QUENCH CRACKING OF THE ALUMINIUM ALLOY 7010

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ABSTRACT

The Al-Zn-Mg-Cu aluminium alloy 7010 is a precipitation hardenable, plate and forging alloy mainly used for strength critical aerospace structural applications. 7010 attains its high strength via a high temperature (475°C) solution heat treatment followed by a rapid quench into water/organic quenchant and a subsequent artificial ageing treatment. The quenching part of this process sets up severe thermal gradients in thick sections, and the inhomogeneous plastic deformation results in residual stresses of yield point magnitude. Occasionally, fracture can occur during the quenching stage.

An investigation into the cause of fracture during quenching has been carried out. Residual stress measurements have been made on the surface of large rectilinear forgings and correlated with values predicted using finite element methods. In addition, elevated temperature plain and notched tensile test have confirmed a significant reduction in the tensile ductility in the temperature range 460-470°C. This loss of tensile ductility is considered the main reason for fracture observed during quenching.

KEYWORDS

7010 aluminium alloy, forging, quench cracking, residual stress.

INTRODUCTION

The aluminium alloy 7010 [1] is a descendant of the Al-Zn-Mg-Cu alloys first used towards the end of Second World War. It is a precipitation hardenable alloy developed in the late 1970s by Alcan International Ltd. (UK) and HDA Forgings Ltd. (UK) under the sponsorship of the UK Ministry of Defence. The alloy was primarily designed as a plate and forging alloy with reduced quench sensitivity allowing for use in thick sections. The alloy has a low combined Fe+Si impurity content for good fracture toughness. The combination of strength, fracture toughness and stress corrosion cracking resistance was improved compared to alloys like 7075T651 and 7079T6. [2] It is mainly used for strength critical aerospace structural applications.

To produce useful strengthening, precipitation hardenable aluminium alloys rely on rapid quenching from the solution heat treatment temperature to suppress the formation of coarse equilibrium second phases. Precipitation nucleation and growth kinetics dictate that the critical temperature range is between 400 and 290°C and the cooling rate through this range must exceed 100°C sec^{-1} for most alloys, although certain chromium containing quench sensitive alloys like 7075 require up to 300°C sec^{-1}. [3] Quenching is normally performed by submerging the material into cold water, or when lower rates of cooling are required, hot water or aqueous solutions of organic quenchants like polyalkylene glycol (PAG) that are inversely soluble in water with respect to temperature. The resulting metastable supersaturated solid solution can then be subject to a controlled decomposition known as aging.

Quenching into cold water by immersion or spraying produces the greatest possible thermal gradients in aluminium alloys and is an ideal quenchant from a mechanical properties perspective. Unfortunately, the severe thermal gradients can result in inhomogeneous plastic deformation occurring as the material passes through the 450-300°C temperature range. [4] In thick components like large forgings, this results in the introduction of surface compressive residual macrostresses of yield point magnitude balanced by tensile sub-surface macrostresses. Reducing the thermal gradients by using heated water and PAG type solutions reduces residual stresses at the expense of aging response, the degree being dependent on the alloy. Residual stresses can cause both warping during machining and dimensional instability and established procedures exist to reduce if not eliminate residual stresses from semi-finished products. The strengthening mechanism of age hardenable aluminium alloys obviously precludes the application of normal thermal methods to relieve the residual stresses induced by quenching and stress relieving heat treatable aluminium alloys is therefore usually performed using mechanical methods.

Quench cracking during the heat treatment of precipitation hardenable aluminium alloys is a technological problem sometimes encountered when thick plate or rectilinear forgings are subject to aggressive quenching procedures like cold water spray or immersion. The severe thermal gradients result in inhomogeneous tensile plastic deformation occurring in the surface regions, which in turn can cause crack initiation and rapid propagation. The phenomenon is not widely reported. The problem is intermittent is not associated with any obvious process variable. A possible candidate for the site of crack initiation is pre existing surface forging defects. When as forged surfaces are heat-treated the problem is more common and instances of up to 70-80% of specific parts affected by splitting have been observed. Machining or sawing of as forged surfaces, reduces the problem but does not eliminate it. Cracking has been observed to occur from machined surfaces and even surfaces that were originally within the forging prior to sectioning before heat treatment. This suggests that the phenomena of quench cracking cannot be solely due to pre-existing forging defects.

This paper quantifies the residual stresses generated during cold quenching of large rectilinear aerospace 7010 forgings and attempts to determine the likely reasons for the initiation and propagation of large cracks during heat treatment.

EXPERIMENTAL

Material details

A 124(ST)x156(LT)x3041(L)mm rectilinear forging was supplied by HDA Forgings Ltd., Redditch, UK. This forging was supplied in the W52 condition; solution heat-treated at 475°C, quenched into water at T<40°C, and then cold compressed 2¼%. Large quench cracks had been detected by the normal rigorous non-destructive testing procedures employed, resulting in the forging being rejected. The cracks had initiated at the surfaces and propagated in the L-T plane (initiated and propagated normal to the ST direction). The surfaces adjacent to cracks were examined for sites of possible crack initiation but the surface roughness near cracks was no different to other areas on the forging. The crack depths varied from 15-35mm.

The specification composition and quantitative analysis of a sample from the forging performed using standard analytical techniques is presented in Table 1.

Table 1 Chemical composition corresponding to the 7010 aluminium alloy specification and chemical analysis results (in brackets), wt%.

	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Zr
7010 **Actual**	0.12 (0.03)	0.15 (0.058)	1.5-2.0 (1.695)	0.10	2.1-2.6 (2.44)	0.05	5.7-6.7 (6.26)	0.06	0.10-0.16 (0.14)

Residual stress prediction.
A finite element program (ABAQUS V5.5) was used to model cooling and residual stress formation during quenching from 475°C to room temperature for full scale and half scale models of the forged block. Temperature dependent values of thermal conductivity [5], specific heat capacity [6], density [7] and thermal expansion [8] were all taken from the literature. The material properties were assumed to be isotropic, with strain rate dependency taken into account and hardening assumed to be linear after initial yield. The plasticity parameters used for the material during the quench were extrapolated from values measured at higher strain rates.

Elevated temperature tensile testing.
Short transverse tensile test pieces were machined from a solution heat-treated forging. Both parallel (D~5.64mm, L~32mm), and notched specimens (notch diameter~5.64mm, notch radius ~0.15mm, 60° angle, parallel diameter 7.0mm), were tested at temperatures from 430 to 500°C. Specimens were heated in situ and allowed to heat to the temperature of testing. Upon attaining the test temperature, the samples were strained to failure with no soak period. Samples were then cooled to room temperature.

Scanning electron microscopy
SEM and qualitative analysis was performed on a Jeol JSM840 scanning electron microscope with a LINK systems AN10000 energy dispersive x-ray facility attached.

OBSERVATIONS

Residual stress prediction
To estimate the thermal and residual stress levels induced by the quenching stage of solution heat treatment, a block corresponding to the same cross section but with reduced length was modelled using ABAQUS. Two conditions were considered to determine the size effect on residual stress. Initially a block of dimension 124(ST)x156(LT)x550(L)mm was modelled and then a half size block of dimension 62(ST)x78(LT)x225(L)mm.

Full size block 124(ST)x156(LT)x550(L)mm. The predicted temperatures as a function of time at three surface locations on a cross section of this block are shown in Figure 1. As expected, the cooling rate of the corner location was much greater than the mid points of the two other faces. It will be noted that the mid point of the face which cracked (L-S) cooled to ~220°C in ~3 seconds after immersion in water at 20°C.

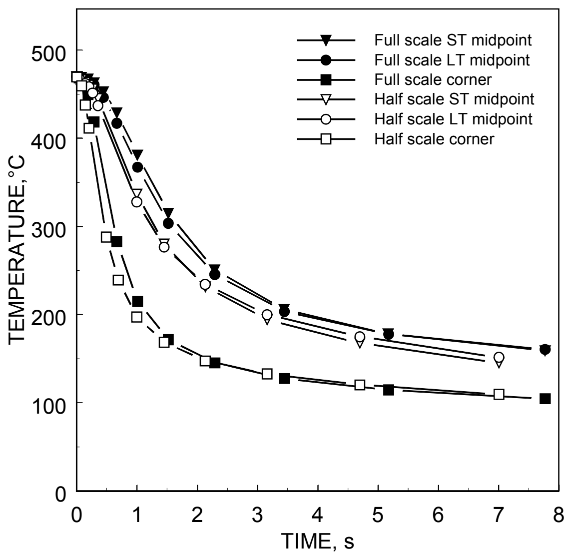

Figure 1. FEA predicted cooling curves for nodes on the surface of a full and half scale block during quenching from 475°C.

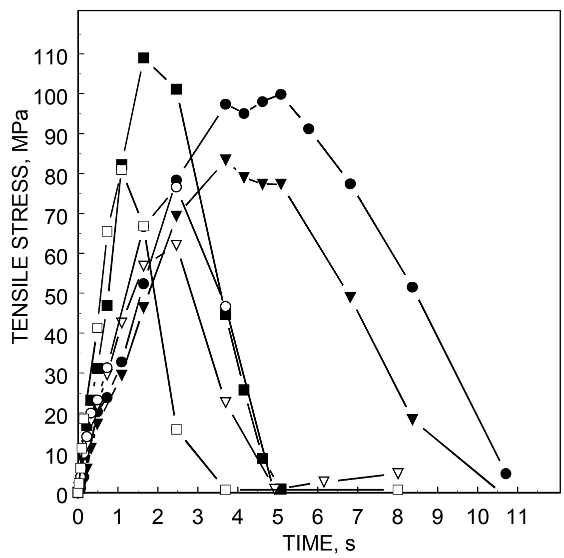

Figure 2. Max. tensile stress at midpoints and corner of full and half scale block during initial stages of quench. Legend as Figure 1.

The maximum tensile stresses developed during quenching are shown in Figure 2. The maximum tensile stress developed on the L-S face during the quench was predicted to be ~85MPa. This stress was developed after ~4 seconds when the material at the surface would have cooled to 200°C. The principal strain rates developed during cooling are indicated in Figure 3.

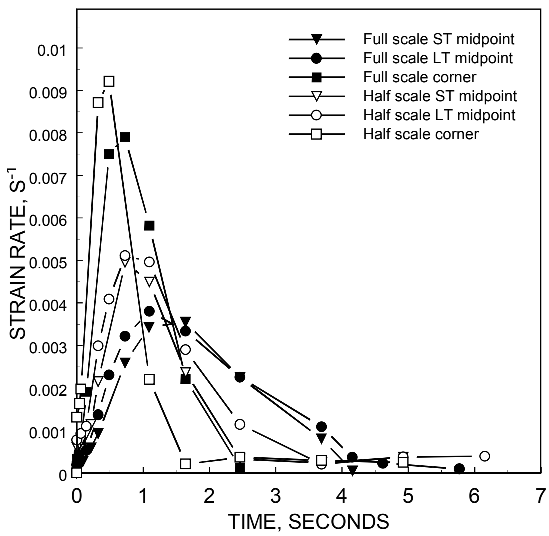

Figure 3. Principal tensile strain rates developed at the face midpoints and corner during quenching of the full and half scale models

The maximum strain rate on the L-S face was ~0.004s^{-1} after ~1 second. The tensile stress developed after this period was ~30MPa and the temperature of the midpoint of the L-S surface had fallen to ~380°C. The resulting residual stresses predicted for the midpoint of the L-S face was ~290MPa.

Half size block 62(ST)x78(LT)x225(L)mm. The predicted temperatures as a function of time at three surface locations on a cross section of this block are also shown in Figure 1. As expected, the cooling rates at the surface and corners of this block were greater than the full-scale block. The maximum tensile stresses developed during quenching are also shown in Figure 2. The maximum tensile stress developed on the L-S face during the quench was predicted to be ~60MPa, 25MPa less than the full-scale block.

This stress was developed after ~2.5 seconds when the material at the surface would have cooled to ~215°C. The principal strain rates developed during cooling are also indicated in Figure 3. The maximum strain rate on the L-S face was ~$0.005 s^{-1}$ after ~0.8 seconds. The tensile stress developed after this period was ~25MPa and the temperature of the midpoint of the L-S surface had fallen to ~350°C. The resulting residual stresses predicted for the midpoint of the L-S face was also ~290MPa. The effect of size on the development of stresses during quenching is complex but the significant observation is that the magnitude of the tensile stresses developed in the surface increase as the block size increases. Plastic deformation will accompany these stresses, as the yield stress will be very low at the initial high temperatures. The amount of plastic strain will increase as the block size increases. If the thermal strains cannot be accommodated by plastic deformation, the material will fracture.

Elevated temperature tensile testing.
It was known that Al-alloys can exhibit low values of tensile ductility when subject to stresses at elevated temperatures. This is normally associated with a change in the failure mode from a shear type gross plastic deformation mechanism to an intergranular separation type mechanism. The temperature at which this change in mechanism takes place is sometimes referred to the equicohesive temperature. Ductility measurements on standard smooth tensile specimens do not always reveal metallurgical or environmental changes that lead to reduced local ductility. The triaxial state of stress at a notch will make plastic flow more difficult, and is a better model of the real (biaxial) state of stress in the surface of the blocks. To determine if 7010 had low elevated temperature ductility, a series of plain and notched elevated temperature tensiles was conducted at temperatures up to 500°C. The strain rate used in the tests was either 0.00167 or $0.01 s^{-1}$ with the majority of tests being performed at the higher value.

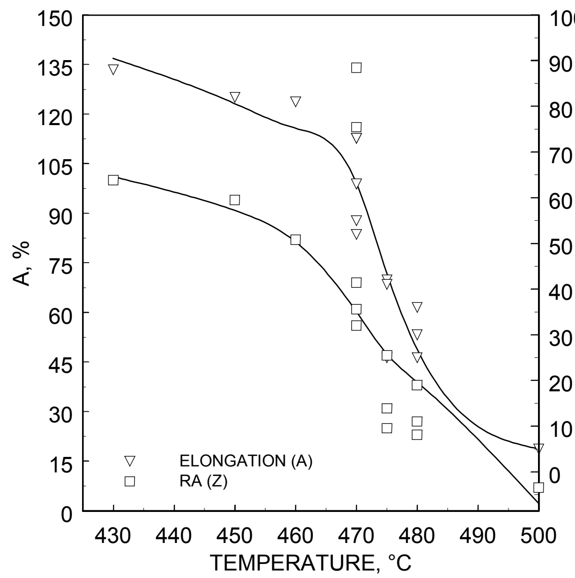

Figure 4 Plain tensile results, elongation (A) and reduction of area (Z).

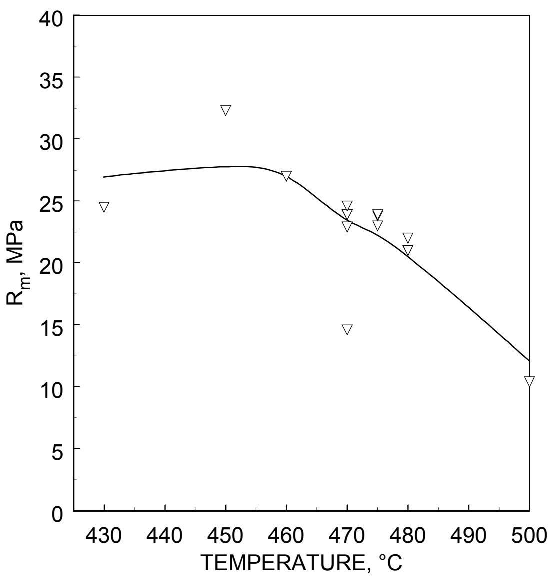

Figure 5 Plain tensile results, tensile strengths (R_m).

Plain elevated temperature tensile results. The observed ductility and reduction of area was noted to decrease with increasing temperature, Figure 4. At temperatures up to 480°C elongations greater than ~38% were recorded. At 500°C the material was macroscopically brittle with a recorded elongation of only 7%. As the solidus of 7010 is known to occur around 488°C then liquation of low melting point phases will account for this low value. Five tests were conducted at 470°C and the measured elongation varied from 56-134%. This appeared to suggest that 7010 was still very ductile at the solution treatment temperature when subject to a uniaxial stress. The tensile strengths were also noted to decrease with increasing temperature, Figure 5.

Notched elevated temperature tensile results. The observed ductility was noted to decrease with increasing temperature, Figure 6. At temperatures up to 460°C elongations greater than 20% were recorded. However, at 470°C the material became macroscopically brittle with a recorded elongation of less than 5%. Four tests were conducted at 470°C and the measured elongation varied from 3.7-4.8%. The tensile strengths recorded were greater than the equivalent plain test, Figure 7. This is normal; a consequence of reduced shear stresses at the notch making plastic flow more difficult. Figure 6 also compares the ductility of plain with notched samples. At 470°C it can be concluded that 7010 becomes notch sensitive.

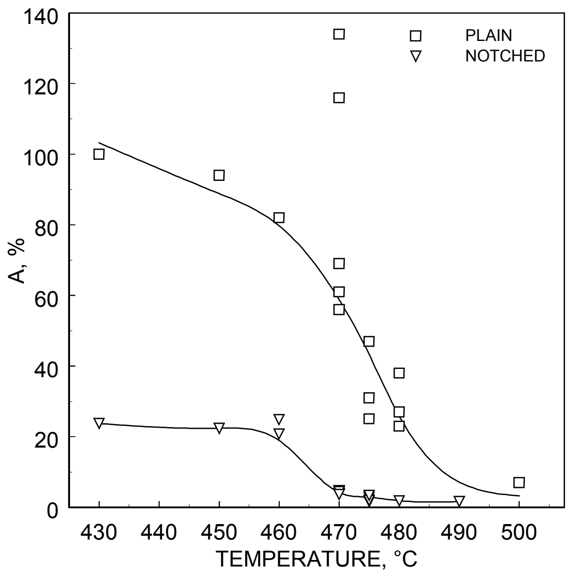

Figure 6 Plain and notched tensile results, elongation (A).

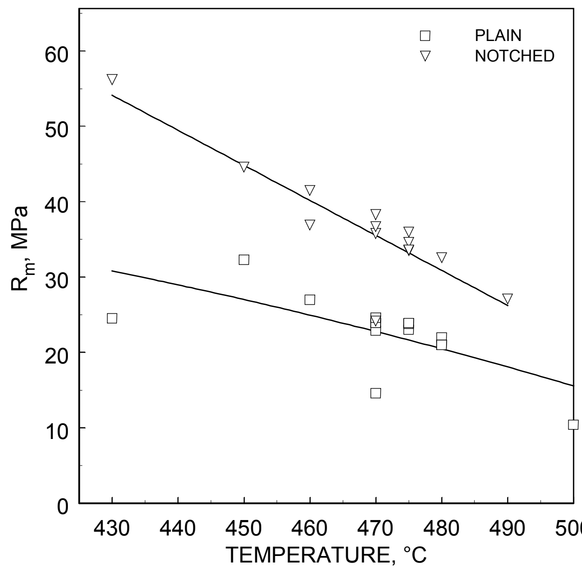

Figure 7 Plain and notched tensile results, tensile strength (R_m).

Fractography

Plain elevated temperature tensile specimens tested between 430 and 470°C exhibited extensive necking and cavitation towards the necked end. Transgranular shear was the dominant failure mechanism. As the temperature of testing increased evidence of intergranular failure became more apparent. This coincided with the reduction in elongation to failure. At 470°C samples that displayed high elongations had mixed mode fracture surfaces but at 475°C and above all fractures were intergranular.

Longitudinal metallographic sections confirmed the presence of extensive precipitation of second phase in grain interiors and boundaries. Clusters of coarse β(AlCuFe) constituent particles were observed the fracture surfaces. Figure 8 is a secondary electron image of a fracture surface from a plain tensile test piece fractured at 500°C. Extensive grain boundary precipitation of a Al-Cu-Zn rich phase can be seen. When a split from the forging was examined the fracture was obviously intergranular with little evidence of plasticity, Figure 9. The characteristics of this fracture were noted to be very similar to those observed on samples known to have failed by stress corrosion cracking.

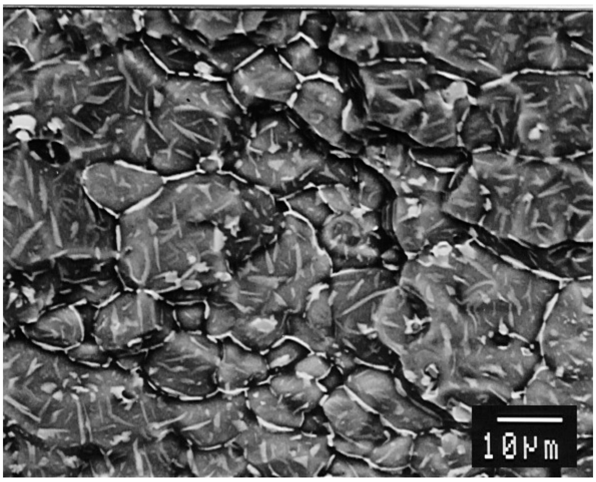

Figure 8 Fracture surface of a plain elevated temperature tensile specimen tested at 500°C.

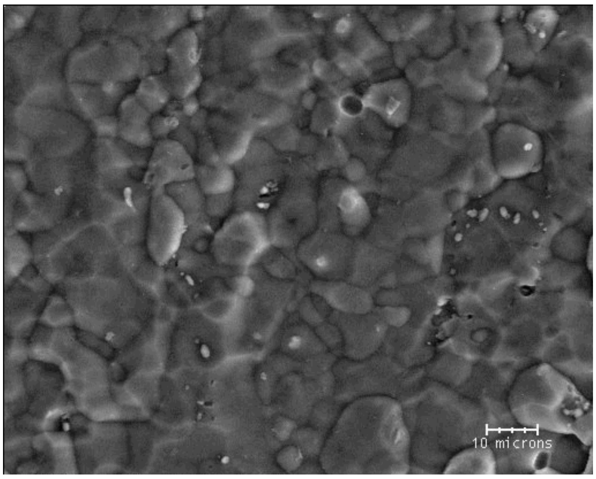

Figure 9 An intergranular fracture surface typical of the fracture surfaces found in the forging.

DISCUSSION

When this project was initiated, two mechanisms were suggested as possible causes of crack initiation and propagation. Simple tensile failure due to thermal contraction of the surface during quenching and the constraint effect of the underlying material, or a stress corrosion cracking mechanism after quenching. The residual stress measurements confirm what would be expected intuitively, that the surface stresses are biaxial, compressive and close to the initial yield point of the material. It would not be expected that a stress corrosion crack could self-initiate from this compressively stressed surface. A relatively large pre-existing defect would be a prerequisite for crack propagation by a stress corrosion cracking mechanism. However, it should be noted that the stress corrosion cracking resistance of 7010 in the W condition is very poor [9] like most other 7xxx series alloys.[10] If a crack were present for whatever reason and the tip of the crack had penetrated sufficiently to be subject to a tensile stress, then crack propagation would occur. The splits examined in this investigation did not look like a stress corrosion crack. No evidence of bifurcation of the split was detected.

If simple tensile failure is the more likely mechanism, due to the notch sensitivity at elevated temperatures, then the obvious question is “Why is cracking only manifest on large forgings?” The reason for this is the large constraint effect of thick forgings increasing the magnitude of the tensile stresses developed at the surface of the forging during the quench. In small forgings these thermal stresses can be wholly accommodated by plastic flow in the outer skin. In large forgings, the thermal stresses cannot be accommodated and in conjunction with the low toughness properties in the S-T direction and the notch sensitivity, the material fractures. It is likely that these fractures can then propagate by a stress corrosion cracking mechanism over a period of days/weeks if a suitable moisture laden atmosphere is present.

This investigation did not determine the root cause of the splitting. There is some recent evidence that the Fe:Si ratio can cause susceptibility to the phenomenon. A technological solution to the problem could be to lower the temperature of the parts prior to quenching. This would rely on a complete understanding of the time temperature property relationships for the alloy in question to ensure that the mechanical properties were not adversely affected.[11]

CONCLUSIONS

1. Modelling of the thermal stress generation during quenching predicts large (85MPa) tensile stresses on the L-S surface of the forgings within a few seconds of immersion.
2. These tensile stresses increase in magnitude as the block size increases due to increasing constraint and increasing thermal inhomogeneity.
3. Notched elevated temperature tensile tests confirm that 7010 becomes notch sensitive as the temperature increases from 460 (elongation~20%) to 470°C (elongation<5%).
4. Extensive grain boundary precipitation is observed in the fracture surface of elevated temperature tensile tests. It is likely these particles are equilibrium phases formed during the heating up phase of the test when insufficient time was allowed to solutionise.
5. The most probable cause of the longitudinal splitting is cracking due to low tensile ductility in the 460-470°C temperature range.
6. A possible solution to the problem is to lower the temperature of the solution heat treatment furnace just prior to quenching. This can only be carried out with reference to the time temperature property curve , and if the relevant material specification allows it.

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