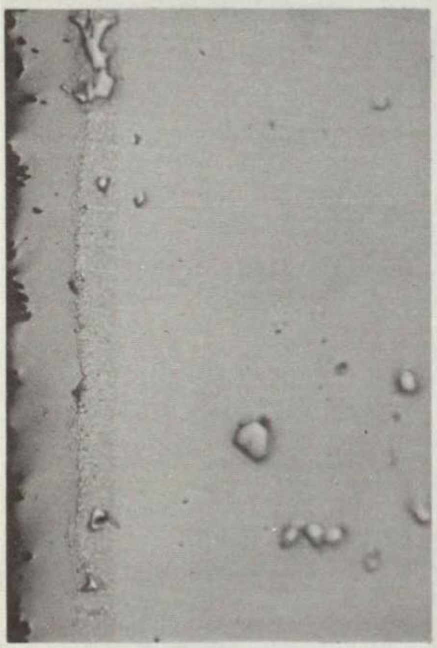
f. Coated, exposed 993 hrs, 982C, carbide etch, added layer transformation, acicular σ

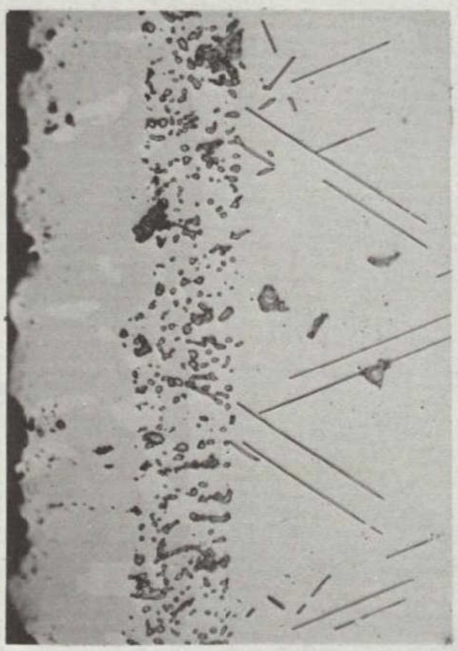
g. Same as (f), 92-5-3 etch coarse grain boundary phases below coating

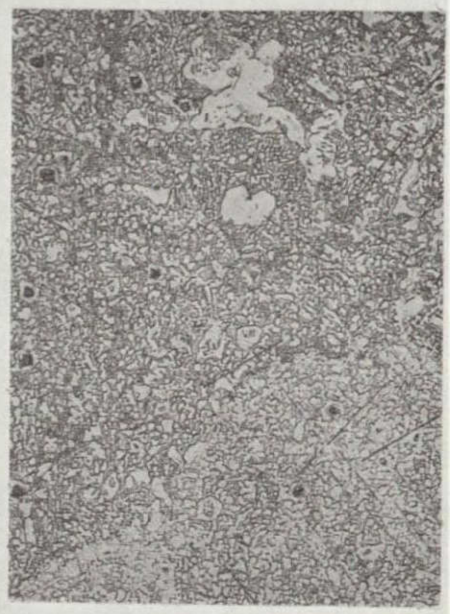
Fig. 46 Typical Structures, Rene 80, 400X

a. As-cast, unetched, eutectic-like MC

b. Heat treated, 92-5-3 etch, eutectic and fine γ'







c. As-coated, unetched, fine fingers, MC in diffusion zone

d. Coated (coarse grain spec.), rupture test 1093C, 127.2 hrs, carbide etch, coating transformation, few large σ needles

e. SM spec., coated, exposed, re-SM, coated, exposed 982C, 92-5-3 etch, coarse γ' & attacked eutectic γ'

Fig. 47 Typical Structures, Rene 120, 400X