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The effect of stress relief heat treatment and strain aging on A737 Grade C pressure vessel steel.

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THE EFFECT OF STRESS RELIEF HEAT TREATMENT AND STRAIN AGING ON A737 GRADE C PRESSURE VESSEL STEEL

by

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Abstract

An investigation of the effect of stress relief heat treatment and strain aging for A737 Grade C vanadium-nitrogen treated pressure vessel steel has been carried out.

The results obtained from the stress relief heat treatment study at 620°C (1150°F) for the normalized and the quenched and tempered plates indicated that a holding time of 10 hours or more gradually reduced the yield strength and tensile strength. The impact transition temperatures of the normalized plate increased almost linearly with the logarithm of the holding time; however, those of the quenched and tempered plate did not have a decrease in toughness due to the effect of stress relief heat treatment up to 300 hours of holding time.

The results obtained from prestraining, aging, and subsequent stress relief heat treatment of A737 Grade C normalized plate indicated that prestrain increased the yield strength, tensile strength, and impact transition temperatures almost linearly with increasing prestrain percentage. Subsequent aging, independent of prestrain percentage, also increased these properties. Stress relief heat treatment at 620°C (1150°F) or 580°C (1076°F) for 2 hours after strain aging decreased these properties but never restored them to their as-received levels. In stress relieving at 620°C (1150°F) for 100 hours after 10% strain aging, in spite of the extreme decrease in yield strength and tensile strength, the toughness was still almost the same as was obtained after strain aging.

Summarizing all of the results, the prestrain percentage and conditions of stress relief heat treatment are significant in controlling toughness loss in this steel and a low temperature stress relieving may be recommended for strain aged steel but its effects will not completely restore original properties.

Introduction

Background

Niobium- and/or vanadium containing high strength microalloy steels have been used in high pressure pipelines and in structures like bridges, buildings, and transport vehicles. They have not, however, been used frequently in pressure vessels in the United States. Cordea[1)] pointed out the principal reason for the lack of use of these steels in pressure vessels. He noted that the design stress for pressure vessels specified by the American Society for Mechanical Engineers (ASME) Boiler and Pressure Vessel Code is essentially based on a fraction of the tensile strength, and, therefore, increasing the yield strength by means of niobium and vanadium addition has little direct advantage.

In the last decade, promising high strength microalloy steels used in pressure vessel applications have been developed. Previous investigations have indicated that these steels, which were originally specified in ASTM specification A737 in 1976, may have high strength, good low temperature toughness, and good weldability with a low percentage of alloying elements.

The Pressure Vessel Research Committee (PVRC) of the Welding Research Council has continued extensive research on the basic characteristics of these new steels. Static and dynamic fracture toughness tests, Charpy impact and drop weight tests, and ambient and low temperature tension tests were conducted on A737 Grade C vanadium-nitrogen treated (VNT) plate developed in both the normalized, and the quenched and tempered conditions. These test results have already been reported[2]. The study of the fracture toughness of submerged arc weldments is also a part of another project at Lehigh University. The work on the effect of strain aging and stress relief heat treatment for this steel was undertaken in order to develop additional basic information as part of the overall program.

Strain Aging

Strain aging is induced in pressure vessel steels during cold forming and subsequent heating during fabrication, long term room temperature aging, or during high pressure proof testing and subsequent service at high temperatures. Generally, strain gives rise to an increase in yield strength, a process referred to as strain or work hardening, and a decrease in notch toughness. Moreover, aging after strain increases yield strength as well as ultimate tensile strength, and decreases notch toughness. Recently, as a result of the increased ability of forming equipment, thicker steel plates can be cold-formed. Consequently, the study of a steels' susceptibility to strain aging has become significant, especially from the standpoint of notch toughness.

It is generally accepted that strain aging is due primarily to the migration of carbon and nitrogen atoms to dislocations and locking them. According to Baird[3], this strain aging process is divided into two categories. One is that only solute atmosphere formation takes place around dislocations produced by straining. Another is that solute atmosphere formation is succeeded by precipitation on dislocation cores.

Effect of strain aging on tensile properties

The general effects of strain aging on the true stress-true strain curves of A737 Grade C normalized steel are illustrated in Fig. 1. Curves A and B show the relationship between true stress and true strain in the as received (normalized) condition and after 5% tensile strain plus 260°C (500°F) for 10 hours aging, respectively. By comparing curve A with curve B, the well known phenomena of strain aging, namely, the increase in yield strength due to strain-hardening and age-hardening, and the presence of the clear upper yield point after aging can be observed. An increase in ultimate tensile strength, and a decrease in elongation and in reduction of area may also take place.

Effect of composition

The effects of steel composition on strain aging are divided into two classes[3) 4)].

1. The first class consists of solutes which can lock dislocations and diffuse sufficiently quickly to produce strain aging. The effectiveness of an element in producing strain aging should be a function of three characteristics: its solubility, its diffusion coefficient, and its severity in locking dislocations.

Carbon and nitrogen have very similar diffusion coefficients in iron and produce identical distortions of the ferrite lattice; therefore, they are expected to produce very similar strain aging effects when present in solution in equal amounts. The main differences between the strain aging effects of carbon and nitrogen result from their widely differing solubilities in iron.

The solubility of nitrogen is high in temperatures above 200°C (392°F) where rapid precipitation can take place, but at 200°C (392°F) the solubility of carbon in equilibrium with cementite falls below 10^{-4}%. However, in aging above 100°C (212°F) there is evidence that fine carbide particles can redissolve to produce extensive strain aging.

The effect of nitrogen on strain aging is generally considered to be fairly directly related to the free nitrogen.

2. The second class consists of elements which themselves do not produce strain aging but which may reduce strain aging by altering the concentration or the mobility of the solute atoms which produce strain aging. These elements can be divided into four categories.

1) Elements in solid solution which interact slightly or not at all with nitrogen and carbon. These are copper, nickel, phosphorous, manganese, silicon, and arsenic.

2) Nitride formers, such as aluminum, boron, silicon, and manganese

3) Carbide formers, such as molybdenum

4) Nitride and carbide formers such as chromium, vanadium, niobium, titanium, and zirconium

In the course of heat treatment which does not produce the precipitation of the alloy carbides or nitrides, little effect can be obtained even when carbide and nitride formers are added.

It must be noted that A737 Grade C contains a low percentage of vanadium, chromium, aluminum, nickel, molybdenum, and silicon, and a relatively high percentage of manganese, which would reduce strain aging; however, A737 Grade C also contains much nitrogen, which increases the strength of the steel after normalizing, and consequently, directly produces strain aging.

Effect of aging temperature.

In the fabrication process and operation of pressure vessels, steels may be exposed to wide ranges of aging temperature or long term aging at elevated temperatures. In general, the effect of increasing the aging temperature, which ranges from 20°C (68°F) to 300°C (572°F)[3], is to accelerate the process toward the fully aged condition in which the properties do not vary greatly. In pressure vessel steels which contain a number of different alloying elements and different heat treatments, however, the effect of aging temperature on each steel may be different. This effect has been investigated by PVRC[5][6]. The results have shown that the maximum effect for most steels was observed at 260°C (500°F), although for some steels, the maximum effect was obtained at 370°C (700°F).

Effect of strain aging on toughness.

From the standpoint of pressure vessel applications, the effect of strain aging on toughness is more important than its effect on tensile properties because both strain and subsequent aging increase the Charpy impact transition temperature. From the previous investigations of PVRC[5][6] and Baird[3], it appears that the factors affecting toughness are similar to those affecting the increase in yield strength during aging. In some steels, however, the recovery in impact properties does not occur[7][8] even though the yield strength falls appreciably because of over-aging. In other steels, either a decrease in transition temperature is observed during aging[5][9] or no change in transition temperature is observed during strain and aging; only the decrease of shelf energy is observed[1) 10)]. These data make obvious the need for further investigation.

Stress relief heat treatment

Stress relief heat treatment is commonly conducted on heavy-wall pressure vessels after welding. It is defined as the uniform heating of a structure below the transformation temperature for a predetermined period of time, followed by uniform cooling. According to the ASME Boiler and Pressure Vessel Code, the minimum holding temperature and holding time at this temperature for carbon and low alloy steels are basically specified as 593°C (1100°F) and 1 hour per inch of thickness, respectively[11)].

Stress relief heat treatment has the following effects:

1) Reducing residual stress by creep process in which elastic strain is converted to plastic strain
2) Reducing localized high hardness and strength of welds and heat affected zones
3) Removing hydrogen introduced into the weld and heat affected zone
4) Removing aging effects
5) Improving ductility and fracture toughness of the weldment as a whole
6) Improving dimensional stability
7) Improving resistance to stress corrosion
8) Increasing strength by precipitation hardening

On the other hand, stress relief heat treatment may cause a loss in toughness, referred to as stress relief embrittlement, and a loss in strength for some steels. This embrittlement mechanism appears to be associated with carbide agglomeration on grain boundaries [6) 12)] or by precipitation hardening.

Effect of stress relief heat treatment after strain aging

During the fabrication process, stress relief heat treatment is performed after strain aging. The effect of stress relief heat treatment after strain aging in many steels has been investigated by PVRC[5)6)] and has been reviewed by Cordea[1)]. These results show that stress relief heat treatment after strain aging improves the toughness obtained after aging takes place; however, in some steels, this heat treatment can not attain the previous level of unstrained condition. The prestraining and precipitation hardening effect are partially retained. On the other hand, in many steels, this stress relief restores the notch toughness to its original value.

Possibly, these unanticipated results are obtained because stress relief heat treatment after strain aging has the multiple function of annealing, stress relief embrittling, tempering, and secondary hardening, depending on the temperature and holding time of the stress relief heat treatment, and on the chemical composition and heat treated history of the steel. Softening during annealing after cold forming, the most expected result, depends upon the cold forming ratio. In the case of low C cold formed steels with low cold forming ratios, higher temperatures and longer holding times

in annealing are required to make adequate softening possible[10].
From the standpoint of strain aging, Garofalo and Smith[8] have shown that adequate recovery after strain aging in normalized low carbon steels is not obtained even though the annealing temperature is 650°C (1202°F). In addition to this inadequate annealing effect, it is possible to obtain the effects of stress relief embrittling, tempering and secondary hardening at the higher stress relief temperature. Thus it is clear, when a new material, such as microalloyed steel is developed, additional information is needed to clarify these stress relief heat treatment effects.

Strengthening mechanisms and toughness of A737 Grade C

Because of the growing need for high yield strength, low impact transition temperature, good weldability, good formability, and minimum cost for structural steels, high strength microalloy steels like VNT steel have been developed since the 1960's[13)-17)].

Normalized condition

Normalizing is usually performed to increase yield strength and notch toughness in thick plate. During the normalizing process, a ferrite-pearlite microstructure is usually obtained. The factors which affect the strength of a steel with a ferrite-pearlite microstructure are[18]:

1. Ferrite grain size

Refinement of polygonal grain size increases the yield strength and decreases the impact transition temperature. It is well known that the yield strength and impact transition temperature vary linearly with the reciprocal of the square root of grain size.

Ferrite grain size depends on the original austenite grain size, transformation temperature, and second phase particles[13].

2. Pearlite content

Pearlite has virtually no effect on yield strength but causes an increase in tensile strength.

3. Solid solution strengthening

Solute elements increase both the yield strength and tensile strength. This result depends on the difference between the atomic size of the element and iron.

4. Dislocation strengthening

In general, decreasing the transformation temperature either by alloying and/or increasing the cooling rate refines the grain size and increases the dislocation density. This increased dislocation density increases the yield strength.

5. Precipitation hardening

Recent developments have combined precipitation hardening with grain refinement by means of vanadium, niobium, or titanium. The effectiveness of these elements depends on their solubility in austenite. Undissolved particles do not strengthen, but they give fine grain size. During cooling, dissolved vanadium, niobium, or titanium precipitates as carbonitrides and subsequently increases strength.

It is significant that only grain refinement decreases the impact transition temperature.

Next, the strengthening mechanism and toughness of A737 Grade C normalized plate will be considered. Vanadium nitride is much more soluble in austenite than aluminum nitride or niobium nitride[16); however, grain refinement by vanadium nitrides occurs at a low normalizing temperature of 900°C (1650°F). The mechanism of grain refinement by vanadium nitrides is considered to be similar to the mechanism of grain refinement by aluminum nitrides; particles precipitated by proper heat treatment pin the grain boundaries and retard grain growth[17). Moreover, precipitation hardening by finely dispersed vanadium carbonitrides occurs effectively during cooling after normalizing because vanadium carbide is virtually completely dissolved at 900°C (1650°F)[16). Vanadium nitride, therefore, appears to be much more effective than vanadium carbide with respect to the refinement of austenite grain size[17). The 0.015-0.02% nitrogen addition further increases the strength. Because it is difficult to maintain the free nitrogen content at a minimum and because vanadium does not appear to be as effective a grain refiner as niobium, vanadium-containing steels are not as tough as equivalent niobium-containing steels. Nevertheless, they can be manufactured with adequate toughness under the appropriate process conditions[17). For example, under low temperature normalizing with an aluminum additive, adequate precipitation hardening caused by the vanadium, and good grain refinement caused by the aluminum can be obtained while maintaining low free nitrogen content.

Manganese decreases the transformation temperature and provides for grain refinement in addition to solid solution hardening. Carbon provides the higher yield strength and tensile strength associated with the hardening effect of vanadium[17]. 0.15-0.20% carbon and 1.00-1.50% manganese-containing steels are generally used in thick plates.

A737 Grade C normalized plate appears to be manufactured mainly in accordance with the mechanism already mentioned. Moreover, a small percentage of molybdenum, nickel, and silicon may contribute to strengthening.

Quenched and tempered condition

The strengthening mechanism of quenched and tempered steel, even if a martensite or an acicular ferrite structure is present, is considered to be caused by the following factors[19) 20)].

1) Grain size

2) Solid solution hardening

3) Precipitation hardening

4) Dislocation density (internal twins only in martensite)

The significance of each factor varies with each steel. In the A737 Grade C quenched and tempered plate, the transformation from austenite to martensite during quenching from 900°C (1650°F) to water may be impossible in 76mm (3 inches) plate because of the relatively low hardenability and heavy thickness except in the regions adjacent to the plate surfaces. It appears, therefore, that quenching was employed to obtain a fine ferrite-bainite structure, which produces both high strength and good toughness. This structure may be partly precipitation hardening during cooling. During tempering, precipitation of vanadium carbides can occur and subsequently increase strength and decrease toughness. This effect, however, may be slight because of the low percentage of vanadium in this steel (0.08%). Consequently, tempering may improve toughness but may not decrease strength significantly.

Object of investigation

The objective of this investigation is to obtain basic information about stress relief heat treatment, strain aging and subsequent stress relief heat treatment in A737 Grade C steel. First, the effect of stress relief heat treatment at 620°C (1150°F) including extended holding times was examined for both normalized and quenched and tempered plates. Next, the effects of strain, strain aging, and subsequent stress relief heat treatment, were examined for normalized plate. Tension and Charpy impact tests were used for the examination of the effects. Test conditions for stress relief heat treatment and strain aging are shown in Table 1.

Testing procedure

Materials

The A737 Grade C VNT plates used in this investigation were provided by Armco Steel Corporation. Normalized and quenched and tempered plates were made from the same heat. Aadland used these plates for fracture toughness tests in a previous investigation[2). During normalizing, the hot-rolled plate was austenitized at 900°C (1650°F) for 30 minutes and cooled in air. The quenched and tempered plate was austenitized at 900°C (1650°F) for 40 minutes and water-quenched. Subsequently, this plate was tempered at 660°C (1220°F) for 30 minutes and cooled in air. The nominal thickness of both plates was 76mm (3 inches). The chemical compositions and mechanical properties as reported by Armco Steel Corporation are listed in Table 2.

Heat treating and straining

Stress relief heat treatment

Blocks approximately 25mm (1 inch) thick were saw-cut from one quarter or three quarter thickness position in the original plates. An electrical furnace with an air circulating fan was used for the heat treatment. The first block was placed in the furnace at a temperature of 620°C (1150°F). The second block was added 200 hours later. The third block was added 70 hours after the second. The fourth block was put into the furnace 20 hours after the third. Finally, the last block was added 8 hours after the fourth. The five blocks were held for additional 2 hours and cooled to 300°C (572°F). The cooling rates from

620°C (1150°F) to 300°C (572°F) and from 500°C (932°F) to 300°C (572°F) for normalized plate were 79°C/hr (142°F/hr) and 61°C/hr (110°F/hr) respectively. The rates for the quenched and tempered plate were 72°C/hr (130°F/hr) and 55°C/hr (99°F/hr). After the heat treatment, two transverse tensile test specimens and transverse Charpy specimens were taken from each block.

Straining.

Large tensile strip specimens for the Charpy specimens were prepared transversely to the plate rolling direction. The sampling method, specimen size, and orientation are illustrated in Fig. 2. Nominal 2%, 5%, and 10% tensile prestrains were induced in large strip specimens at room temperature using a Tinius Olson 534 kN (120,000 lb) testing machine with a constant crosshead speed of 1.27mm/min. (0.05 in./min.).

The strain distribution down the length of the nominally 2%, 5% and 10% prestrained specimens are shown in Fig. 3, 4, and 5. Examination of these figures shows that the nominally 5% prestrained specimens were almost exactly 5% except for one of the specimens in that sequence, which was also close to 5%. The nominal 10% prestrained specimens, which were consistently slightly lower than 10%, averaged about 9.5%, and the 2% prestrained specimens were a little higher than 2%, averaging about 2.5%. Moreover, the 10% prestrained specimens, as might be expected, were also less uniform than the lower prestrained specimens. It does not appear, however, that these results will have a significant effect on the experiment since the choice of specific strain level was somewhat arbitrary.

After prestraining, Charpy specimens were machined longitudinally to the strain, i.e., transverse to the plate rolling direction.

Modified button-head tensile specimens measuring 6.76mm (0.266 in.) in diameter were also prepared transverse to the plate rolling direction as shown in Fig.2. Nominal 2%, 5%, and 10% prestrains were induced at room temperature using a 44.5 kN (10,000 lb) instron testing machine with a gage length of 25.4mm (1 in.) extensometer at a constant crosshead speed of 0.25mm/min. (0.01 in./min.). Actual prestrains for nominal 2%, 5%, and 10% prestrains were 2.0 to 2.2%, 4.9 to 5.1%, and 9.7 to 10.0% respectively.

After prestraining, standard button-head tensile specimens were machined by slightly reducing the diameter of the parallel portion. The diameters of some specimens after 10% nominal prestraining, however, became smaller than the nominal standard diameter. These diameters of the tested specimens did not deviate from the nominal standard diameter by more than 0.15mm (0.006 in.). These slight deviations do not seem to be significant.

Aging

The aging of the specimen was performed in the same forced air furnace used for the stress relief heat treatment studies. Standard aging conditions, an aging temperature of 260°C (500°F) and a holding time of 10 hours, were chosen in order to obtain the fully aged condition. Moreover, another aging temperature of 370°C (700°F) was chosen for the 5% nominally prestrained condition in order to examine the effect of the aging temperature. Charpy specimens and tensile specimens were prepared in the same way they were machined after prestraining.

Stress relief heat treatment after strain aging

Stress relief heat treatment after strain aging was conducted in the same manner as stress relief heat treatment of the original plate. Although a temperature of 620°C (1150°F) for 2 hours was chosen as a standard stress relief condition, additional stress relief conditions of 580°C (1076°F) for 2 hours and 620°C (1150°F) for 100 hours were used for the 10% strain-aged specimens. The cooling rates for each stress relief heat treatment are listed in Table 3. Charpy specimens and tensile specimens were machined in the same manner they were machined after prestraining.

Mechanical testing

Tensile tests

The tensile specimens were of the standard button-head type which had 6.35mm (0.25 in.) diameter and 25.4mm (1.000 in.) gage length; however, some 10% prestrained specimens had slightly smaller diameters. The orientation of all tensile specimens was transverse to the plate rolling direction. Testing was carried out using a 44.8 kN (10,000 lb) instron universal tester with a constant crosshead speed of 1.27mm/min. (0.05 in./min.). An extensometer was used to determine the 0.2% offset yield strength. Other procedures were conducted according to the ASTM standard A370 and E8. Ultimate tensile strength, 0.2% offset yield strength, percent elongation, and reduction of area were determined.

Charpy impact tests

ASTM standard type A Charpy specimens which are full size with a 2mm V-notch were machined transversely to the rolling direction. A Satec system SI-1D impact tester with a maximum impact energy of 325J (240 ft-lb) was used. Testing was performed on each material using a wide range of temperatures in order to obtain full transition curves. 2-methylbutane cooled by liquid nitrogen was used for test temperatures below -60°C (-76°F). The testing procedure was conducted according to ASTM standards A370 and E23. Impact energies were obtained directly from the impact tester, and lateral expansions and shear fracture appearances were measured.

The 20.3J (15 ft-lb), the 67.8J (50 ft-lb) energy transition temperatures, the 0.89mm (35 mil) lateral expansion transition temperature, and the 50% shear fracture appearance transition temperature were chosen as criteria from the transition curves.

Optical metallographic examination

Conventional optical metallographic examinations were conducted on broken Charpy impact specimens. The plane examined was parallel to the plate rolling direction and perpendicular to the plate rolling surfaces. The samples were mounted in bakelite for polishing. Silicon carbide papers were used to polish the samples. Next, the surfaces were polished with alumina polishing powder. Polishing was completed with 0.06 micron alumina polishing powder. The samples were etched in 2% nital for about 15 seconds. A Zeiss Axiomat metallograph was used to carry out the microscopic examinations.

Electron metallographic examination with TEM and SEM

The electron metallographic examinations were performed using a Philips EM400 transmission electron microscope. The thin foil specimens used in this examination were prepared from the broken Charpy impact specimens. First, thin plates measuring 0.38mm (0.015 in.) in thickness were sliced with a diamond blade and then punched into discs with 3mm (0.12 in.) diameters. These discs were polished to a thickness of about 0.07mm (0.003 in.) using silicon carbide papers. Finally, they were jet-polished with a 2% perchloric acid-methanol solution cooled to $-60^{\circ}C$ to $-70^{\circ}C$ ($-76^{\circ}F$ to $-94^{\circ}F$) at 110V.

In addition to the thin foil transmission electron metallographic examination, an ETEC scanning electron metallographic examination was carried out on the as-received and 620°C (1150°F) for 300 hours stress relieved materials for the observation of the general morphology of the carbides at the grain boundaries. The samples were prepared using broken Charpy impact specimens. These samples were prepared in the same manner they were prepared for optical metallographic examinations and then these samples were Au-Pd coated.

Test results

Effect of stress relief heat treatment

Tensile test results

The tensile test results for the normalized, and the quenched and tempered plates are shown in Table 4, Fig. 6, and Fig. 7. As already noted by Aadland[2], the normalized plate did not satisfy the minimum yield requirement for A737 Grade C, 415 MPa (60 ksi), in as-received condition: however, the quenched and tempered plate satisfied that requirement in as-received condition. Change in the yield strength or the tensile strength of neither the normalized, nor the quenched and tempered plates was observed after the stress relief heat treatment at 620°C (1150°F) for 2 hours. Holding times of 10 hours or more, however, gradually reduced the yield strength and the tensile strength for both the normalized, and the quenched and tempered plates. Apparently, no appreciable secondary hardening was obtained in either normalized or quenched and tempered plate. The yield strength of the normalized plate decreased less than that of the quenched and tempered plate. After heat treatments lasting 10 hours or more, the tensile strength of the normalized plate became lower than the minimum tensile requirement for A737 Grade C (550 MPa (80 ksi)). After a holding time of 100 hours or more, the yield strength and tensile strength of the plate quenched and tempered at 660°C (1220°F) were lower than the strength required by A737 Grade C. The tensile test results for A737 Grade B niobium treated low sulphur normalized plate are shown in Fig. 6[12]. The rate of decrease in the yield strength and tensile strength is similar to A737 Grade C normalized plate; however, the degree of this change in A737 Grade C is less than in A737 Grade B. Although the reduction of area and elongation showed some experimental deviations, they gradually increased with the increasing holding time.

Charpy impact test results

The Charpy impact test results for the normalized plate are shown in Table 5 and Fig. 8. The impact energy and lateral expansion transition curves are illustrated in Fig. 10 and 11. Each transition temperature chosen in this investigation increased almost linearly with the logarithm of the holding time. For example, Tr20.3J (Tr15 ft-lb) increased from -95°C (-139°F) in the as-received condition to -46°C (-71°F) after 300 hours of stress relief heat treatment. It appears that the shelf energy also increased as the holding time increased. It should be noted, however, that the temperature at which 100% of the shear fracture appearance was obtained also increased with the increased holding time.

The impact test results for A737 Grade B niobium treated normalized plate are similarly plotted in Fig. 8[12)]. It appears that the transition temperatures of A737 Grade B increased sharply when the holding time was 100 hours or longer.

The results for the quenched and tempered plate are shown in Table 5 and Fig. 9. The impact energy and lateral expansion transition curves are illustrated in Figs. 12 and 13. The transition temperatures of the plate quenched and tempered at 660°C (1220°F), which was higher than the stress relieving temperature by 40°C (70°F), did not seem to suffer the effect of stress relief heat treatment at 620°C (1150°F) lasting for 300 hours. Moreover, the transition temperatures, except for the Tr20.3J (Tr15 ft-lb), of the quenched and tempered plate were slightly lower than those of the normalized plate before stress relief heat treatment. As a result, it was obvious that the toughness of the quenched and tempered plate increased relative to the normalized plate as stress relief holding time increased.

Optical metallographic examination results

The optical microstructures examined are shown in micrographs 1 and 2. Micrographs 1a and 1b show a typical normalized microstructure with strong ferrite-pearlite bands. The ASTM grain size number is 10. Micrographs 1c and 1d show the microstructure of the normalized plate after 30 hours and 300 hours of stress relief heat treatment respectively. The spheroidization of the pearlite is observed in these photographs, especially in photograph 4. Micrographs 2a and 2b show the fine ferrite and probably tempered bainite microstructure of the quenched and tempered plate with an ASTM grain size of 12.5. Spheroidization of carbides is also observed after stress relief heat treatment in micrographs 2c and 2d.

Electron metallographic examination results

Results with TEM

The observed results are shown in micrographs 3 and 4. The fine round-shape precipitates with a random distribution were observed in the as-received material as shown in micrograph 3a. These precipitates could not be identified from their diffraction pattern; however, the STEM analysis showed that they were all vanadium precipitates, probably vanadium carbonitrides. Moreover, few dislocations, which are sometimes induced during the thin foil specimen preparation process, were observed. Mircograph 3b and 3c show the general morphology and decomposed pearlite after 620°C (1150°F) X 300 hours of stress relief heat treatment respectively. The coarse carbides at the grain boundaries are not clearly observed with this thin foil technique as pointed out by Shinohe[12).

Micrograph 4a also shows fine vanadium precipitates in the as-received material of the quenched and tempered plate. In this material, in addition to fine vanadium precipitates, some larger compound precipitates were observed. Micrograph 4b and 4c show the difference in general morphology of carbides in the as-recieved , and in the 620°C (1150°F) X 300 hours of stress releived material. In the stress relieved material, the carbieds at grain boundaries were a little coarser than those in the as-received material. Some dislocations were observed in these specimens; however, these were induced during the thin foil specimen preparation process.

Results with SEM.

Micrographs 5 and 6 show the general morphology of the carbides in the as-received condition and after 620°C (1150°F) for 300 hours for the normalized, and the quenched and tempered plates.

In the as-received normalized material, a ferrite-pearlite banded structure and fine carbides in a ferrite matrix are observed. Micrographs 5c and 5d show spheroidal cementite in the prior pearlite regions and coarse carbides at grain boundaries in the stress relieved material.

In the quenched and tempered plate, spheroidal carbides, most of which are at grain boundaries, (with some, however, inside the grain) are observed in both the as-received and stress relieved materials. It is obvious from the micrographs 6a and 6b that the carbides in the stress relieved material are coarser than those in the as-received material.

Effect of strain aging

Tensile test results

The tensile test results for the various conditions of prestrain are shown in Table 6, Fig. 14, and Fig. 15. Fig. 16 shows the effect of the prestraining, aging, and subsequent stress relief heat treatment on the yield strength and tensile strength. Prestrain increased the yield strength and tensile strength almost linearly with the increasing prestrain percentage. The increase in yield strength, about 22 MPa (3.3 ksi)/1% strain, was more extreme than the increase in tensile strength, about

8.2 MPa (1.2 ksi)/1% strain. It is natural that the elongation obtained after prestraining decreased with increasing prestrain; however, an appreciable change was not noted in reduction of area.

Subsequent aging, independent of the prestrain percentages, increased the yield strength about 70 MPa (10 ksi) as shown in Fig. 16. The increase in tensile strength was expected to show a trend similar to the increase in yield strength; however, this trend was not consistant. The elongation and reduction of area decreased as the prestrain increased. During the 2%, 5% and 10% strain aging process, the total increases were 116 MPa (16.9 ksi), 201 MPa (29.2 ksi), and 275 MPa (39.9 ksi) in yield strentch, respectively, and 50 MPa (7.2 ksi), 87 MPa (12.5 ksi), and 122 MPa (17.7 ksi) in the tensile strength, respectively.

A comparison of the results obtained after 260°C (500°F) aging and 370°C (700°F) aging show that the yield strength and tensile strength obtained after 370°C (700°F) aging were slightly higher than those obtained after 260°C (500°F) aging. But, the comparison of the Charpy impact results obtained after 260°C (500°F) aging and 370°C (700°F) aging showed little difference between them as shown in Table 7, Fig. 22, and Fig. 23. Therefore, the aging condition of 260°C for 10 hours probably produces a fully aged condition in this steel.

Subsequent stress relief heat treatment at 620°C (1150°F) for 2 hours decreased the yield strength and tensile strength to levels lower than those of the prestrained but still higher than those of the as-received. A typical true stress-true strain curve after 5% strain aged and subsequently 620°C (1150°F) for 2 hours stress relieved material is illustrated in Fig. 1. The results obtained after different stress relief heat treatments for 10% strain aged materials are shown in Fig. 26. After 620°C (1150°F) for 100 hours of stress relief heat treatment, the level of the tensile strength decreased approximately to the level of the as-received; however, the yield strength was still higher than the as-received. The yield strength and tensile strength obtained after stress relief heat treatment at 580°C (1076°F) for 2 hours almost equaled the yield strength and tensile strength obtained at 620°C (1150°F) for 2 hours.

<u>Charpy impact test results</u>

The Charpy impact test results for the various conditions of prestrain are shown in Table 7, and Fig. 17, 18, and 19. The impact energy transition curves and lateral expansion transition curves are illustrated in Fig. 20 to 25. These results, especially those illustrated in Fig. 19, show that the changes in transition temperature increased linearly as the prestrain increased. (The increase rate of Tr67.8J (Tr50 ft-lb) was about 3.4°C (6°F)/1% strain.) Moreover, the transition temperatures increased independently of the prestrain percentages following aging (for example, the increase in Tr67.8J was approximately 20°C (45°F) and then decreased with subsequent stress relief heat treatment. Nevertheless, they were still a little higher than those obtained after prestraining. The maximum upward shift of Tr67.8J (Tr50 ft-lb) which occurred in the 2%, 5%, and 10% strain aging were 22°C (40°F), 35°C (63°F), and 56°C (101°F) respectively.

Almost no difference in impact properties after 260°C (500°F) aging and 370°C (700°F) aging was noted (Table 7, Fig. 22, and Fig. 23).

A difference in the relationship between the changes in strength and the changes in transition temperature was noted after the stress relief heat treatment at 620°C (1150°F) for 100 hours as shown in Fig. 26. Although the level of tensile strength decreased to the level of the as-received, the level of transition temperatures still almost equaled the levels of the strain aged material. This relationship between the change in strength and the change in impact transition temperatures is considered to be the same as the relationship after the stress relief heat treatment for the as-received plate. These results show that the stress relief heat treatment conditions affect the results of stress relief heat treatment after strain aging. The results also indicate that a higher temperature and/or longer holding time of stress relief heat treatment may initially decrease the impact transition temperature increased after strain aging but ultimately it may increase the impact transition temperature.

Optical metallographic examination results

The optical microstructures examined after 10% prestraining, 260°C (500°F) X 10 hours of strain aging, and subsequent stress relief heat treatment at 620°C (1150°F) for 2 hours and 100 hours are shown in micrographs 7a, 7b, 7c, and 7d respectively. No significant change was observed in these microstructures except for a little spheroidization of pearlite in stress relieved material at 620°C (1150°F) for 100 hours.

Electron metallographic examination results with TEM

Special attention was given to the preparation of the thin foil specimens because dislocations are induced during specimen preparation as shown in micrograph 4b and 4c.

Many dislocation tangles, shown in micrograph 8a, were observed in the 10% strained material. These tangles were distributed at random, not uniformly. It was obvious that these dislocations were induced during the prestraining process. The white round points observed in micrograph 8a are holes where precipitates have fallen out. The strengthening mechanisms here are undoubtedly dislocation tangles and the precipitates which pin the dislocations. The same features were found in 10% strain aged material, as shown in micrographs 8b and 8c, however, the difference between these materials, namely aging effects, is not clear. Some dislocation tangles were still observed in the material stress relieved at 620°C (1150°F) X 2 hours as shown in micrograph 9.

Discussion

Effect of stress relief heat treatment

During stress relief heat treatment in A737 Grade C VNT steel, the yield strength and tensile strength decreased in both the normalized and quenched and tempered plate. Pronounced secondary hardening did not occur. During stress relief heat treatment at a high temperature for an extended holding time, the pearlite or carbide agglomerate spheroidizes and then dissolves gradually in the ferrite matrix. Subsequently, the dissolved carbon precipitates both at grain boundaries and in the matrix as in high temperature tempering of C-Mn steels. In vanadium treated steels, this process can be delayed[21)]. After stress relief heat treatment at 620°C (1150°F) for 300 hours, however, the reprecipitated carbides or coarsened carbides were observed at grain boundaries as shown in micrographs 3c, 4c, 5c, 5d, and 6b.

Probably, spheroidization of the pearlite and reduction in the amount of pearlite caused the decrease in the yield strength and tensile strength, while the growth of the repricipitated carbides caused the increase in the impact transition temperatures. This embrittlement mechanism appears to be the same mechanism observed in A737 Grade B niobium treated steels[12)] and C-Mn steels[22)]. In the quenched and tempered plate during stress relief heat treatment, the general effects of tempering and recovery of a less tangled dislocation structure, might compensate for increase in the impact transition temperatures. As a result, a change in these properties might not have occurred.

Effect of strain aging

The as-received condition including the heat treatment history is significant in the investigation of strain aging. The material used in this investigation, A737 Grade C VNT normalized plate, has few dislocations and little internal stress in the as-received material and mechanical property tests on material given stress relief heat treatment at 620°C (1150°F) for 2 hours (which showed change in neither the yield strength nor tensile strength and little decrease in the impact transition temperatures) support this contention.

During straining, the interaction among dislocations induced by prestraining and between dislocations and precipitates resulted in an increase in the yield strength which was almost linear to the increase in strain, classical strain hardening. After prestraining, residual stress and strain energy remained in the steel. Strain hardening resulted in an increase in the impact transition temperatures which also increased linearly with the strain.

During aging, a tempering effect and recovery effect in addition to the aging effect itself may be expected depending on the temperature and the holding time of the aging. The aging at 260°C (500°F) or 370°C (700°F) for 10 hours performed for A737 Grade C normalized plate seemed to have a classical aging effect which resulted in the increase in the yield strength, tensile strength, and impact transition temperatures independent of the prestrain percentage in the range of 2% to 10% prestrain.

The susceptibility to strain aging of this steel, which contains much nitrogen, is not high considering the results obtained by previous researchers[1) 5) 6)]. The generally accepted reason for this is that during normalizing at a temperature of 900°C (1650°F), nitrogen precipitates as vanadium nitride or aluminum nitride; consequently, the amount of soluble nitrogen itself is not high but sufficient to cause aging. The precentage of nitrogen which precipitates as VN or AlN at 900°C (1650°F) can be calculated roughly using the solubility relationships of VN and AlN: $\log(V)(N) = -8330/T + 3.40 + 0.12(\%Mn)$[16)] and $\log(Al)(N) = -7400/T + 1.95$[23)]. The calculated result shows that 0.0130% of the 0.0158% total nitrogen content would precipitate at 900°C (1650°F).

Stress relief heat treatment has the various effects mentioned previously. The results of stress relief heat treatment in the as-received material show neither a secondary hardening effect nor a tempering effect. The results, however, show a softening effect and an embrittling effect after the extended holding time. Therefore, after stress relief heat treatment at 620°C (1150°F) or 580°C (1076°F) for 2 hours after strain aging, the effect of removing the aging effects was mainly complete. In addition to this effect, a slight softening effect (a partial but incomplete annealing effect because of the relatively low prestrain percentages and the addition of vanadium) was obtained. As was shown, the decreases in yield strength and tensile strength after stress relief heat treatment were greater than the increases after the aging. In spite of the decrease in the yield strength and tensile strength, the transition temperatures were still almost the same as those obtained after stress relief heat treatment at 620°C (1150°F) for 100 hours following 10% strain aging. This seemed to show that the stress relief embrittling, softening caused by the incomplete annealing effects and spheroidization of carbides (as observed after stress relief heat treatment in the as-received material) occurred after aging.

The results obtained in the series of prestraining, aging, and subsequent stress relief heat treatment experiments indicate that the strain percentage induced by cold forming and the conditions of stress relief heat treatment are significant for A737 Grade C normalized steel because although stress relief heat treatment can remove the aging effects, it barely removes the prestraining effects. Moreover, stress relief heat treatment results in softening and embrittlement under extended holding times and/or high temperatures. The results of the stress relief heat treatments after 10% strain show that a lower temperature stress relieving may be recommended for A737 Grade C normalized steel.

Conclusion

From the results of the experiments on the effect of stress relief heat treatment on A737 Grade C in the normalized and the quenched and tempered condition, the following conclusions can be drawn.

1. Changes in the yield strength and tensile strength of either the normalized, or the quenched and tempered plate were not observed after stress relief heat treatment at 620°C (1150°F) for 2 hours.
2. The stress relieving holding times of 10 hours or more at 620°C (1150°F) gradually reduced the yield strength and tensile strength in both the normalized and the quenched and tempered plates.
3. The transition temperatures of the normalized plate increased almost linearly with the logarithm of the stress relieving holding time.
4. This embrittlement appeared to be due to carbide coarsening at grain boundaries.
5. The impact transition temperatures of the quenched and tempered plate did not show the effect of stress relief heat treatment at 620°C (1150°F) for holding times of 300 hours.

From the results of the experiments on the effects of strain aging on A737 Grade C normalized plate, the following conclusions can be drawn.

1. Prestrain increased the yield strength, tensile strength and

impact transition temperatures almost linearly with the increasing prestrain percentage. The increase rates for these properties were:

YS = 22 MPa (3.3 ksi)/1% strain
TS = 8.2 MPa (1.2 ksi)/1% strain
Tr67.8J (Tr50 ft-lb) = 3.4°C (4°F)/1% strain

2. Subsequent aging, independent of the prestrain percentage, increased the yield strength about 70 MPa (10 ksi). The impact transition temperatures also increased: for example, Tr67.8J (Tr50 ft-lb) increased about 20°C (45°F).
3. Stress relieving at 620°C (1150°F) or 580°C (1076°F) for 2 hours after strain aging decreased the yield strength and tensile strength to levels lower than those of the prestrained material but remained still higher than those of the as-received material. It also decreased the impact transition temperatures; however, these were still a little higher than those obtained after prestraining.
4. In stress relief heat treatment at 620°C (1150°F) for 100 hours after 10% strain aging, in spite of a substantial decrease in yield strength and tensile strength, the impact transition temperatures were still almost the same as those obtained after strain aging because of stress relief embrittlement.
5. The results obtained in the series of prestraining, aging, and subsequent stress relief heat treatment experiments indicate that the prestrain percentage and conditions of stress relief heat treatment are significant and a lower temperature of stress relieving may be recommended for this steel.

Table 1 Test condition

1. Stress relief heat treatment

   Temperature: 620°C (1150°F)
   Holding time: 2, 10, 30, 100, 300 hours

2. Straining, aging and stress relief heat treatment

Strain level	Aging	Stress relief
None	None	None
None	260°C X 10 hours	None
None	260°C X 10 hours	620°C X 2 hours
2%	None	None
2%	260°C X 10 hours	None
2%	260°C X 10 hours	620°C X 2 hours
5%	None	None
5%	260°C X 10 hours	None
5%	370°C X 10 hours	None
5%	260°C X 10 hours	620°C X 2 hours
5%	370°C X 10 hours	620°C X 2 hours
10%	None	None
10%	260°C X 10 hours	None
10%	260°C X 10 hours	620°C X 2 hours
10%	260°C X 10 hours	620°C X 100 hours
10%	260°C X 10 hours	580°C X 2 hours

Table 2 Chemical composition and mechanical properties of A737 Grade C (Data supplied by Armco Steel Corp. except where noted)

1. Chemical composition (wt.%)

Analysis	C	Mn	P	S	Si	V	N	Ni	Cr	Mo	Al
Ladle (Heat 42034)	.20	1.29	.010	.007	.30	.084	.014	*	*	*	*
Check (Bethlehem Steel)	.18	1.29	.005	.007	.30	.09	.0158	.15	.18	.08	.009

* : Not reported

2. Mechanical properties

2.1. Tensile properties (Room temperature, orientation not specified)

Heat treatment	Yield strength MPa	(ksi)	Tensile strength MPa	(ksi)	Elongation (GL=2") %
Normalized	422	(61.2)	581	(84.2)	26.0
Quenched and Tempered	560 524	(81.2) (76.0)	618 623	(89.7) (90.4)	25.0 29.5

2.2. Charpy impact properties (0°C (32°F))

Heat treatment	Impact energy			
	Transverse J	(ft-lb)	Longitudinal J	(ft-lb)
Normalized	121	(91)	178	(131)
Quenched and Tempered	Data not reported			

Table 3 Cooling rates of stress relief heat treatment after strain aging

Strain	Stress relief	Cooling rate 620°C - 300°C (1150°F - 570°F) °C/hr	(°F/hr)	Cooling rate 500°C - 300°C (932°F - 570°F) °C/hr	(°F/hr)
0%	620°C X 2 hrs.	104	(187)	76	(137)
2%	620°C X 2 hrs.	104	(187)	76	(137)
5%	620°C X 2 hrs.*	29	(52)	22	(40)
10%	620°C X 2 hrs.	105	(189)	77	(139)
10%	620°C X 100 hrs	79	(142)	59	(106)
10%	580°C X 2 hrs.	114**	(205)**	91	(164)

Laboratory furnace: Lindberg

*: Heavy duty

**: Cooling rate from 580°C t0 300°C (1076°F to 572°F)

Table 4 Tensile test results after stress relief heat treatment

Material	Stress relief	Y.S. MPa	Y.S. (ksi)	T.S. MPa	T.S. (ksi)	El %	R.A. %
Normalized	As received	399 396	(57.8) (57.4)	562 564	(81.5) (81.8)	30.0 29.5	65.4 62.3
	620°C X 2 hrs.	394 392	(57.1) (56.9)	558 563	(80.9) (81.6)	29.8 32.5	65.7 65.9
	620°C X 10 hrs.	392 390	(56.9) (56.5)	549 542	(79.6) (78.6)	33.0 30.9	68.1 68.4
	620°C X 30 hrs.	377 368	(54.7) (53.3)	530 524	(76.9) (76.0)	33.5 31.2	67.7 69.0
	620°C X 100 hrs.	368 374	(53.3) (54.3)	508 509	(73.7) (73.8)	35.0 35.5	71.5 71.9
	620°C X 300 hrs.	354 359	(51.3) (52.1)	493 492	(71.5) (71.4)	32.8 33.4	71.7 70.8
Quenched and Tempered	As received	473 467	(68.6) (67.7)	601 583	(87.2) (84.5)	29.0 24.7	68.4 69.5
	620°C X 2 hrs.	468 475	(67.9) (68.9)	590 598	(85.6) (86.8)	28.9 30.9	69.5 68.4
	620°C X 10 hrs.	445 479	(64.5) (69.5)	566 600	(82.1) (87.0)	26.6 32.9	67.5 66.8
	620°C X 30 hrs.	450 445	(65.2) (64.6)	578 572	(83.8) (82.9)	31.7 29.7	69.7 70.2
	620°C X 100 hrs.	427 412	(62.0) (59.8)	550 542	(79.7) (78.6)	30.9 27.8	70.3 73.8
	620°C X 300 hrs.	411 392	(59.6) (56.7)	527 520	(76.4) (75.5)	34.3 31.5	71.7 73.0

Table 5 Charpy impact test results after stress relief heat treatment

Material	Stress Relief	Tr20.3J (Tr15ft-lb) °C (°F)	Tr67.8J (Tr50ft-lb) °C (°F)	Tr0.89mm (Tr35mil) °C (°F)	50%SA FATT °C (°F)	Shelf energy J(ft-lb)
Normalized	As received	-95 (-139)	-20 (-4)	-38 (-36)	-26 (-15)	106 (78)
	620°C X 2 hrs.	-77 (-107)	-11 (12)	-35 (-31)	-13 (9)	105 (77)
	620°C X 10 hrs.	-68 (-90)	-13 (9)	-33 (-27)	-10 (14)	117 (86)
	620°C X 30 hrs.	-59 (-74)	0 (32)	-26 (-15)	2 (36)	110 (81)
	620°C X 100 hrs.	-57 (-71)	0 (32)	-25 (-13)	3 (37)	136 (100)
	620°C X 30 hrs.	-46 (-51)	4 (39)	-23 (-9)	8 (46)	133 (98)
Quenched and Tempered	As received	-79 (-110)	-23 (-9)	-43 (-45)	-38 (-36)	120 (88)
	620°C X 2 hrs.	-77 (-107)	-25 (-13)	-41 (-42)	-27 (-17)	117 (86)
	620°C X 10 hrs.	-75 (-103)	-22 (-8)	-35 (-31)	-22 (-8)	118 (87)
	620°C X 30 hrs.	-78 (-108)	-28 (-18)	-42 (-44)	-23 (-9)	116 (85)
	620°C X 100 hrs.	-70 (-94)	-23 (-9)	-37 (-35)	-22 (-8)	139 (102)
	620°C X 300 hrs.	-70 (-94)	-32 (-26)	-43 (-45)	-27 (-17)	139 (102)

Table 6 Tensile test results on strain aging (A737 Gr. C normalized plate)

Condition	YS MPa	YS (ksi)	TS MPa	TS (ksi)	El %	RA %
As received	399 396	(57.8) (57.4)	562 564	(81.5) (81.8)	30.0 29.5	65.4 62.3
0% strain + aging	400 410	(58.0) (59.5)	580 580	(84.1) (84.1)	32.5 33.0	68.6 65.9
0% strain + aging + SR	403 374	(58.4) (54.3)	551 537	(79.9) (77.9)	34.8 30.8	66.7 67.6
2% strain	457 432	(66.3) (62.6)	610 578	(88.5) (83.9)	31.0 33.5	61.8 67.3
2% strain + aging	507 521	(73.5) (75.5)	607 618	(88.1) (89.6)	30.8 31.4	65.5 65.9
2% strain + aging + SR	428 402 435	(62.1) (58.3) (63.1)	590 559 592	(85.6) (81.1) (85.9)	33.0 33.3 32.5	61.1 65.6 65.4
5% strain	524 514	(76.0) (74.5)	590 578	(85.6) (83.8)	26.6 26.5	64.1 64.6
5% strain + aging	614 583	(89.0) (84.6)	653 625	(94.7) (90.7)	19.7 22.7	64.4 61.4
5% strain + aging (370°C)	613 566	(88.9) (82.1)	670 629	(97.2) (91.2)	23.5 25.9	63.1 64.0
5% strain + aging + SR	436 421	(63.2) (61.1)	586 576	(85.0) (83.5)	26.9 25.3	67.4 63.0
5% strain + aging + SR (370°C)	467 445	(67.7) (64.6)	616 603	(89.3) (87.5)	27.7 27.0	67.4 66.4

Aging: 260°C X 10 hours, unless otherwise noted
SR: 620°C X 2 hours, unless otherwise noted

Table 6 continued

Condition	YS MPa (ksi)	TS MPa (ksi)	El %	RA %
10% strain	605 (87.7)	623 (90.4)	22.6	61.6
	638 (92.5)	667 (96.7)	23.7	64.3
10% strain + aging	703 (101.9)	709 (102.9)	8.7*	61.9
	650 (94.2)	667 (96.8)	7.5*	61.9
	682 (98.9)	692 (100.4)	19.9	61.6
	656 (95.2)	671 (97.3)	20.2	59.6
10% strain + aging + SR	501 (72.6)	623 (90.4)	27.0	65.2
	501 (72.6)	631 (91.5)	26.8	64.7
10% strain + aging + SR (100hrs)	449 (65.1)	567 (82.3)	26.7	67.2
	445 (64.5)	560 (81.3)	12.1*	68.8
10% strain + aging + SR (580°C)	503 (73.0)	618 (89.7)	12.3*	61.0

Aging: 260°C X 10 hours, unless otherwise noted

SR: 620°C X 2 hours, unless otherwise noted

* : Fractured outside of the gage length

Table 7 Charpy impact test results on strain aging (A737 Gr. C normalized plate)

Condition	Tr20.3J (Tr15ft-lb) °C (°F)	Tr67.8J (Tr50ft-lb) °C (°F)	Tr0.89mm (Tr35mil) °C (°F)	50%SA FATT °C (°F)	Shelf energy J (ft-lb)
As received	-95 (-139)	-20 (-4)	-38 (-36)	-26 (-15)	106 (78)
0% strain + Aging	-80 (-112)	-18 (0)	-34 (-29)	-22 (- 8)	107 (79)
0% strain + Aging + SR	-67 (- 89)	-13 (9)	-32 (-26)	-28 (-18)	101 (74)
2% strain	-71 (- 96)	-15 (5)	-29 (-20)	-20 (- 4)	90 (66)
2% strain + Aging	-70 (- 94)	2 (36)	-16 (3)	-7 (19)	97 (71)
2% strain + Aging + SR	-59 (- 74)	-7 (19)	-22 (-8)	-20 (- 4)	98 (72)
5% strain	-72 (- 98)	-4 (25)	-18 (0)	-13 (9)	99 (73)
5% strain + Aging	-45 (- 49)	15 (59)	-6 (21)	-3 (27)	75 (55)
5% strain + Aging(370°C)	-57 (- 71)	20 (68)	5 (41)	2 (36)	91 (67)
5% strain + Aging + SR	-60 (- 76)	2 (36)	-8 (18)	0 (32)	102 (75)
5% strain + Aging(370°C)+ SR	-54 (- 65)	2 (36)	-14 (7)	-11 (12)	97 (71)
10% strain	-38 (- 36)	14 (57)	-2 (28)	-6 (21)	76 (56)
10% strain + Aging	-21 (- 6)	36 (97)	12 (54)	8 (46)	82 (60)
10% strain + Aging + SR	-51 (- 60)	20 (68)	2 (36)	6 (43)	95 (70)
10% strain + Aging + SR (100hrs.)	-17 (1)	30 (86)	8 (46)	9 (48)	82 (60)
10% strain + Aging + SR (580°C)	-42 (- 44)	22 (72)	-4 (25)	-6 (21)	80 (59)

Aging: 260°C X 10 hours, unless otherwise noted
SR: 620°C X 2 hours, unless otherwise noted

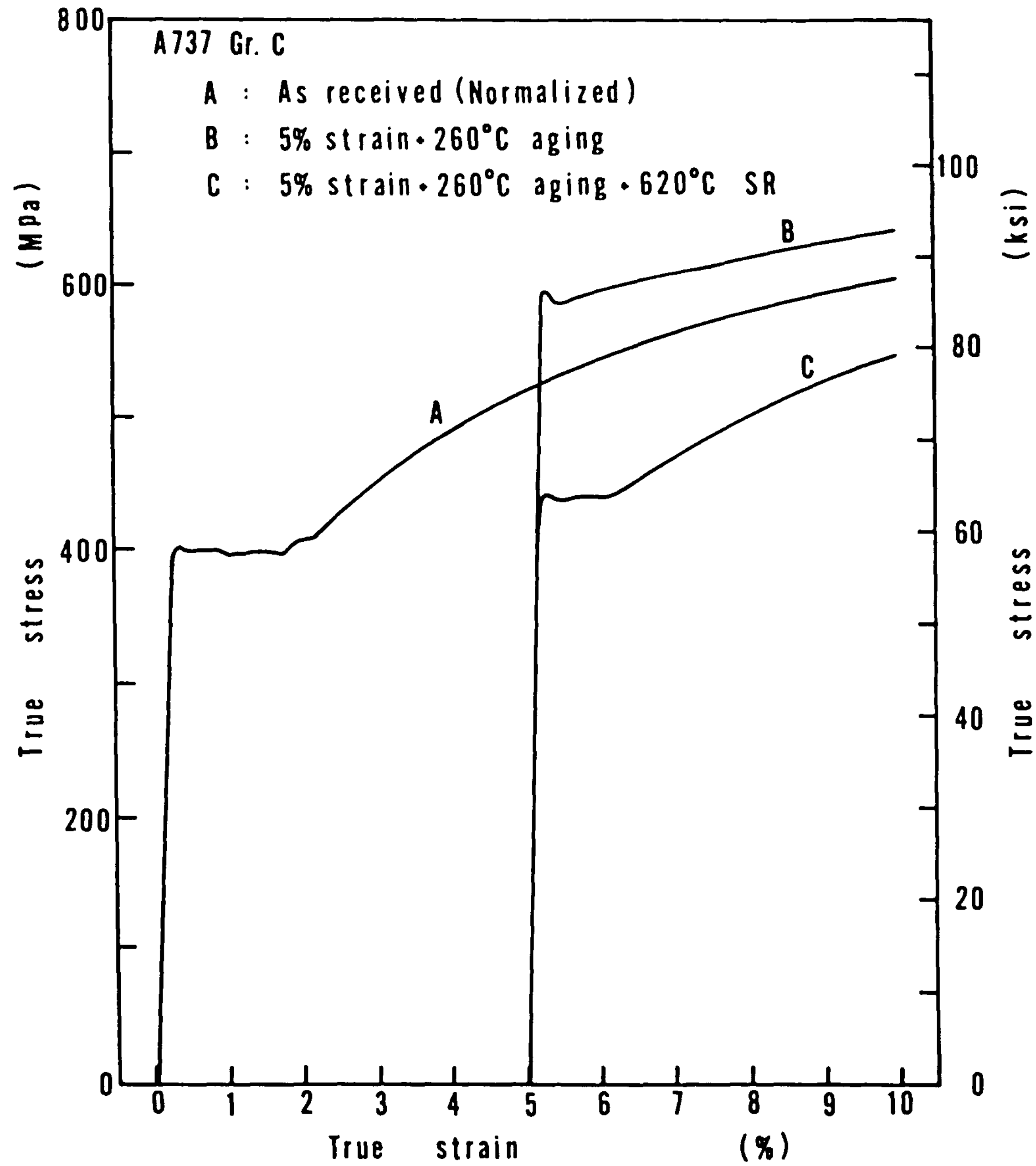
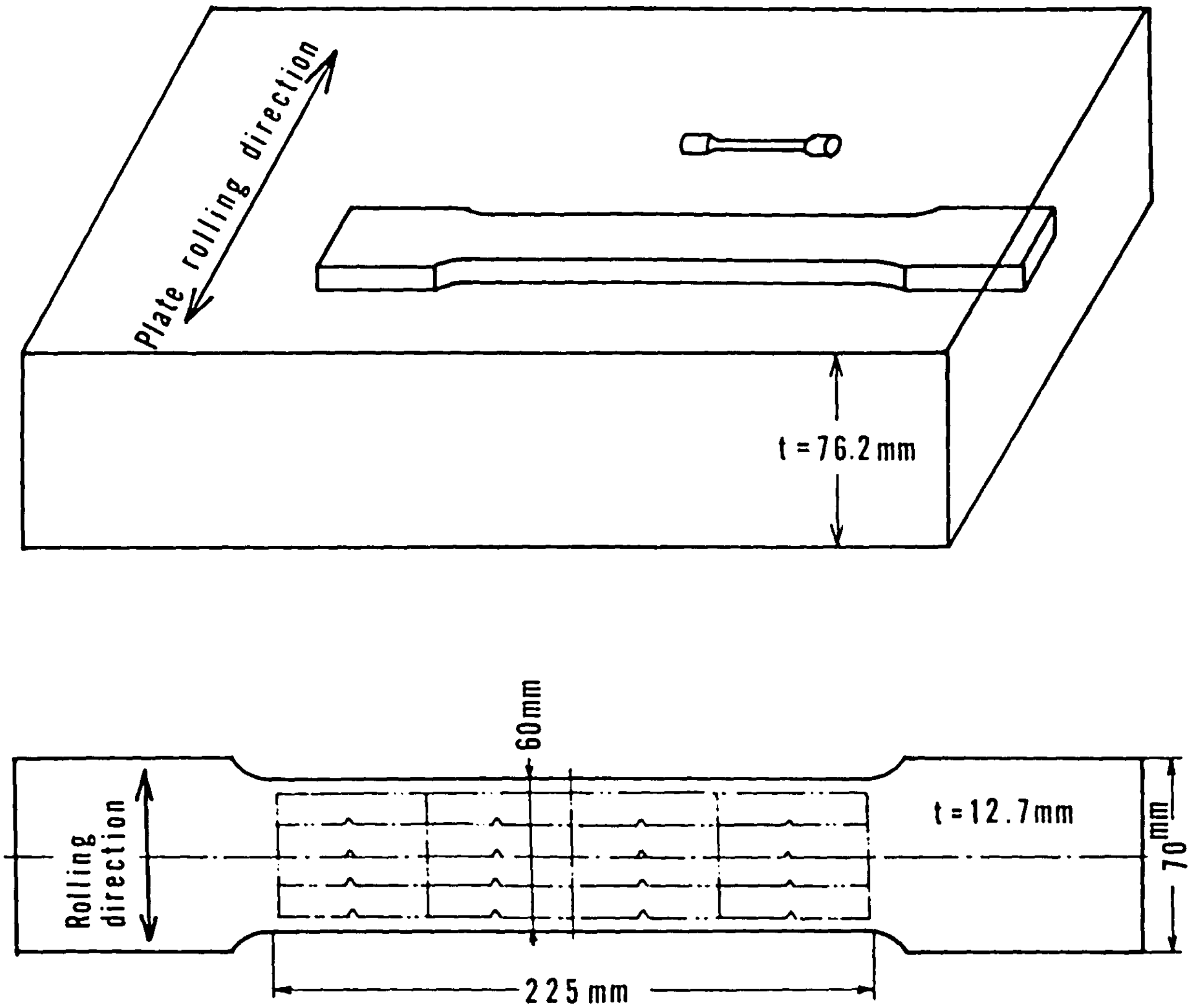


Fig. 1 True stress - true strain curves of A737 Grade C normalized plate



Strain : Transverse to rolling direction

Specimen : Longitudinal to strain
(Transverse to rolling direction)

Fig. 2 Specimen orientation and sampling method

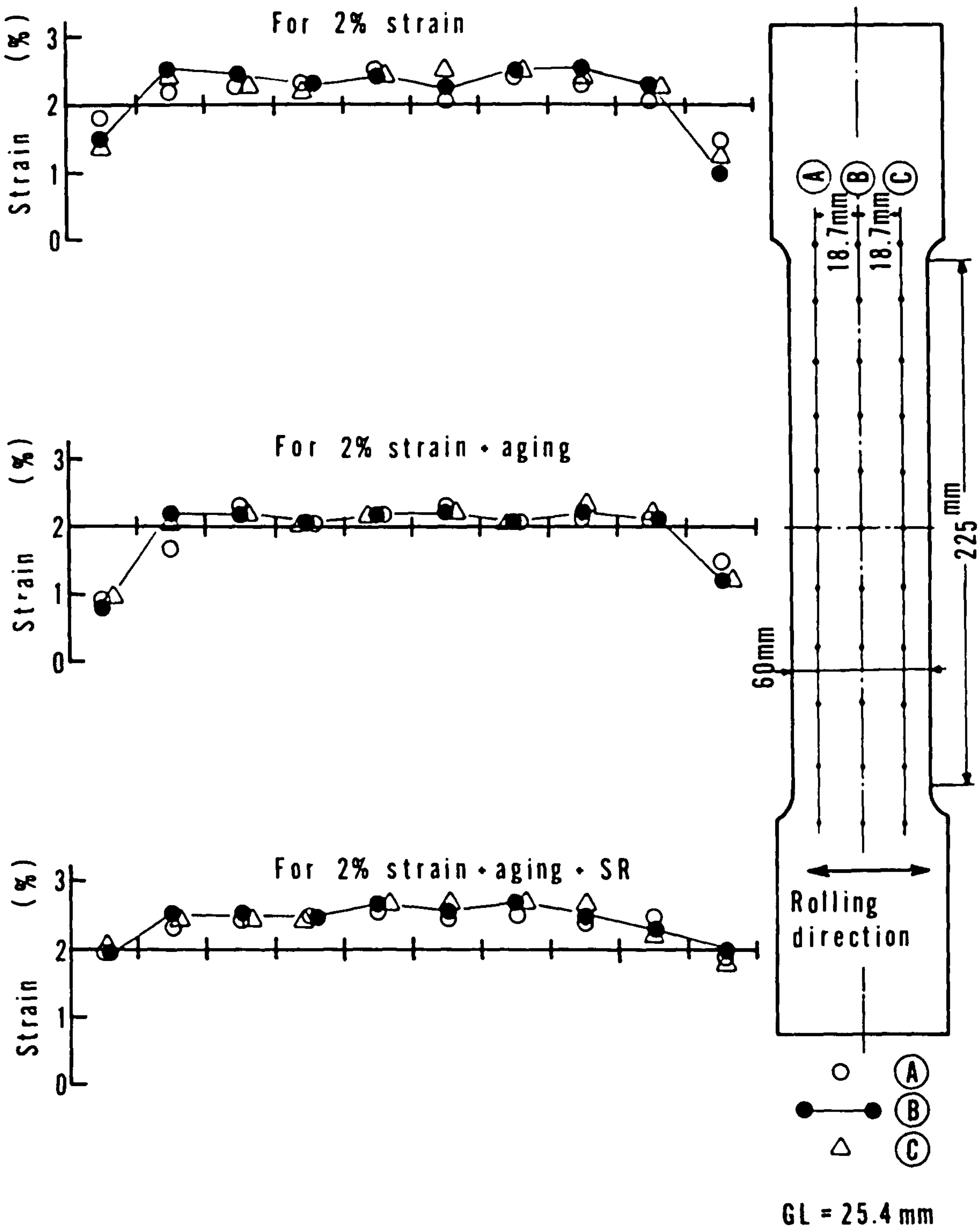


Fig. 3 Distribution of nominal 2% tensile strain

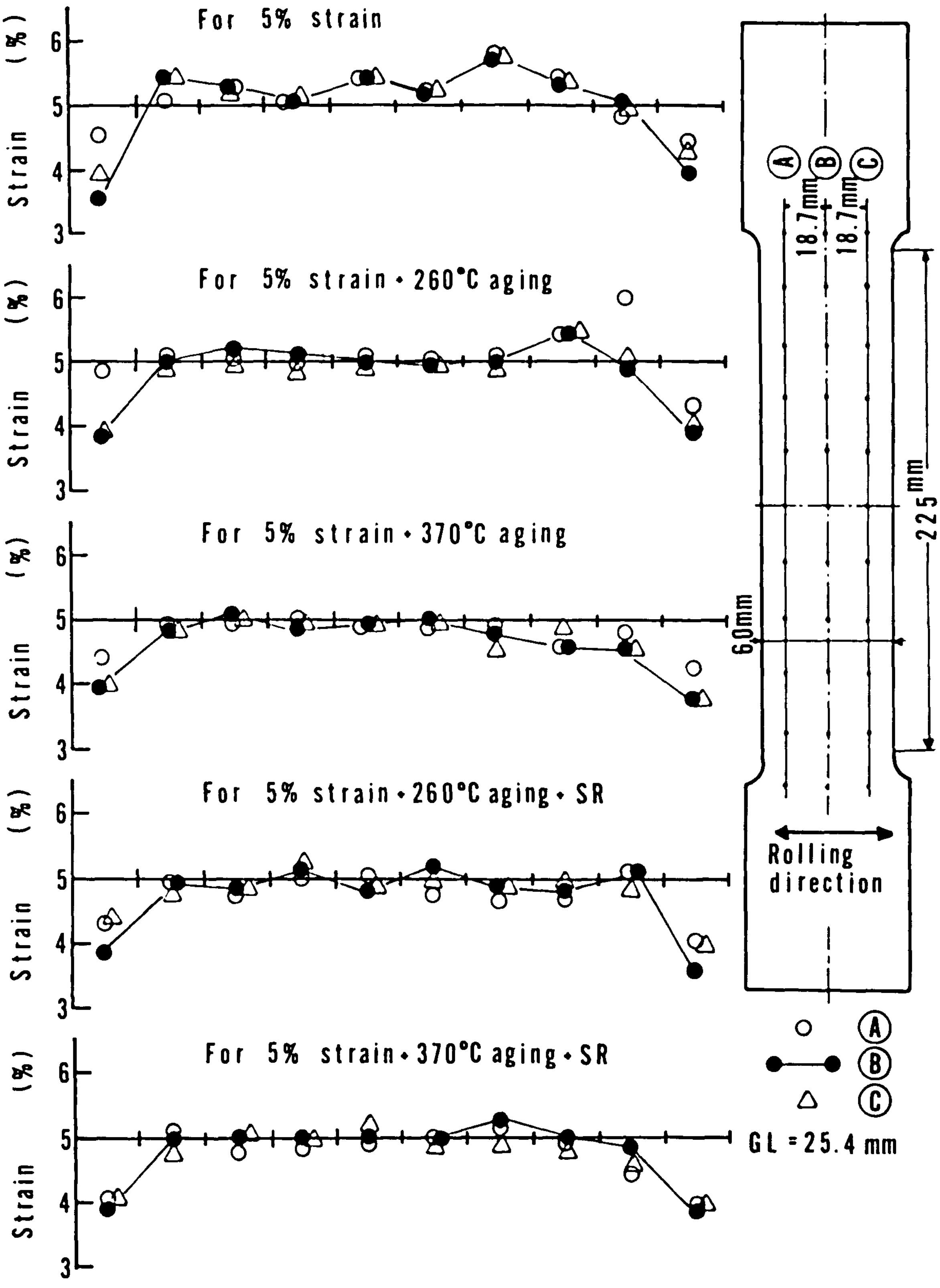


Fig. 4 Distribution of nominal 5% tensile strain

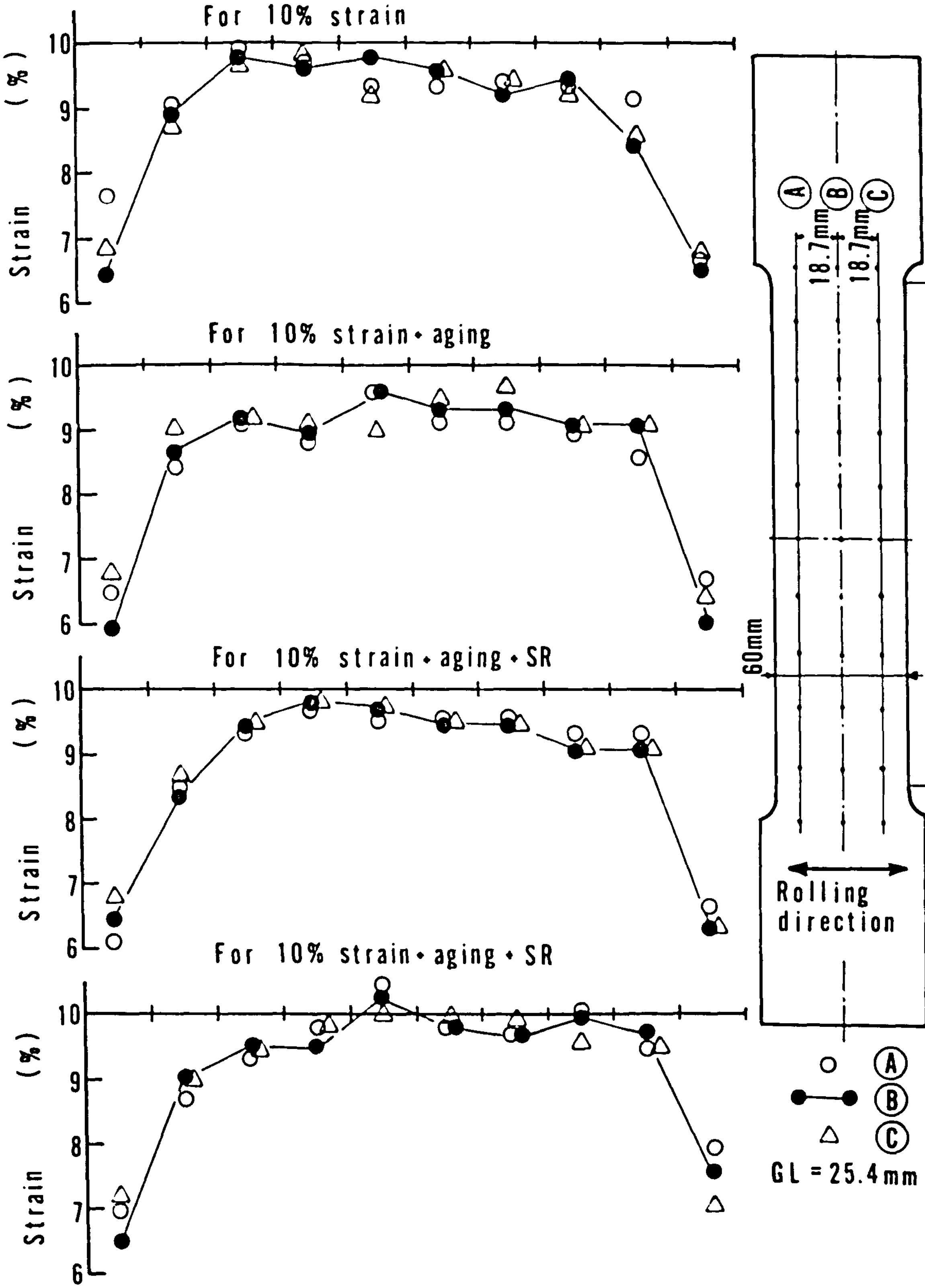


Fig. 5 Distribution on nominal 10% tensile strain

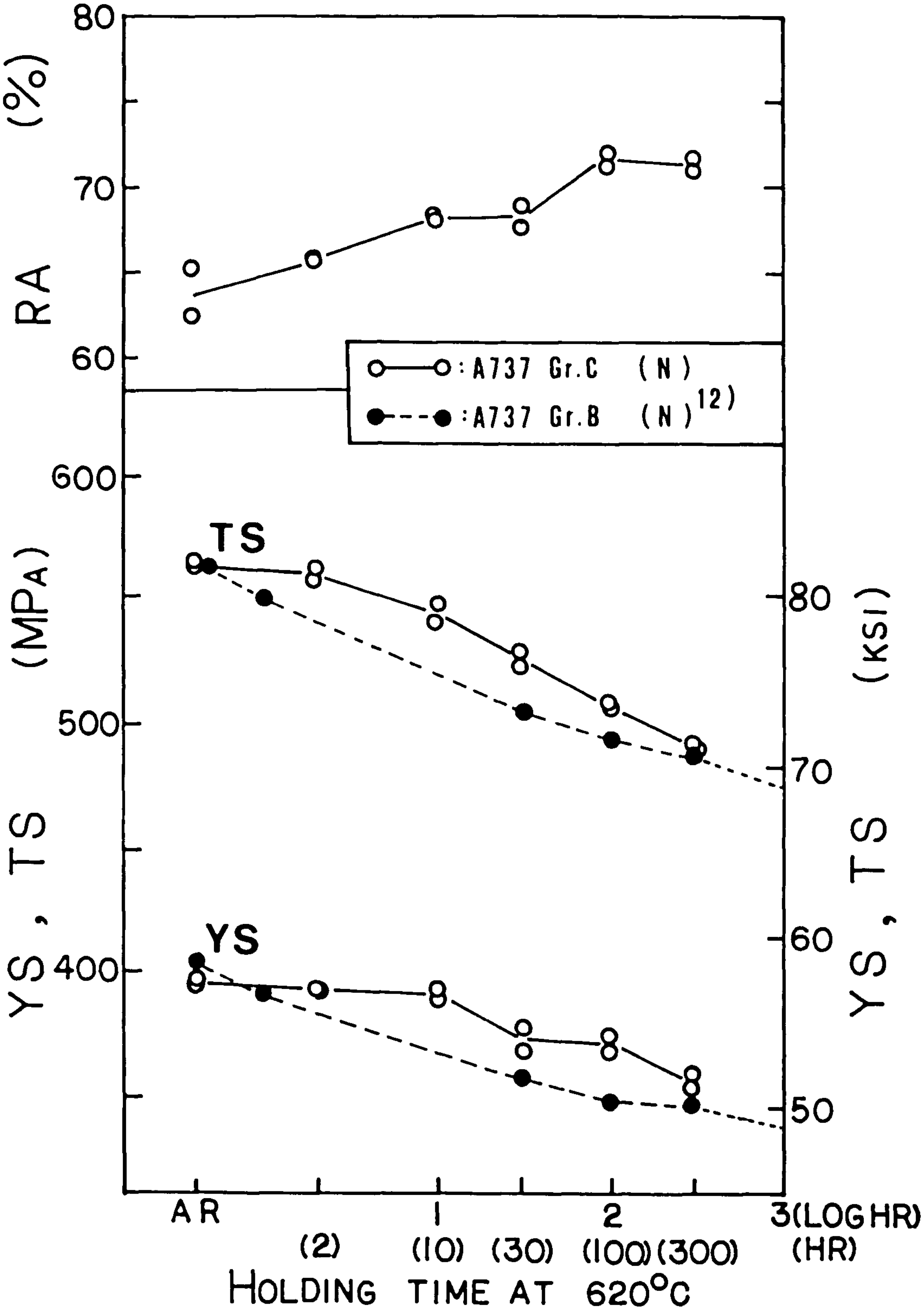


Fig. 6 Effect of stress relief heat treatment on the tensile properties for A737 Grade C normalized plate

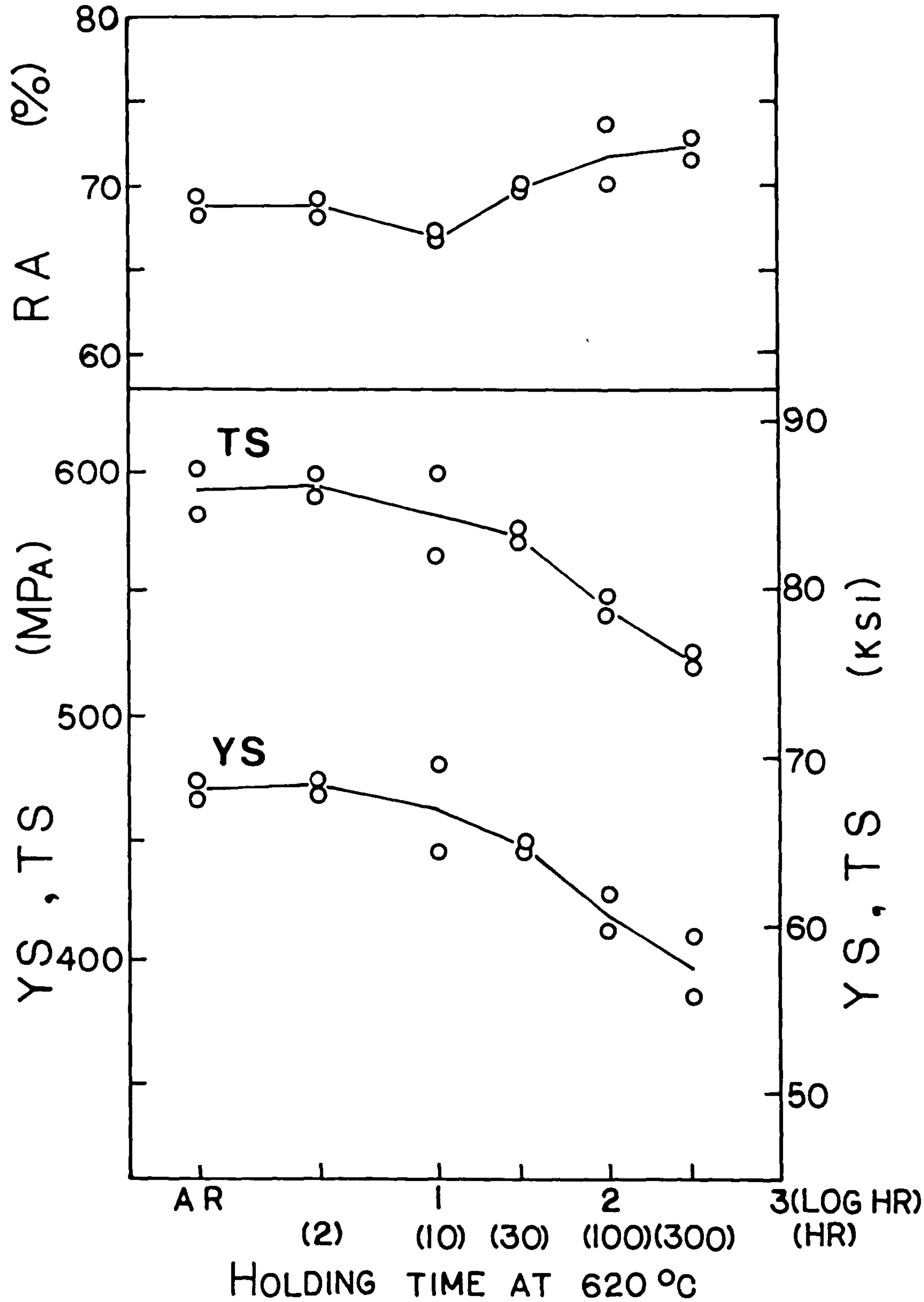


Fig. 7 Effect of stress relief heat treatment on the tensile properties of A737 Grade C Q & T plate

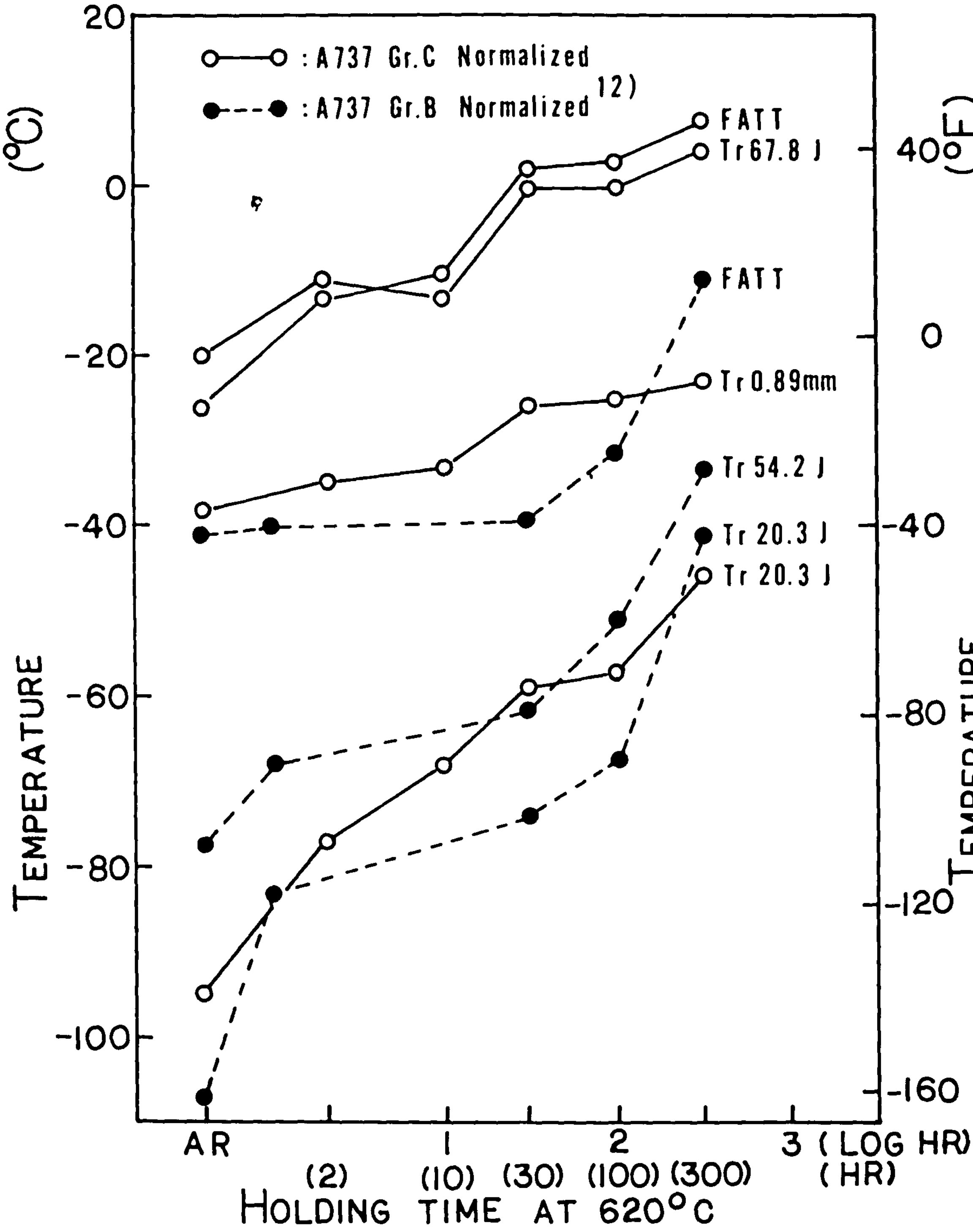


Fig. 8 Effect of stress relief heat treatment on the impact transition temperatures for A737 Gr. C normalized plate

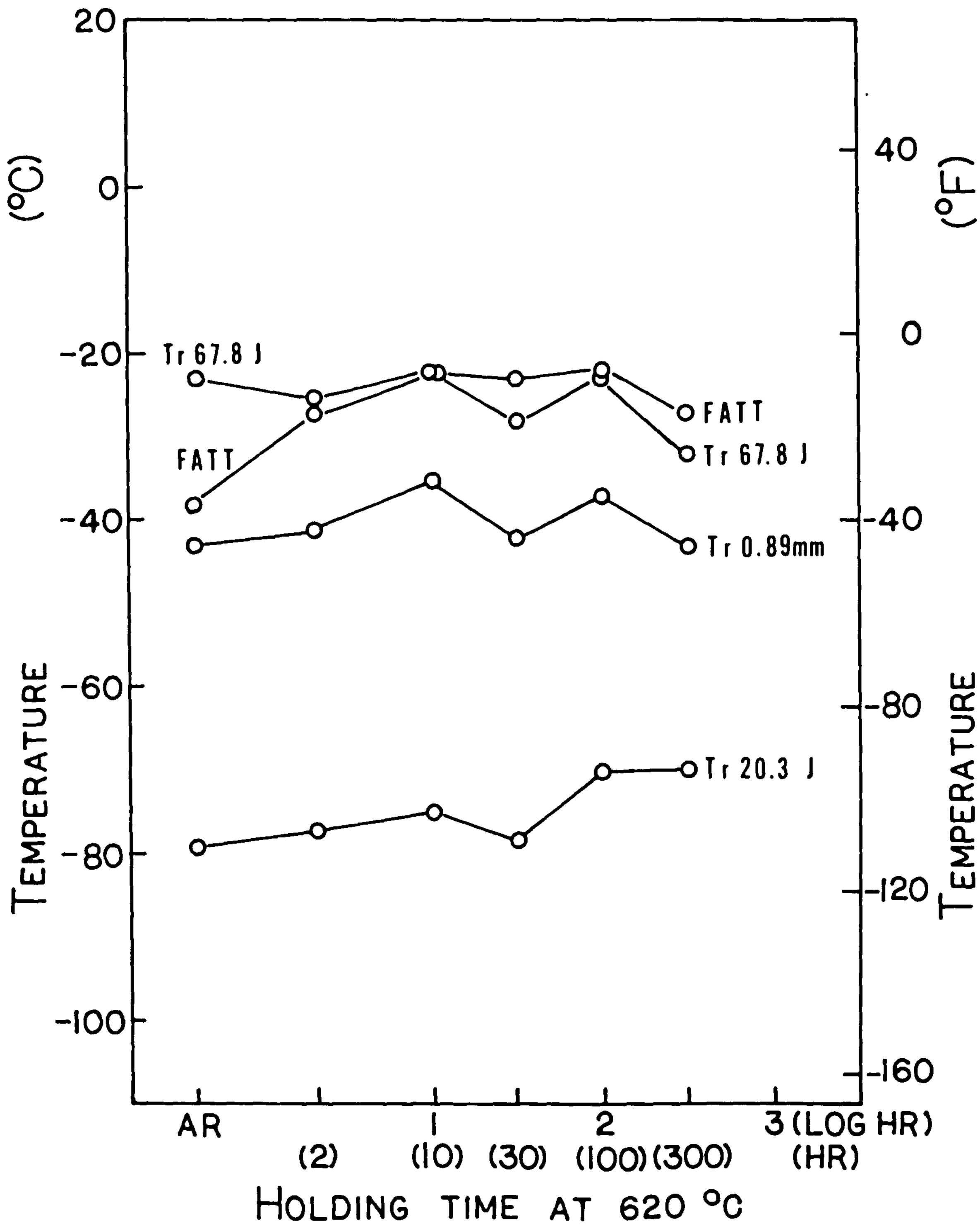


Fig. 9 Effect of stress relief heat treatment on the impact transition temperatures for A737 Gr. C quenched and tempered plate

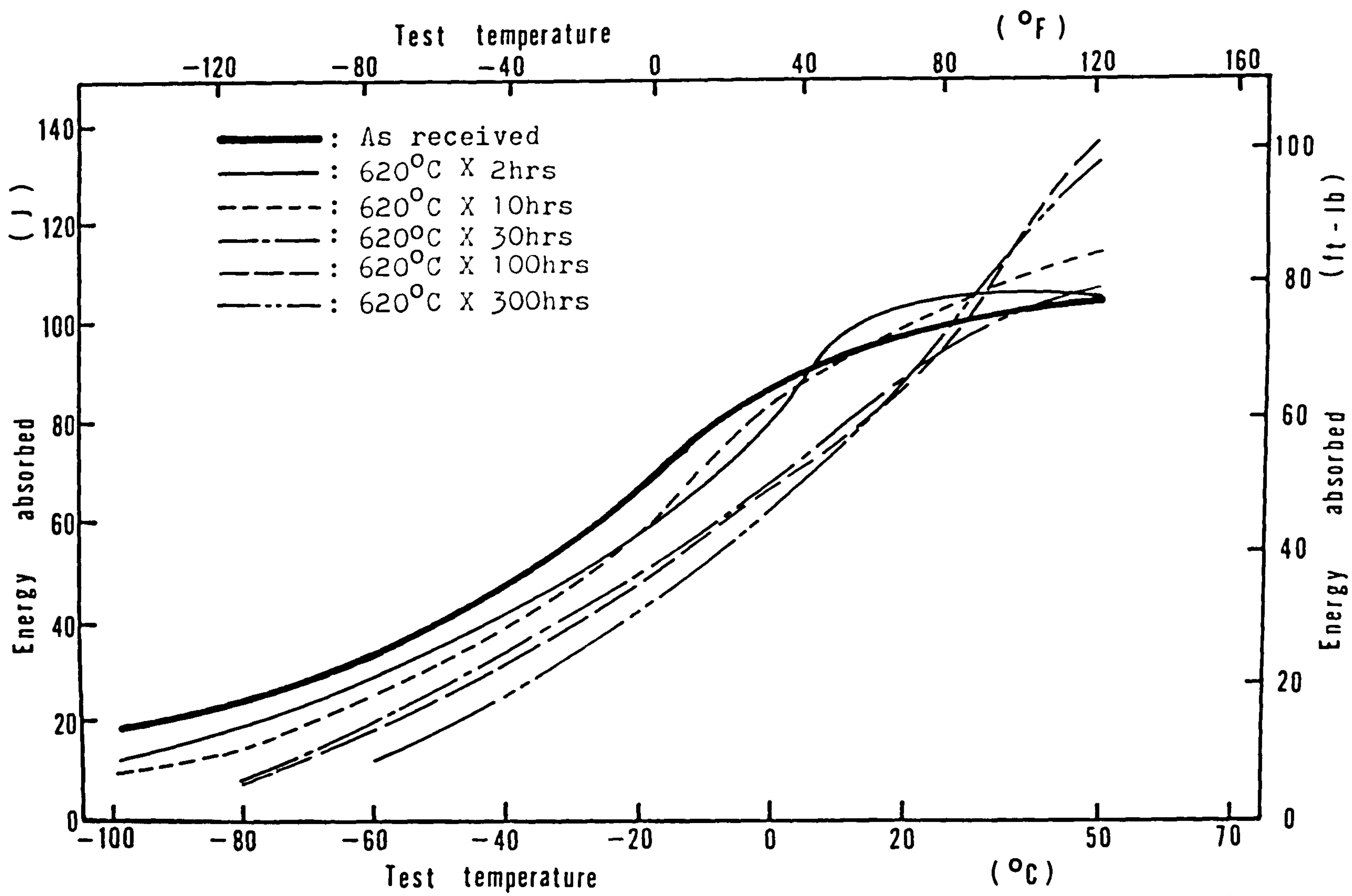


Fig. 10 Effect of stress relief heat treatment on the impact energy transition curves for A737 Grade C normalized plate

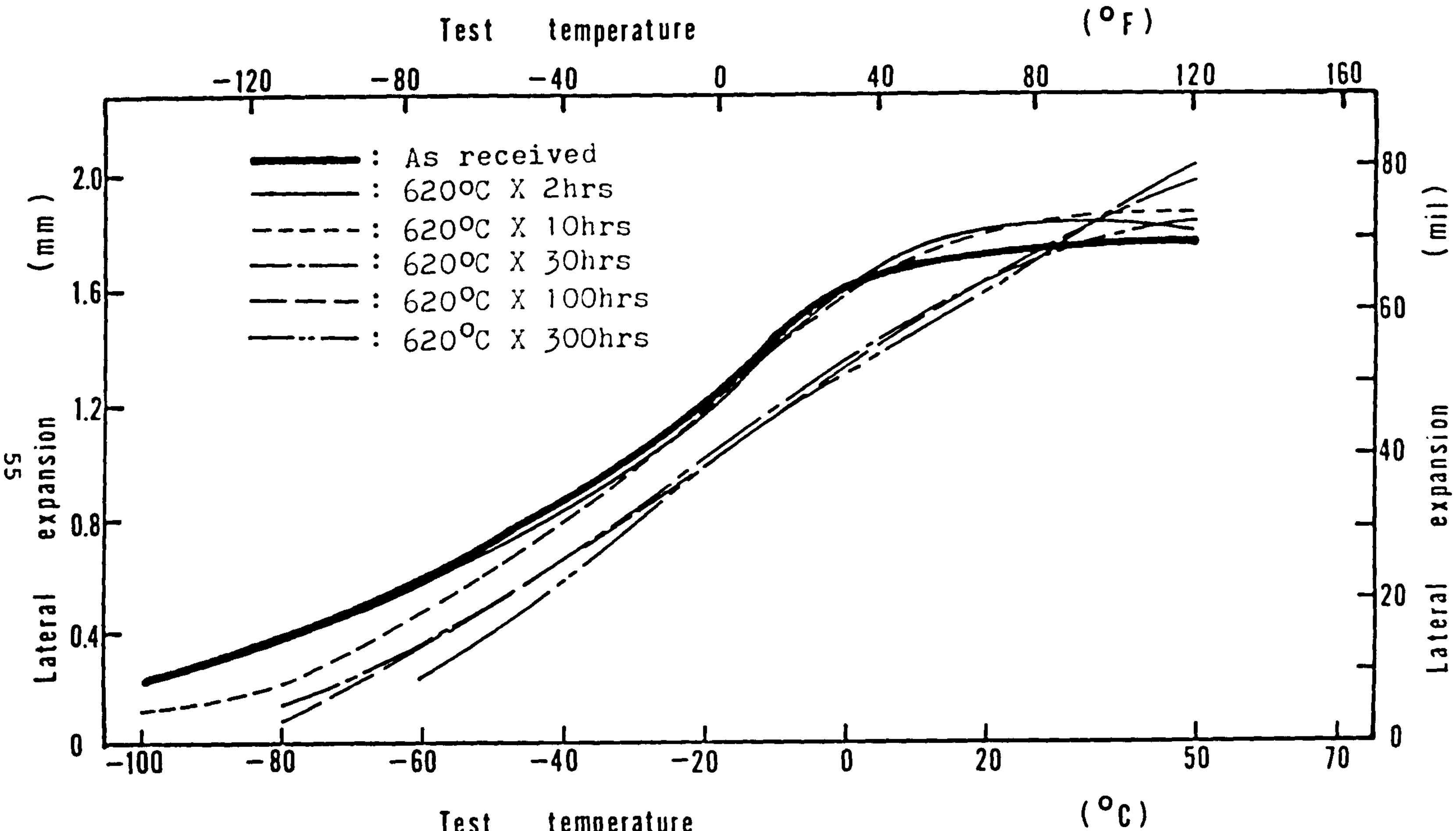


Fig. 11 Effect of stress relief heat treatment on the lateral expansion transition curves for A737 Grade C normalized plate

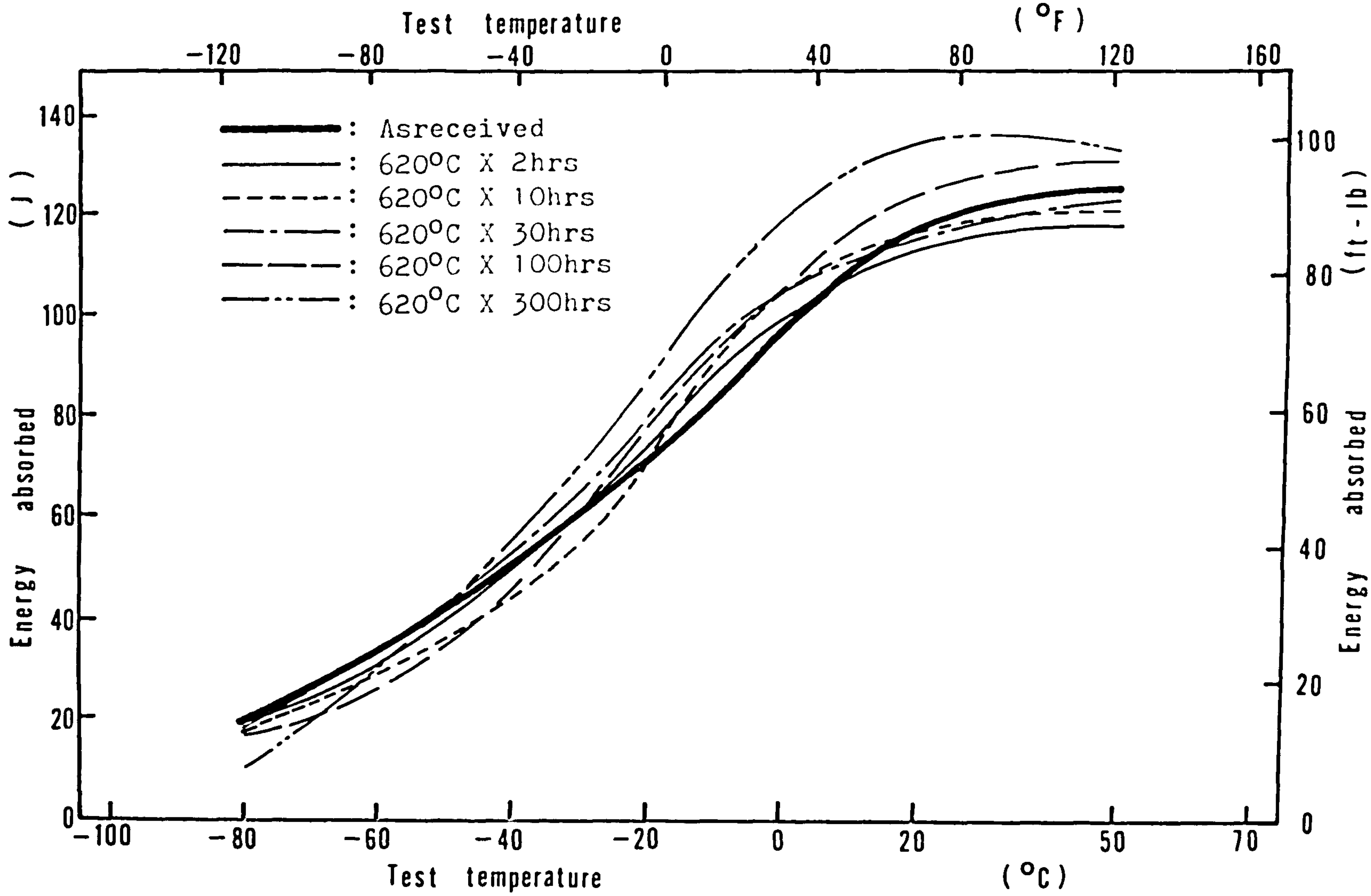


Fig. 12 Effect of stress relief heat treatment on the impact energy transition curves for A737 Grade C quenched and tempered plate

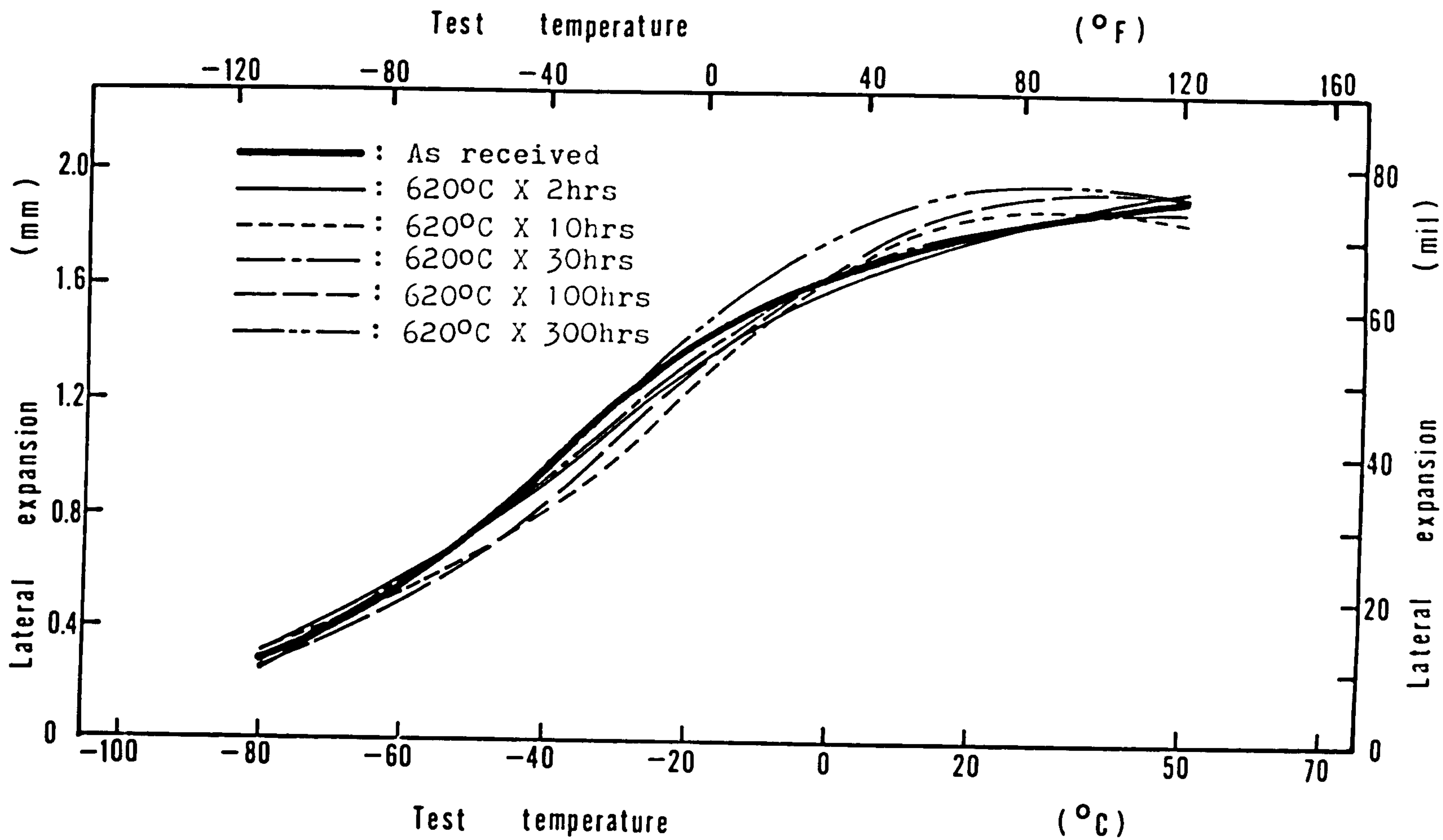


Fig. 13 Effect of stress relief heat treatment on the lateral expansion transition curves for A737 Grade C quenched and tempered plate

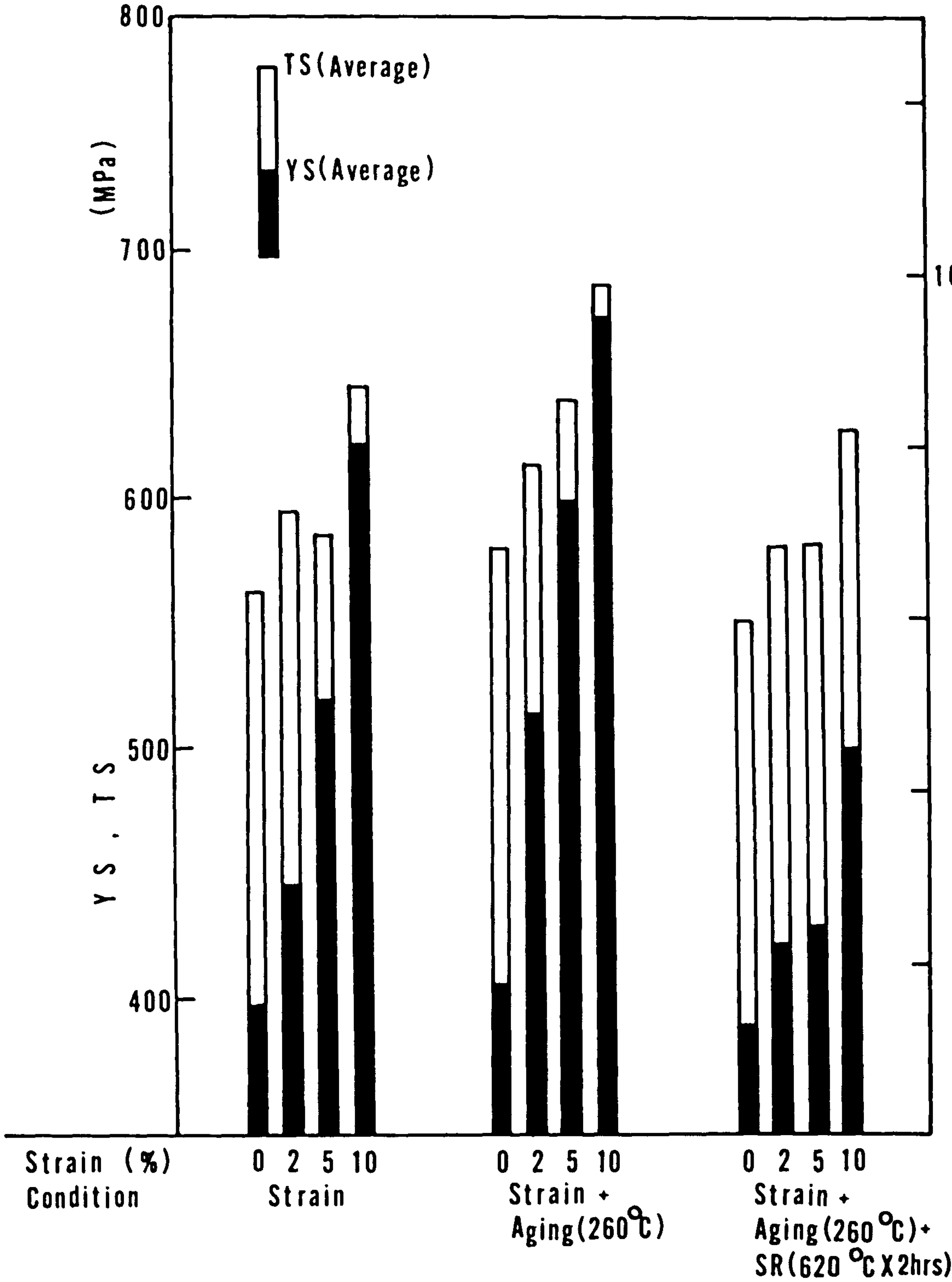


Fig. 14 Effect of strain aging on the yield strength and the tensile strength for A737 Gr. C normalized plate

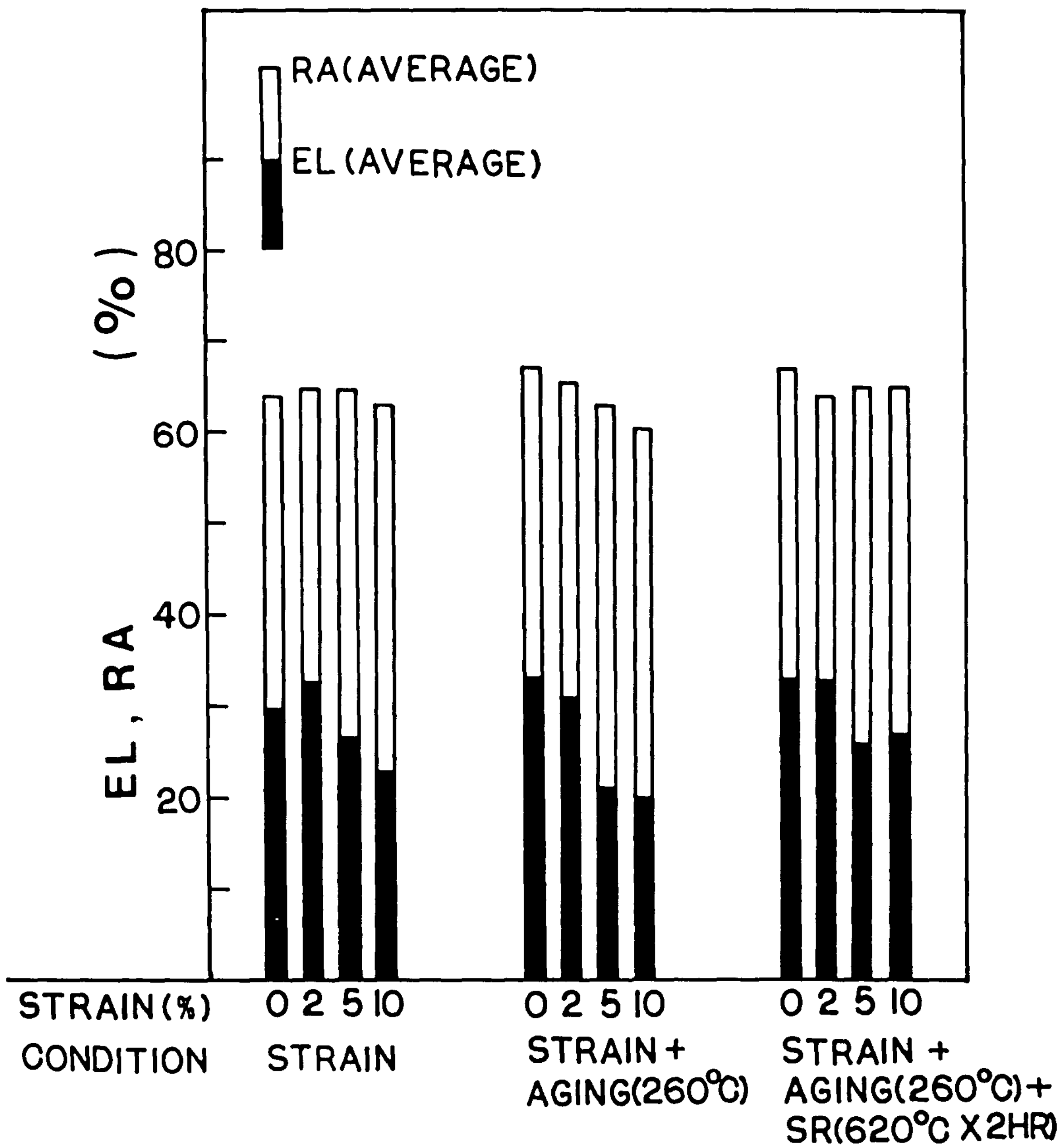


Fig. 15 Effect of strain aging on the elongation and the reduction of area for A737 Gr. C normalized plate

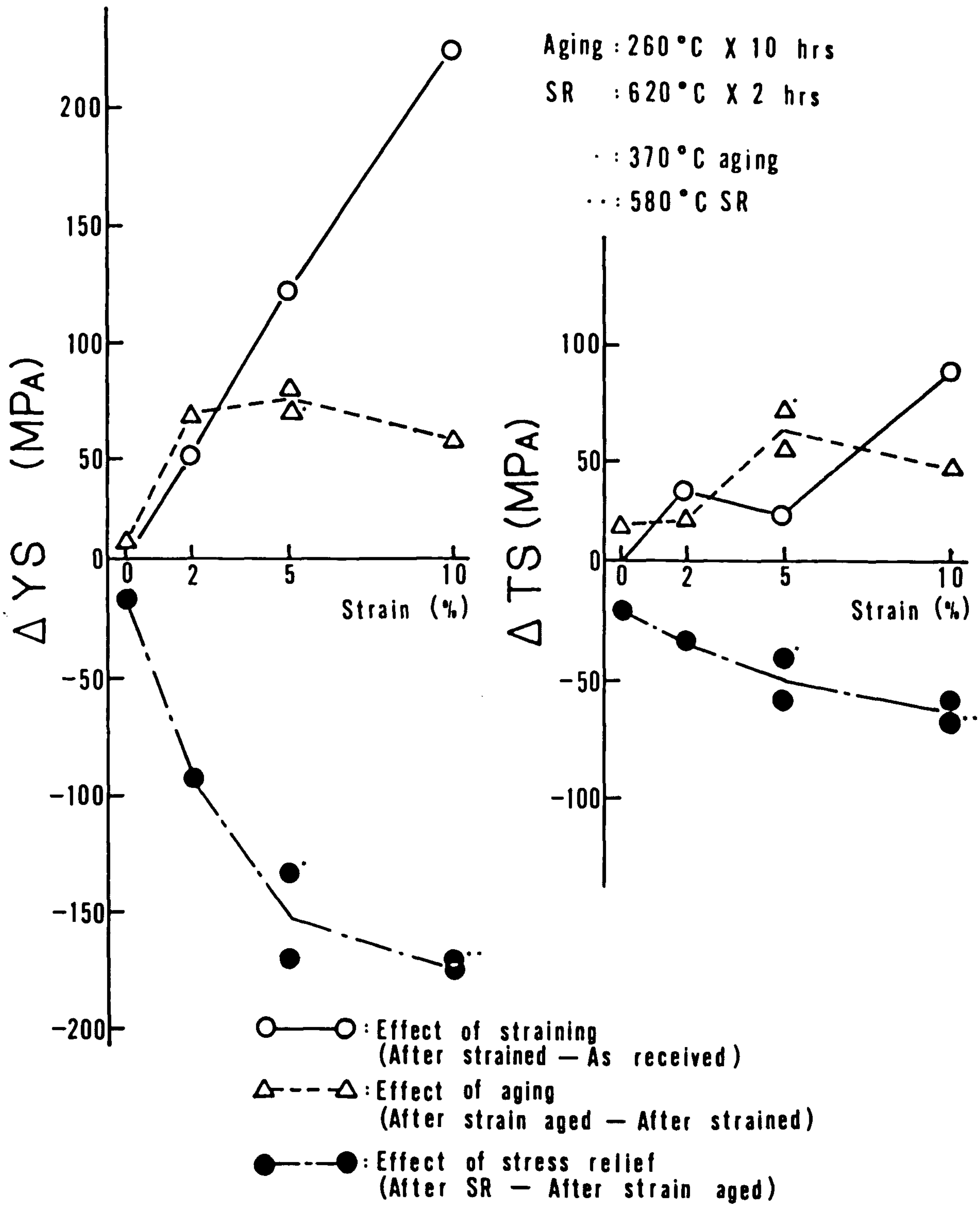


Fig. 16 Effect of straining, aging, and stress relief on the yield strength and the tensile strength

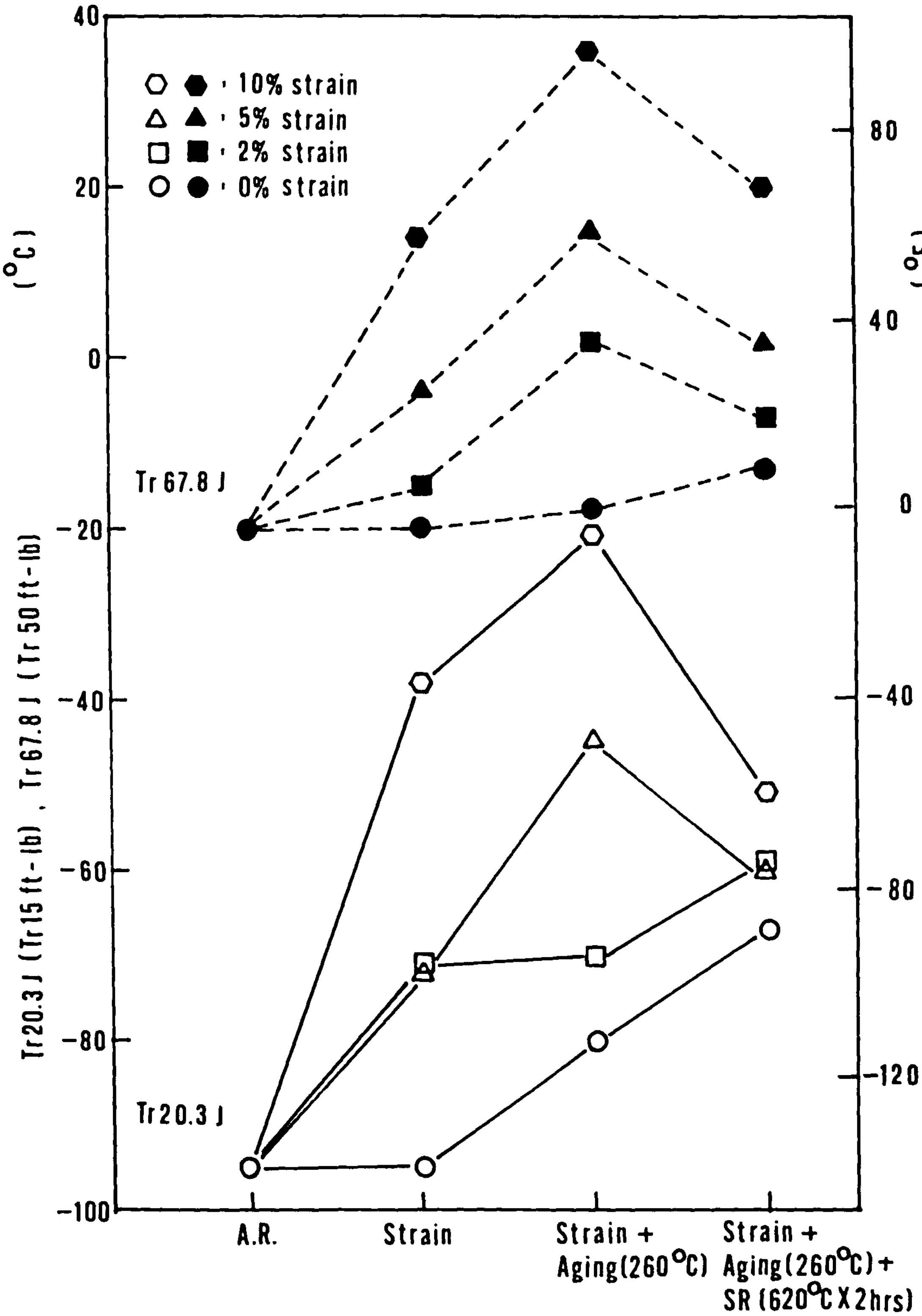


Fig. 17 Effect of strain aging on the Tr20.3J and the Tr67.8J for A737 Gr. C normalized plate

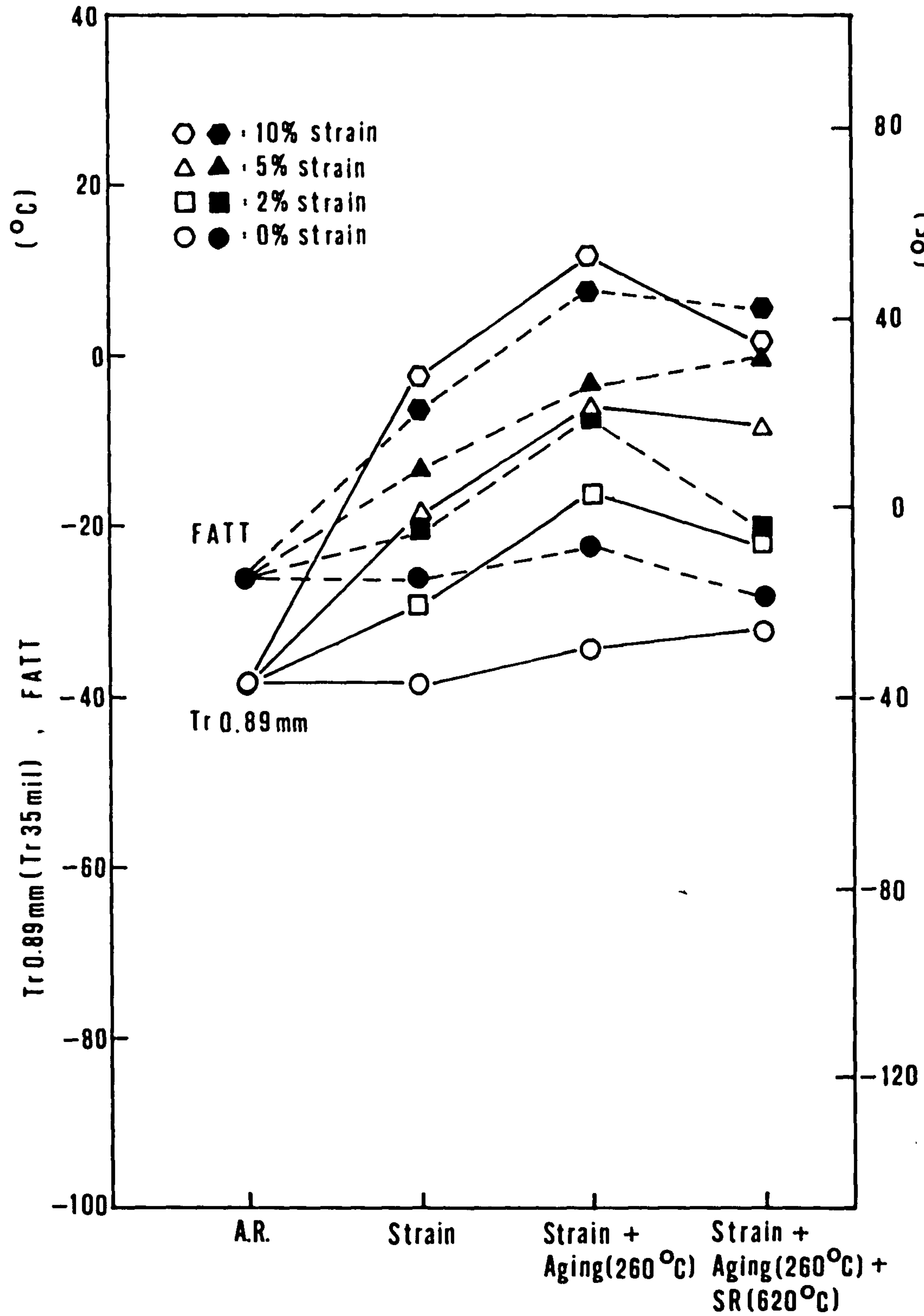


Fig. 18 Effect of strain aging on the Tr0.89mm and the FATT for A737 Gr. C normalized plate

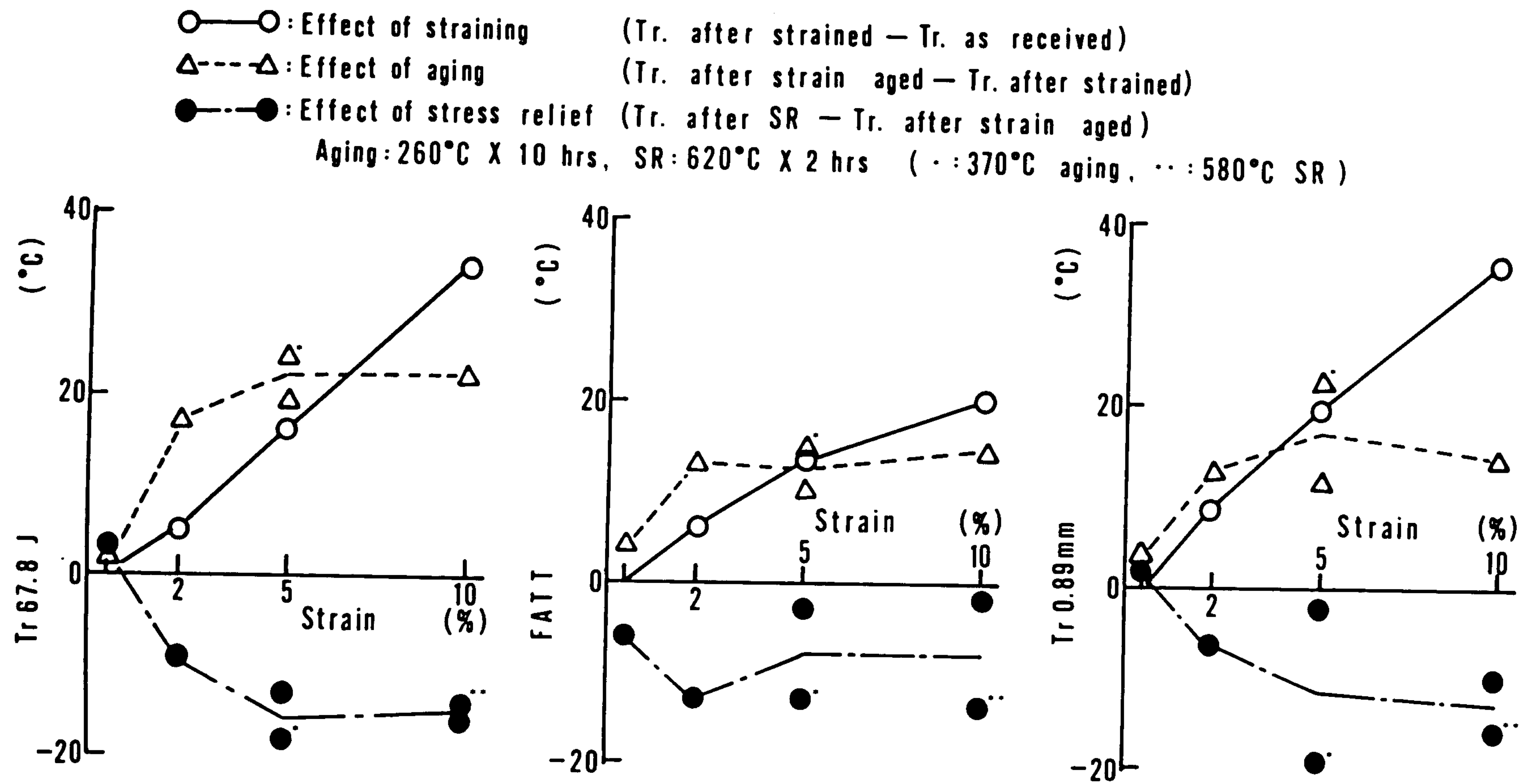


Fig. 19 Effect of straining, aging, and stress relief on the Tr67.8J, Tr0.89mm, and FATT

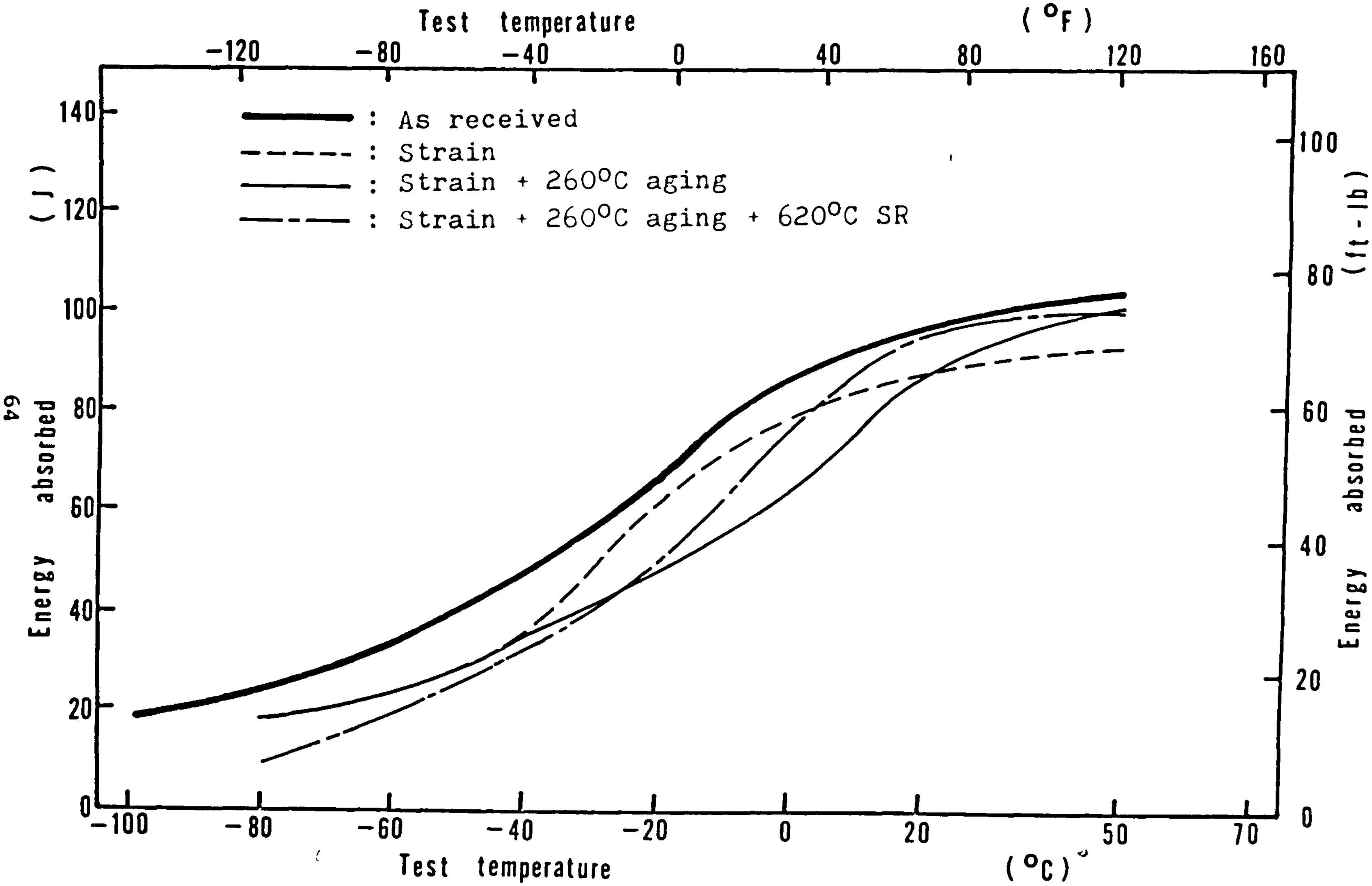


Fig. 20 Effect of 2% strain aging on the impact energy transition curves for A737 Grade C normalized plate

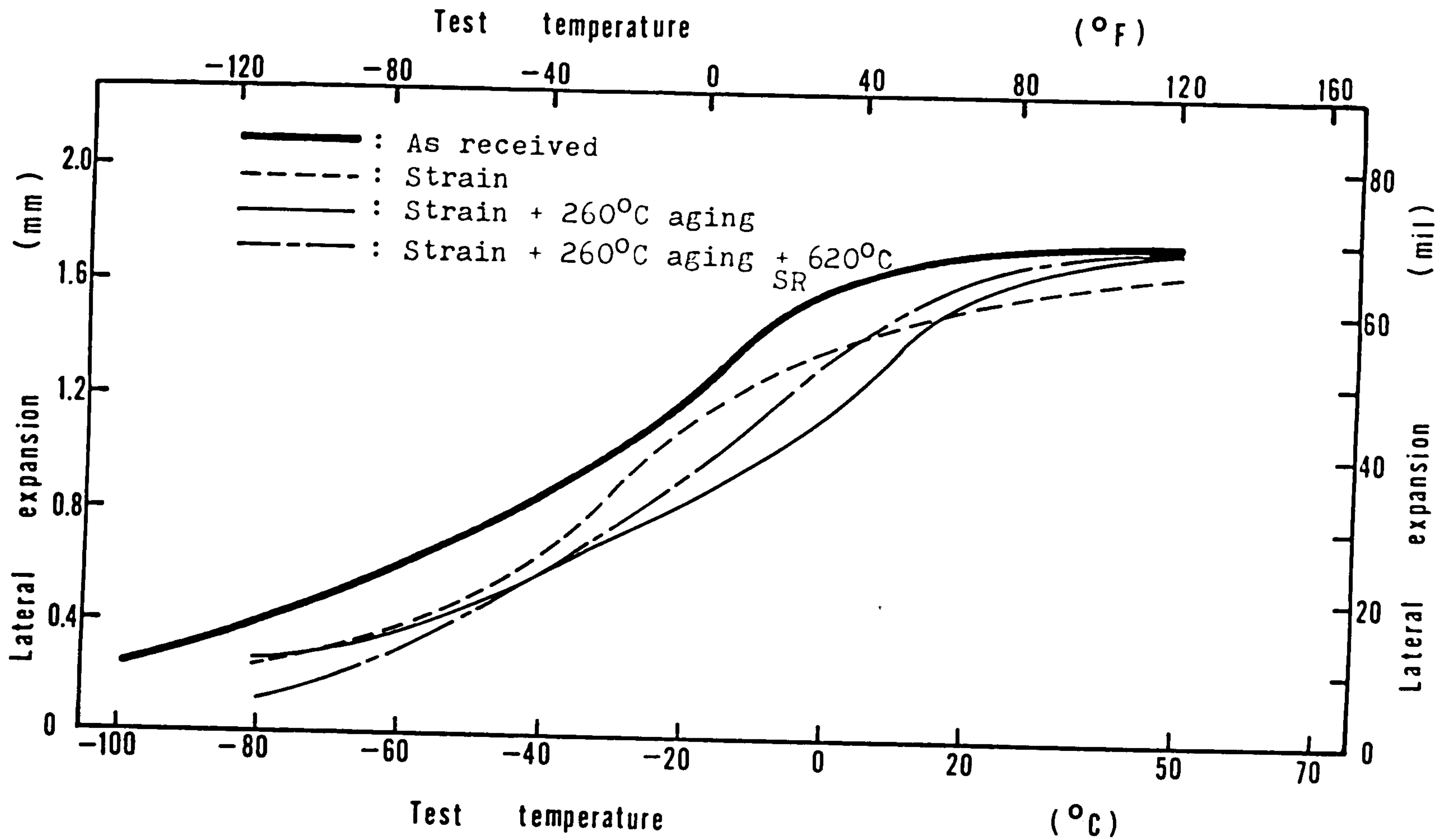


Fig. 21 Effect of 2% strain aging on the lateral expansion transition curves for A737 Grade C normalized plate

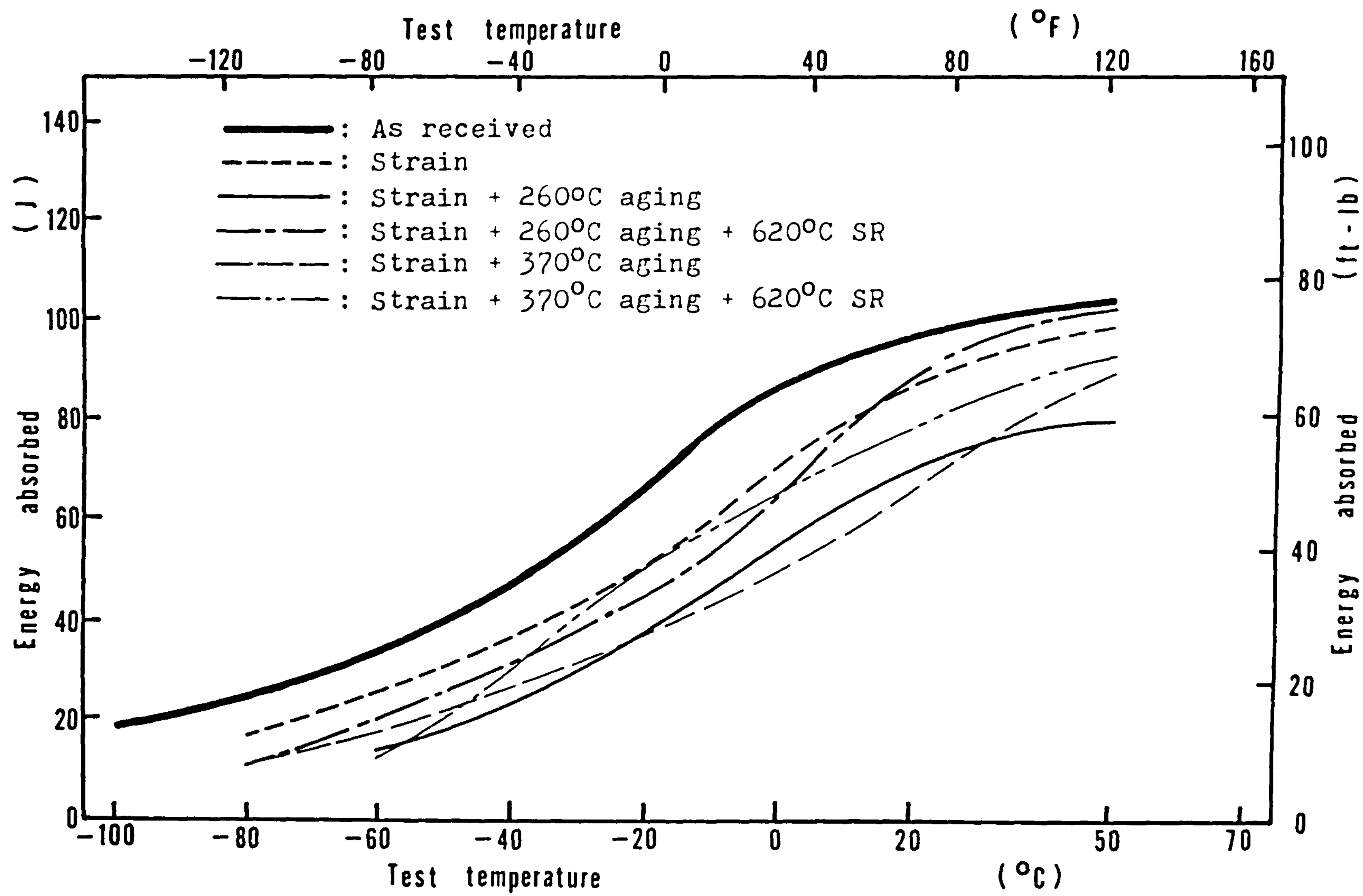


Fig. 22 Effect of 5% strain aging on the impact energy transition curves for A737 Grade C normalized plate

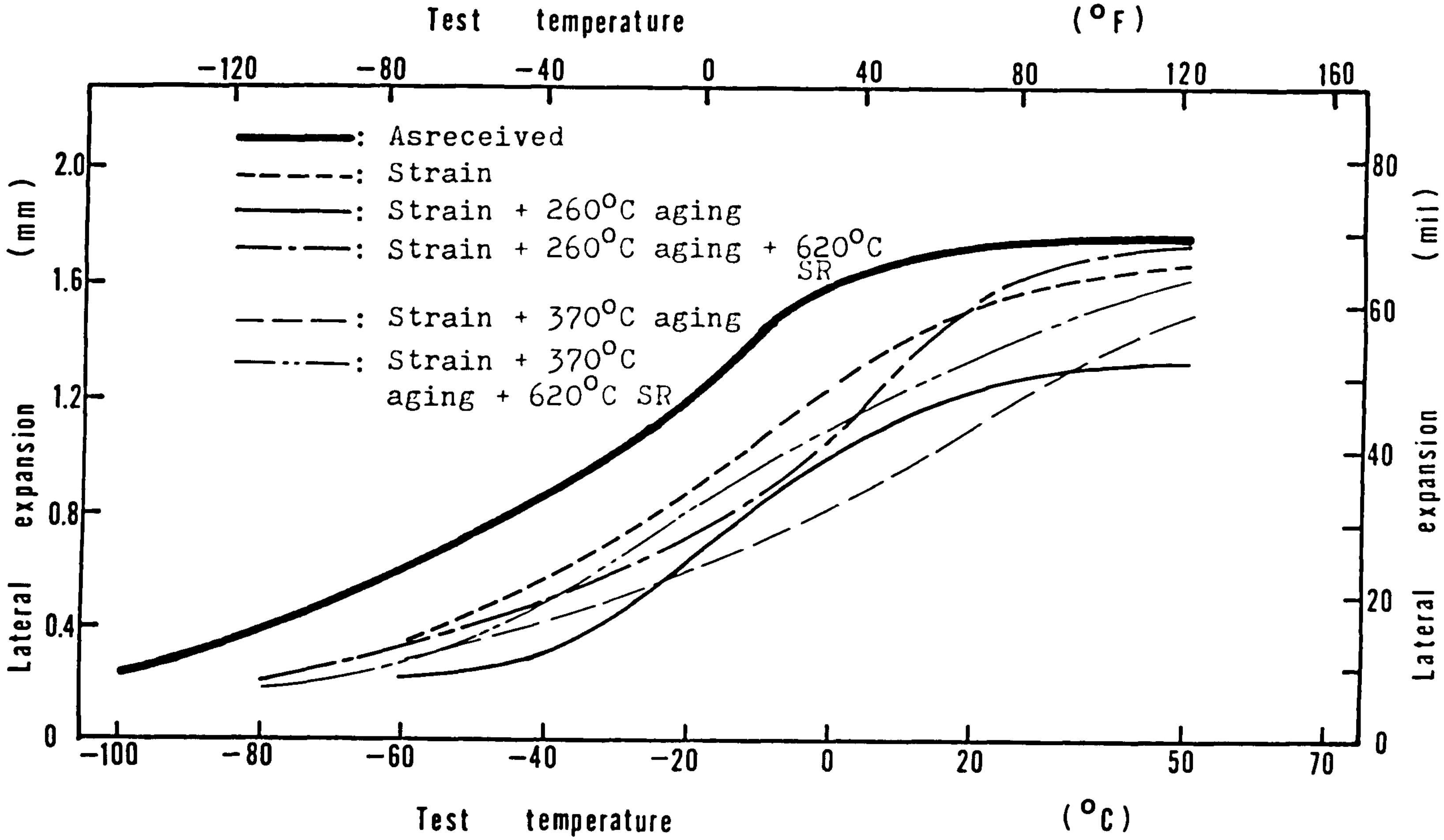


Fig. 23 Effect of 5% strain aging on the lateral expansion transition curves for A737 Grade C normalized plate

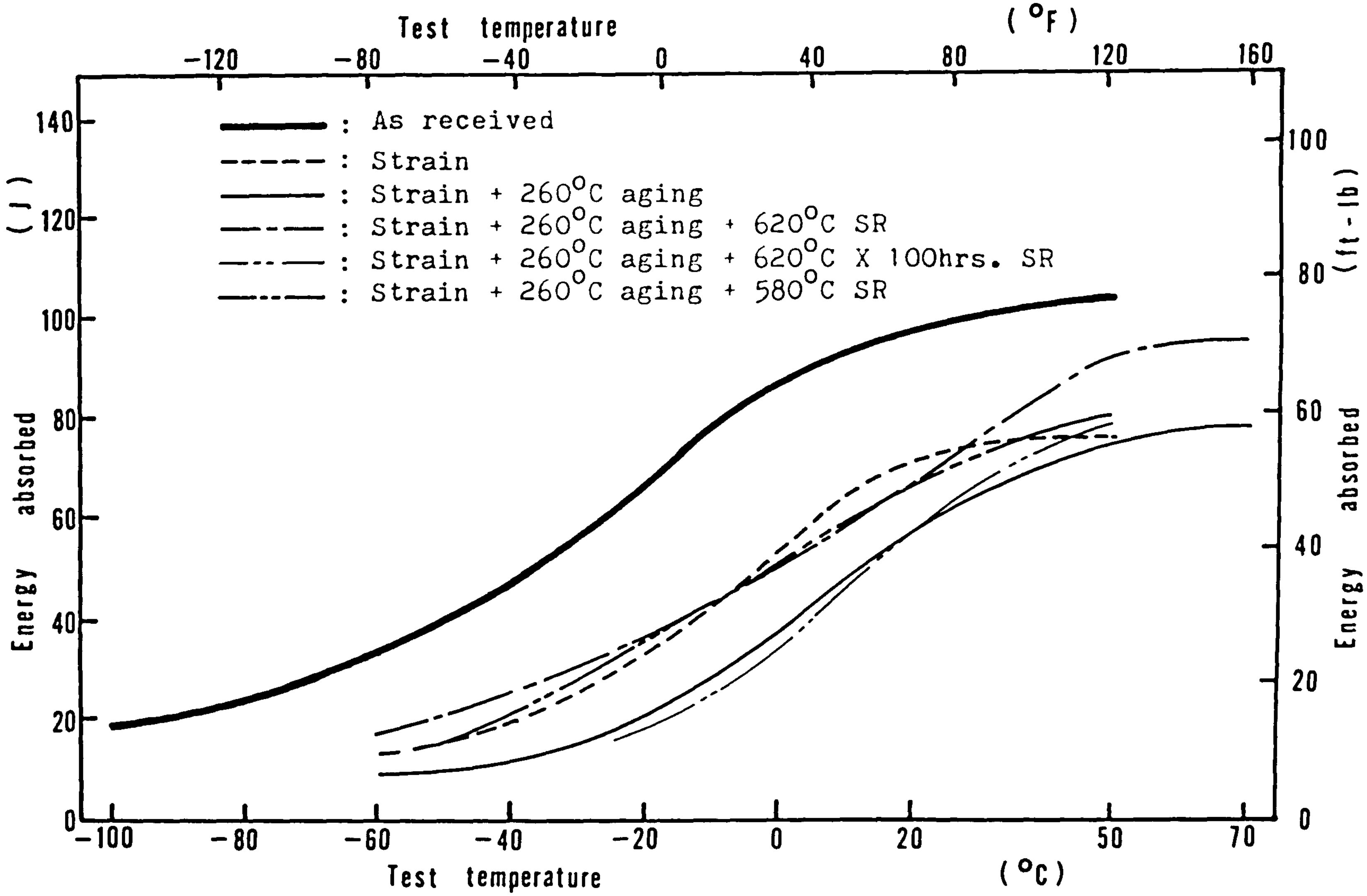


Fig. 24 Effect of 10% strain aging on the impact energy transition curves for A737 Grade C normalized plate

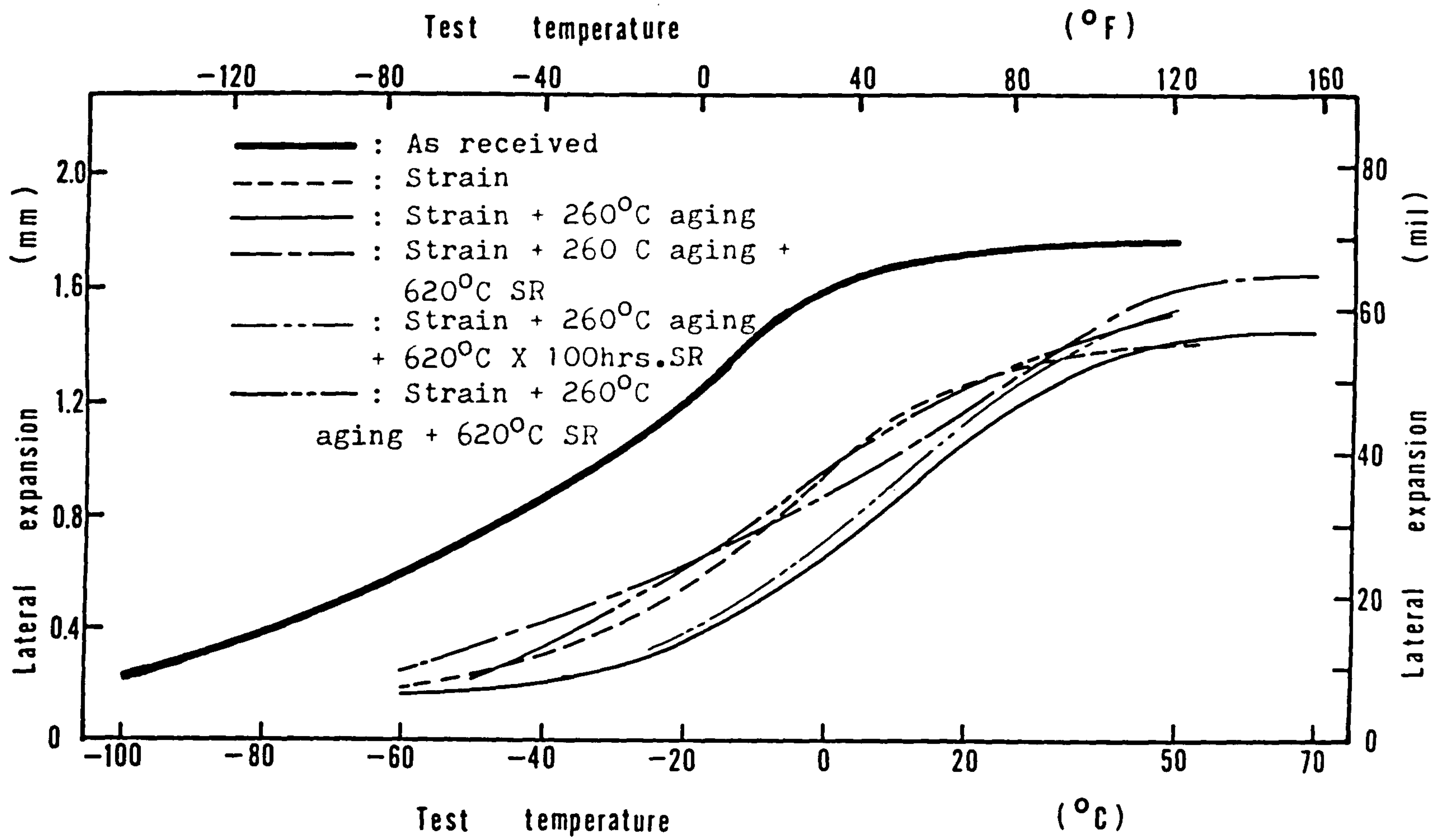


Fig. 25 Effect of 10% strain aging on the lateral expansion transition curves for A737 Grade C normalized plate

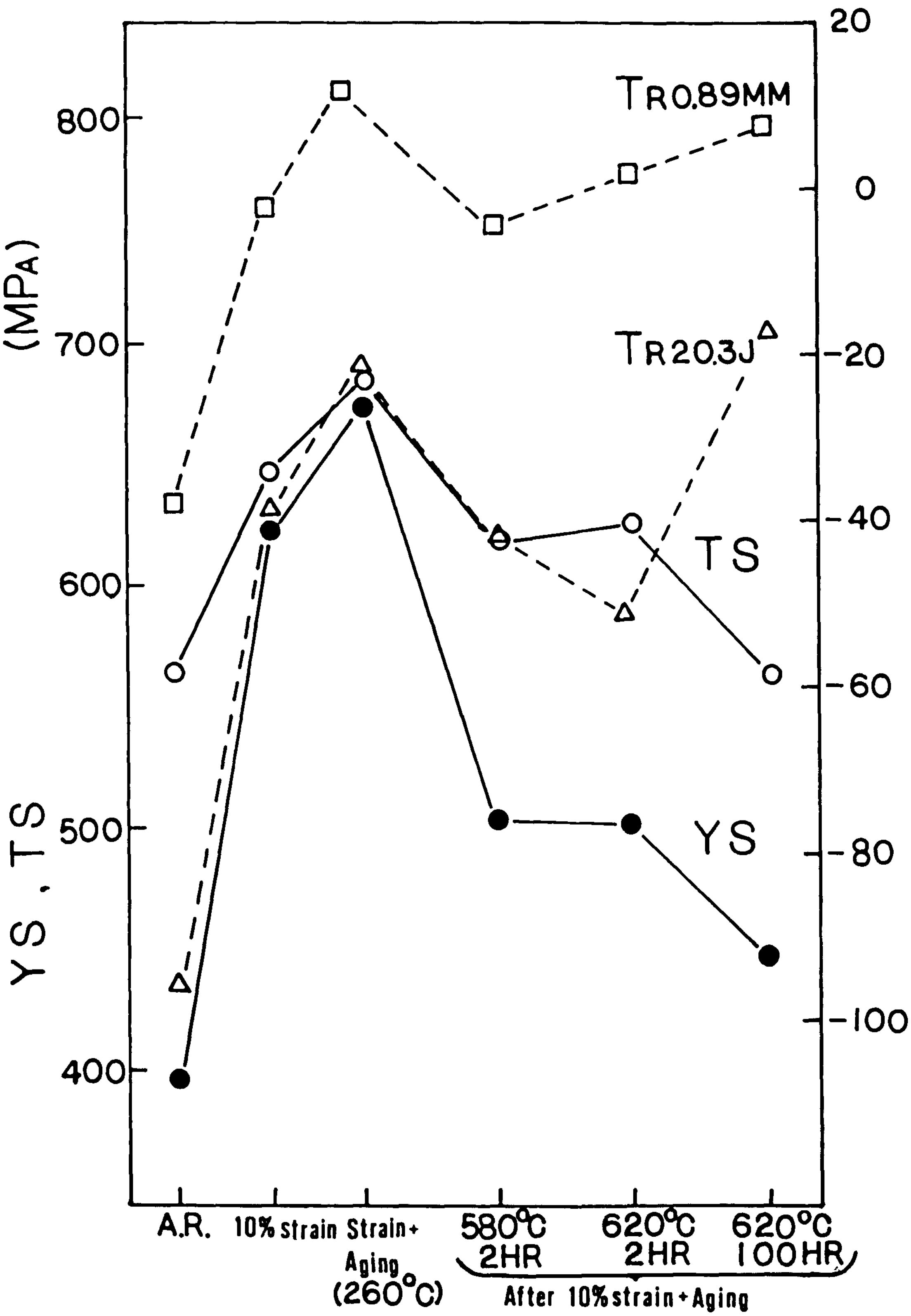
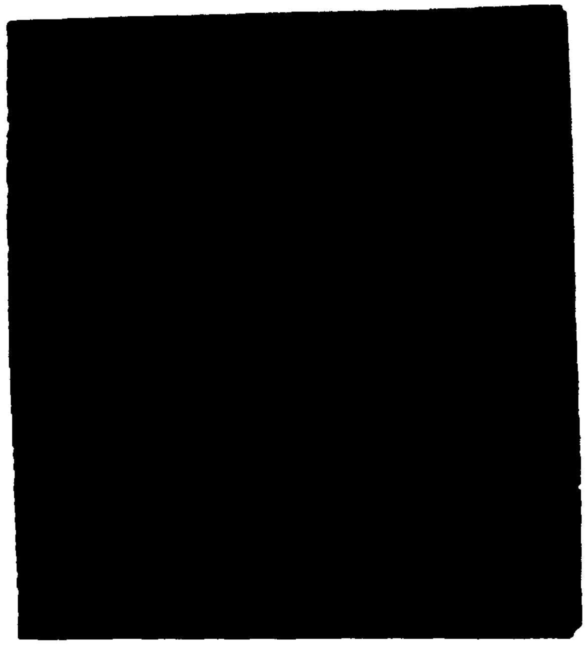
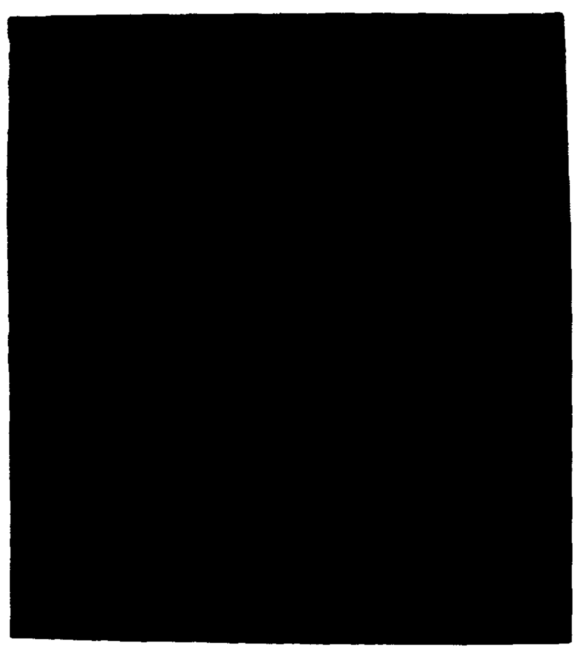


Fig. 26 Effect of stress relief after 10% strain aging on the tensile properties and the Charpy impact properties.

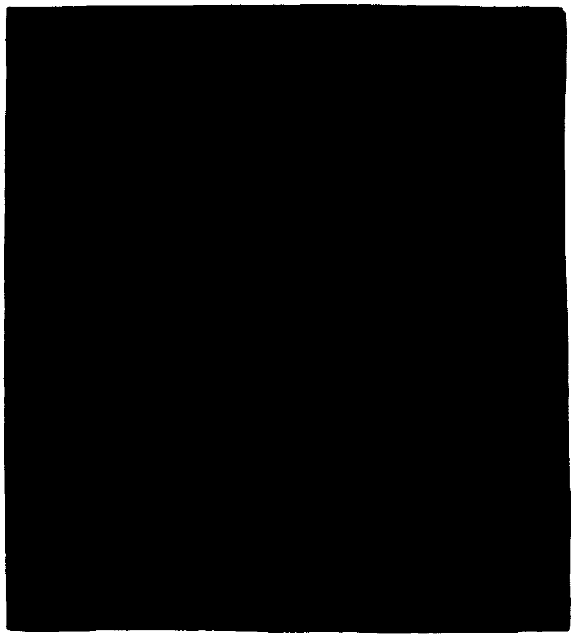


1a. As received (X100)

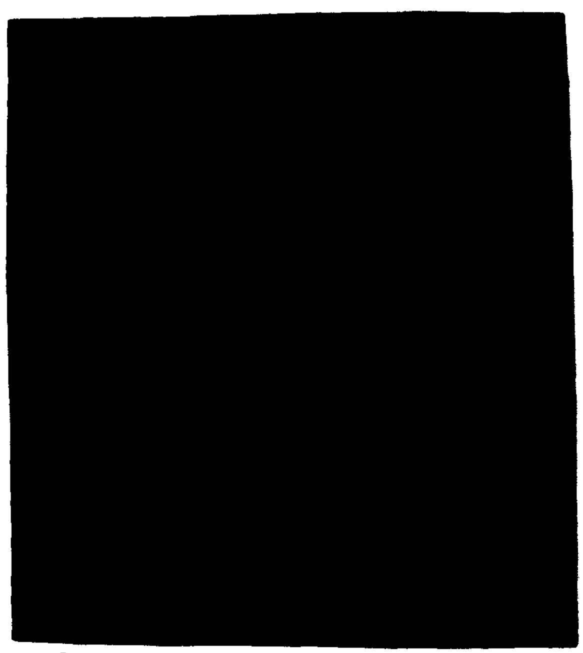


1b. As received (X500)

Micro. 1 Microstructures of A737 Grade C normalized plate

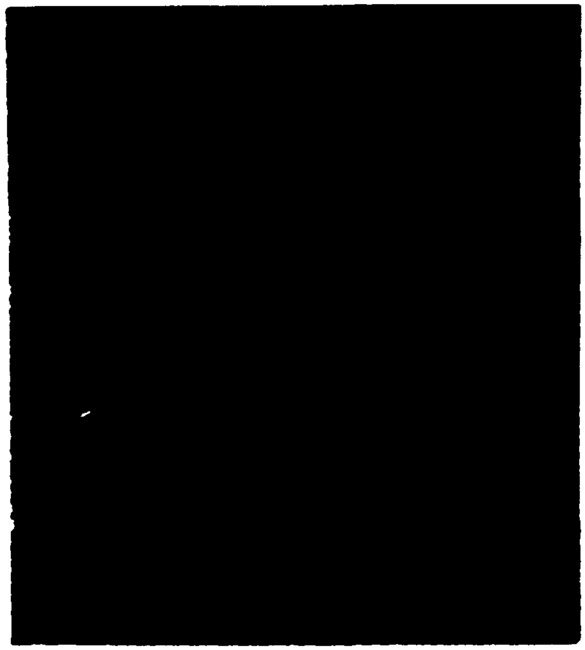


1c. 620°C X 30 hours SR (X500)

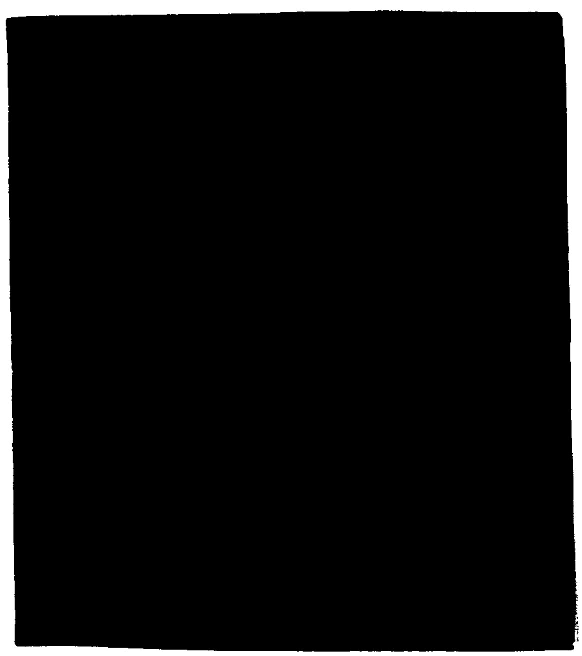


1d. 620°C X 300 hours SR (X500)

Micro. 1 continued

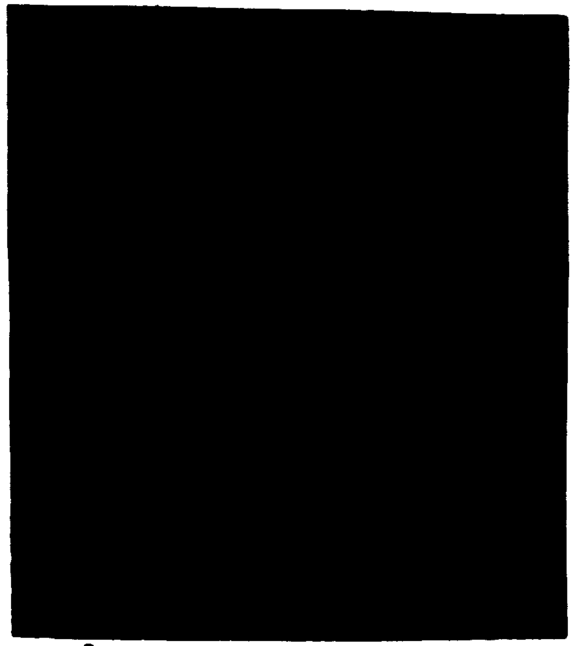


2a. As received (X100)

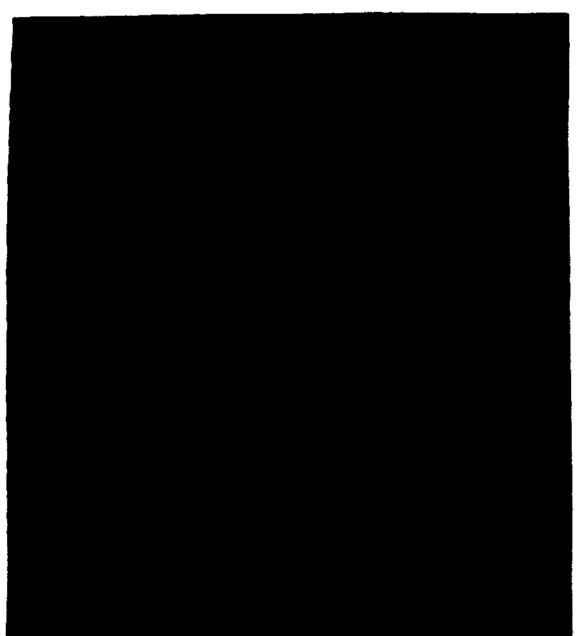


2b. As received (X500)

Micro. 2 Microstructures of A737 Grade C quenched and tempered plate



2c. 620°C X 30 hours SR (X500)



2d. 620°C X 300 hours SR (X500)

Micro. 2 continued

3a. As received (X80,000)

3b. 620°C X 300 hours SR (X46,000)

3c. 620°C X 300 hrs. SR (X13,000)

Micro. 3 Electron micrographs of A737 Grade C Normalized plate

A: Cementite
B: Compound precipitate
C: Vanadium precipitates

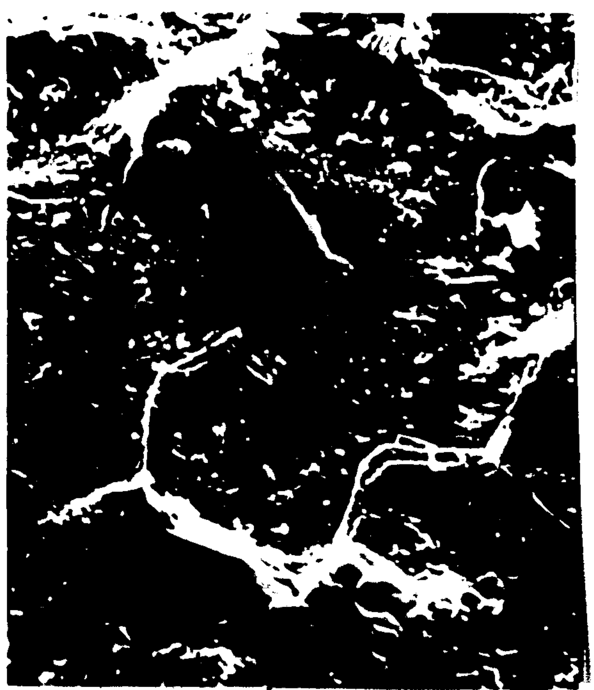
4a. As received (X80,000)

4b. 620°CX300 hrs. SR (X8,000)

4c. 620°CX300hrs. SR (X8,000)

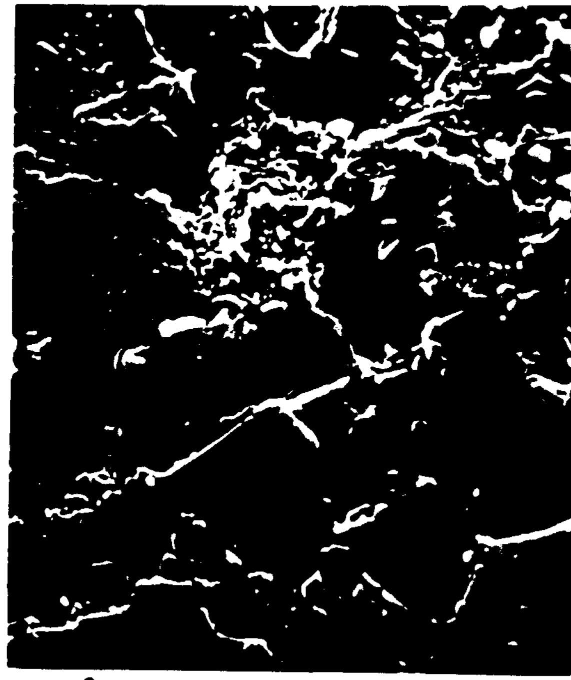
Micro. 4 Electron micrographs of A737 Grade C quenched and tempered plate

5a. As received (X3000)



5b. As received (X5000)

Micro. 5 Electron micrographs of A737 Grade C normalized plate by SEM



5c. 620°C X 300 hrs. SR (X3000)

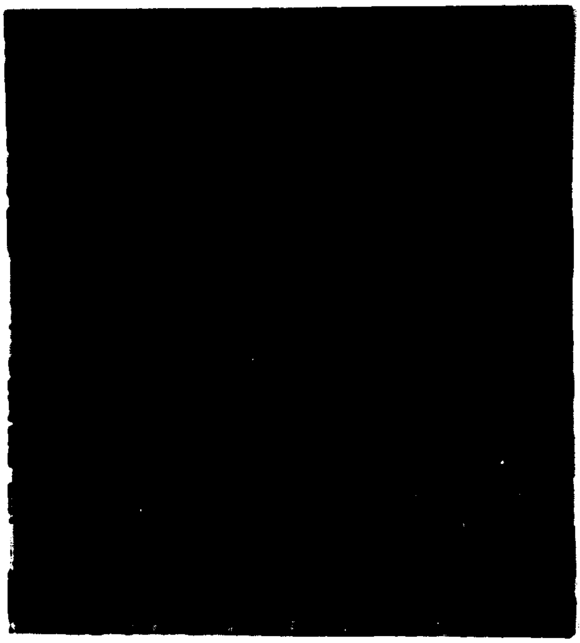
5d. 620°C X 300 Hrs. SR (X5000)

Micro. 5 continued

6a. As received (X5000)

6b. 620°C X 300 hrs. SR (X5000)

Micro. 6 Electron micrographs of A737 Grade C quenched and tempered plate by SEM

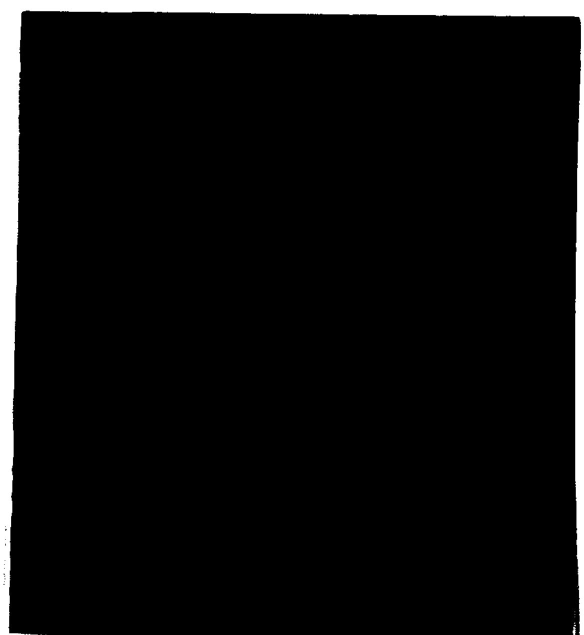


7a. 10% strain (X500)

7b. 10% strain + 260°C aging (X500)

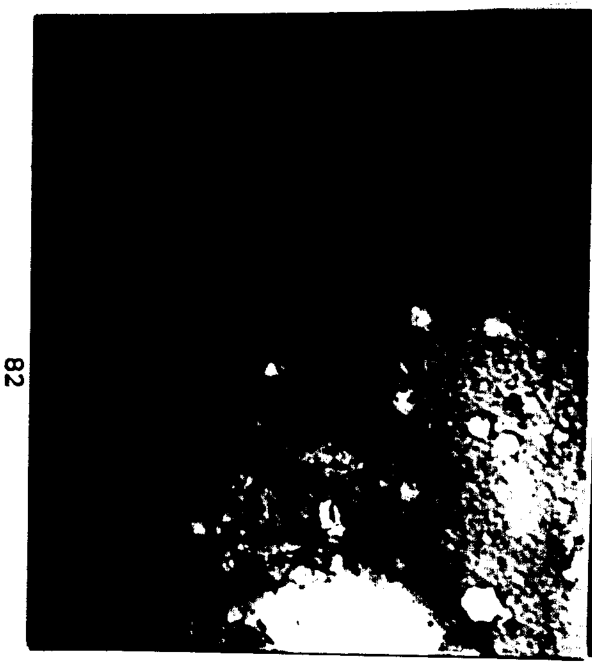
Micro. 7 Microstructures of A737 Grade C normalized plate

7c. 10% strain + 260°C aging + 620°C X 2 hrs. SR (X500)

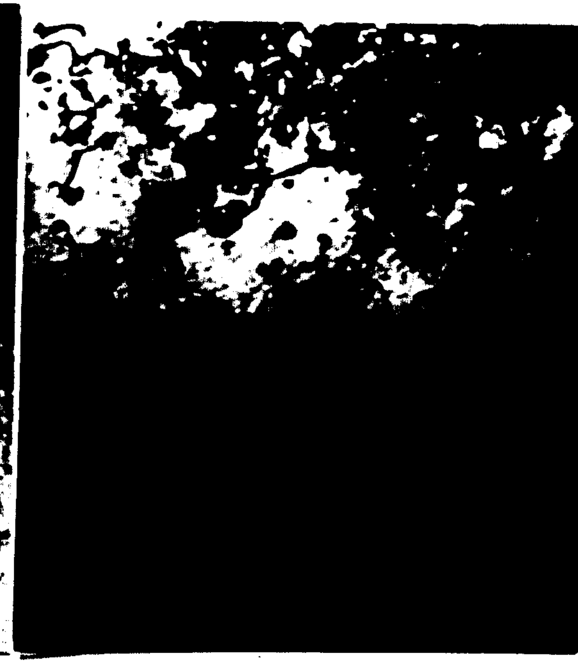


7d. 10% strain + 260°C aging + 620°C X 100 hrs. SR (X500)

Micro. 7 continued



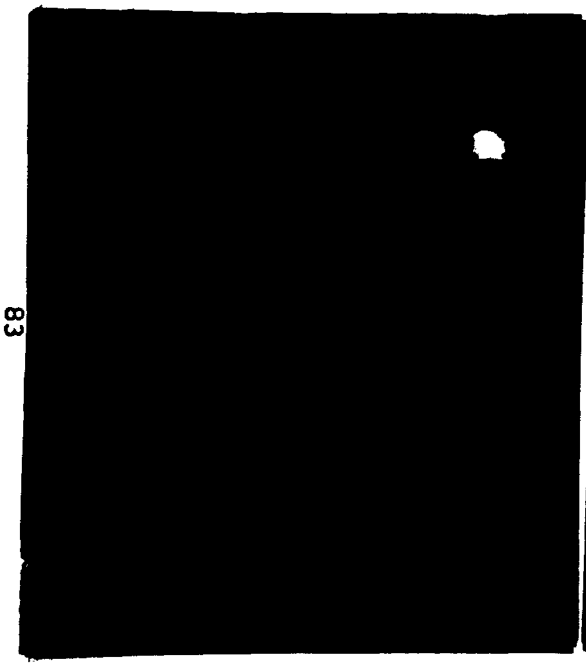
8a. 10% strain (X80,000)



8b. 10% strain + 260°C aging (X80,000)

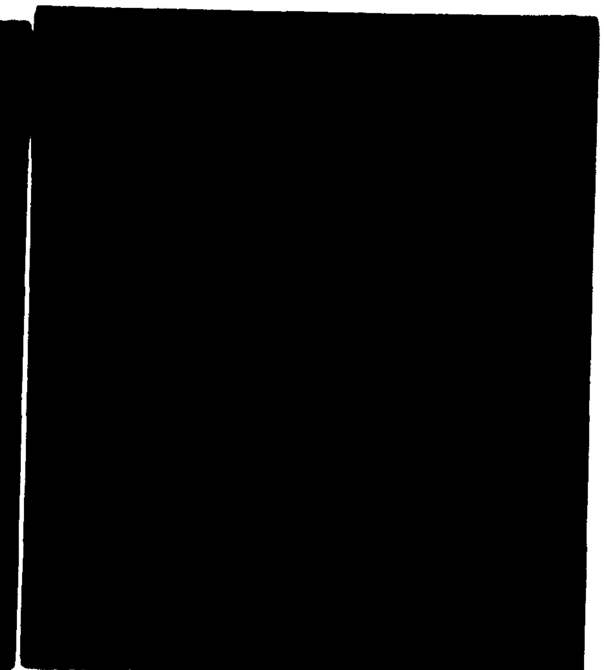
8c. 10% strain + 260°C aging (X170,000)

Micro. 8 Electron micrographs of A737 Grade C normalized plate



9a. (X80,000)

9b. Same area as 9a
DF image (X80,000)



9c. (X410,000)

Micro. 9 Electron micrographs of A737 Grade C normalized plate after 10% strain + 260°C aging + 620°C X 2hrs. stress relieving

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Biography

Makoto Sekizawa, the son of Mineo and Tokumi Sekizawa, was born in Matsumoto, Japan, on March 18, 1949. He received his Bachelor of Engineering degree from Tohoku University, Sendai, Japan, in 1971. He joined Kawasaki Steel Corporation in 1971. He is a member of the Iron and Steel Institute of Japan, and Japan Welding Society. His Master of Science degree in the Department of Metallurgy and Materials Engineering was conferred by Lehigh University, Bethlehem, Pa. 1981.