The Magnitude of Heat Treatment Induced Residual Stresses and the Thermal Stress Relief of Aluminium Alloys

J.S. Robinson and D.A. Tanner

Department of Materials Science and Technology, University of Limerick, Ireland.

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Abstract. 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. An unavoidable consequence of the rapid quenching of thick sections is the severe thermal gradients that quickly develop in the material. The attendant inhomogeneous plastic flow can then result in the establishment of residual stresses. Established procedures exist to minimise residual stress by quenching into boiling water or organic quenchants at the expense of ageing response. Residual stresses can also be relieved after solution heat treatment by the application of plastic deformation in a controlled manner. A limited degree of thermal stress relief is also reported to occur during subsequent artificial ageing treatments, especially duplex ageing treatments. It is generally accepted that the size of the residual stresses induced during quenching cannot exceed the yield strength of the material. However, for precipitation hardened aluminium alloys, stress magnitudes as measured by standard techniques can exceed the uniaxial stress required to cause plastic flow during tensile tests conducted immediately after quenching. An investigation to explain these observations involving measuring as-quenched tensile properties and room temperature stress relief in heat treatable and non-heat treatable aluminium alloys has been conducted. Two alloys were investigated: 7010, an Al-Zn-Mg-Cu precipitation hardenable alloy and 5251, a non-heat treatable medium strength Al-Mg-Mn alloy. Tensile properties were determined by heat-treating test specimens at 475°C, cold water quenching and then testing without delay to avoid significant microstructural modification. The progress of stress relief at room temperature was then monitored utilising test coupons and standard x-ray diffraction techniques. Natural ageing of 7010 leads to a rapid increase in strength and a subsequent locking in of residual stresses (this cannot occur in 5251) and the change in residual stress is monitored as a function of time. An attempt has also been made to determine the efficacy of x-ray diffraction to monitor thermally induced stress relief. X-ray diffraction and hole drilling techniques to ASTM E837 were utilised to follow the progress of isothermal stress relief at room temperature and 200°C in both 7010 and 5251.

Introduction.

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 needs to exceed 100°C sec^{-1} for most alloys. 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 such as 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 ageing.

Quenching into cold water by immersion or spraying produces the greatest possible thermal gradients in aluminium alloys for a given section and is an ideal quenchant from a mechanical properties perspective. Unfortunately, severe thermal gradients can result in inhomogeneous plastic deformation occurring as the material passes through the 450-300°C temperature range.[1] In products such as large forgings and plate, this results in the introduction of surface compressive residual macrostresses balanced by tensile sub-surface macrostresses, where the stress pattern is a

reflection of the geometry of the component and of the temperature gradients generated throughout. Several researchers have indicated that surface compressive stresses in cold water quenched plate and forging alloys can have magnitudes >200MPa using the layer removal technique.[2-5] A more recent paper using the compliance technique indicates subsurface stress magnitudes >200MPa while surface stresses were approximately 150MPa.[6]

Reducing the thermal gradients by using heated water and PAG type solutions reduces residual stresses at the expense of ageing 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 relief of heat treatable aluminium alloys is usually performed using mechanical methods.[5, 7-10]

The mechanism by which residual stresses are introduced during quenching implies that these stresses cannot exceed the yield stress of the material. As the flow stress of aluminium alloys decrease with increasing temperature the maximum residual stress should bear some relation to the strength properties measured at room temperature.[11] The biaxial nature of surface residual stresses may increase the magnitude above the measured uniaxial tensile properties but only by a small degree. It would therefore be expected that the maximum as-quenched residual stress should be similar in magnitude to the uniaxial tensile properties measured at room temperature.

New retrogression and reageing heat treatments that can be applied to certain 7XXX series alloys to improve the combination of mechanical properties utilise short durations at intermediate temperatures (200-240°C).[12] There is some evidence that limited stress relief occurs during normal ageing treatments,[4, 13] and the prospect of greater stress relief occurring during the retrogression treatment has been considered in this paper. X-ray diffraction is used to detect changes in the surface residual stress of materials subject to temperatures where glide type creep mechanisms can occur, and as such is used to monitor the progress of mechanically induced residual stress relaxation although this technique is not always infallible.[14, 15]

Experimental.

Materials and heat treatment. 7010 specimens for tensile testing and residual stress measurement were cut from a large rectilinear open die forging. 5251 specimens were cut from 16mm thick rolled plate. The registered compositions of these two alloys are presented in Table 1. All heat treatments were conducted in an air-circulating furnace with temperature control of ±2°C. Quenching was performed by plunging specimens into a large volume water tank utilising manual agitation.

Table 1 Chemical composition corresponding to Al alloy specifications, wt%.

Alloy	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Zr	Al
7010	0.12 max	0.15 max	1.5-2.0	0.10 max	2.1-2.6	0.05 max	5.7-6.7	0.06 max	0.10-0.16	Bal.
5251	0.40 max	0.50 max	0.15 max	0.10-0.50	1.7-2.4	0.15 max	0.15 max	0.15 max	-	Bal.

Tensile testing. Tensile testing of the test pieces was performed in accordance with ASTM B557-84[16] using a non standard round LT tension test piece geometry of gauge length 30mm, on a Dartec 500kN servo-hydraulic load frame utilising a 25mm gauge length extensometer. The test piece diameter was 6mm. Samples were tested at a strain rate of $3x10^{-4}$ s^{-1}. The minimum delay between quenching and testing was 3 minutes. The duration of the tensile test was approximately 20 minutes but the $R_{p0.2}$ was achieved within the second minute of testing.

X-ray diffraction. X-ray diffraction residual stress measurements were taken at the centre of one 60*60mm face (corresponding to the LT-L plane) in the L or LT direction using a Philips X'Pert x-ray diffractometer for both 5251 and 7010 aluminium alloy samples (16mm thick). Scan parameters were controlled using Philips X'Pert Data Collector (V1.2a) software with 2θ (2θ -

angle between source and diffracted x-ray beam) values chosen to encompass the Cu-Kα doublet for the {422} planes: 136°<2θ<139°.

Eight scans were performed for each measurement using different ψ values within the range 0≤ψ≤60° (ψ – angle between the surface normal and the bisector of source and diffracted x-ray beam). Only positive tilting was used as negative tilting can lead to poor quality peaks when using the ω (omega) diffraction geometry. Pseudo-negative tilting (rotating the sample through 180° around the ϕ (phi) axis) did not affect the results achieved.

The resulting spectra were analysed using Philips PC-Stress Software (version 2.61) with peak locations determined with a parabola fitting technique. In cases where scans resulted in diffracted intensity peaks less than 400 counts, these scans were ruled out and in all cases a minimum of 6 scans were used to calculate the curve fit in the d (lattice spacing) versus $Sin^2\psi$ plots. Combined with this, where the standard deviation in the final results was greater than 15MPa, these results were repeated with 16 scans between the range 0≤ψ≤60° to reduce the scatter and improve the curve fit and allow scans with peaks below 400 counts to be removed. The elastic constant values used were taken from literature for the {422} planes.[17]

The validity of the x-ray diffraction technique described above has been verified in as yet unpublished work, through a round-robin exercise coordinated by the National Physical Laboratory (NPL), UK.[18] In that exercise, measurements were recorded on a sample of solution heat treated and cold water quenched 7010 plate and compared with x-ray measurements taken at other laboratories throughout the UK and Ireland. The results recorded at the University of Limerick compared well with the other investigators and the residual stress was reported by all the laboratories to be compressive in the range 150-220MPa.

Hole drilling residual stress measurement. Residual stresses were measured by a hole drilling strain gauge method as detailed in ASTM E837-95[19] at the centre of a 60*60mm (LT-L) face in each of the samples measured. The strain gauge rosettes (type CEA-06-062UL-120) were attached by Measurements Group, UK Limited. The Measurements Group RS200 milling guide and assembly was used for introducing the hole through the strain gauge rosette as detailed in the procedure provided by Measurements Group.[20] An orbiting technique was used to introduce a hole of diameter 1.88mm, resulting in a hole diameter to mean gauge diameter ratio of approx. 0.37 – within the parameters recommended by ASTM E837-95. The hole was drilled to a depth of 2mm.

Results.

Tensile properties. The as quenched and naturally aged tensile properties of 7010 are presented in Figure 1. The as quenched $R_{p0.2}$ of 7010 was of the order of 150MPa with a measured proportional limit of 110MPa. This alloy strengthened significantly after an initial incubation period of approximately 35 minutes. The alloy continued to strengthen over time and did not stabilise for the test duration (1530 hours). The properties of 7010 (in the W52 condition) naturally aged for 5 years are displayed in this figure as horizontal reference lines. 5251 is not a heat treatable alloy and does not undergo any strengthening phase transformation after quenching and "natural ageing". The as quenched $R_{p0.2}$ of this alloy was approximately 55MPa with a measured proportional limit of approximately 35MPa. Testing further specimens after a delay of 790 hours confirmed that the properties of this alloy were stable. The room temperature strain hardening characteristics of 7010 are shown in Figure 2. The strain hardening exponent n and the corresponding strength coefficient K (MPa) were calculated by assuming the plastic portion of the true stress true strain curve up to the tensile strength could be described by a simple power law type equation, $\sigma = K\varepsilon^n$. The as-quenched room temperature strain-hardening characteristics of 7010 can be described as having low K with a corresponding high n ($\sigma = 670\varepsilon^{0.32}$). This resulted in an overall low strain hardening rate, but with increasing natural ageing, K increased and n decreased with a resulting increase in the strain hardening rate (after 1530 hours: $\sigma = 795\varepsilon^{0.17}$). These observations are consistent with established precipitation and strain hardening theory.[21] The stable strain hardening characteristics of the 5251 alloy can be described by $\sigma = 384\varepsilon^{0.34}$. This alloy did not strain harden as rapidly as 7010.

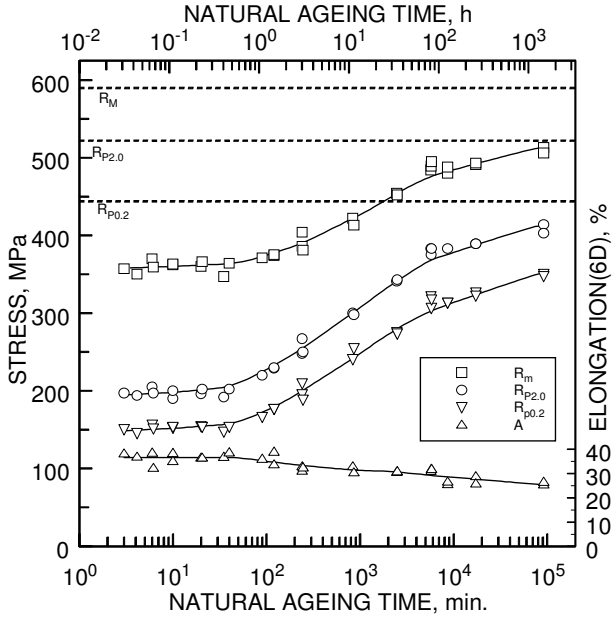

Figure 1 Tensile properties of naturally aged 7010. Horizontal reference lines are for 7010W52 after 5 years natural ageing.

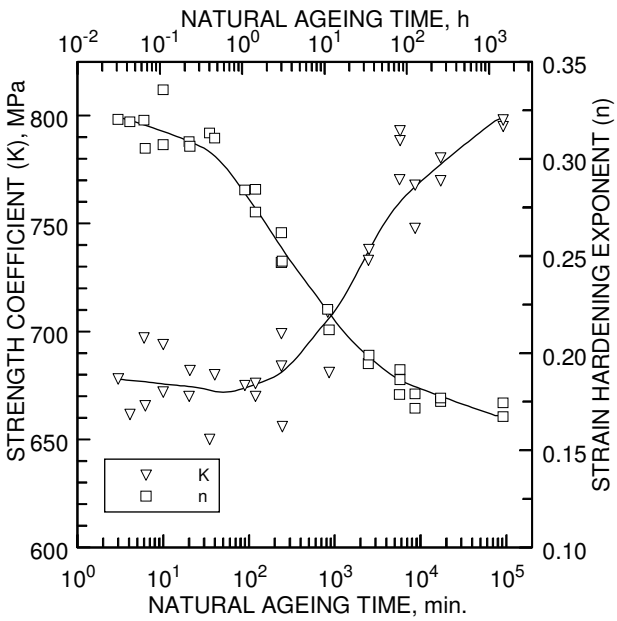

Figure 2 Room temperature strain hardening characteristics of 7010.

X-ray diffraction residual stress measurements. To determine the temperature from which 7010 can be quenched and still develop significant residual stress, a specimen was cold water quenched from successively increasing temperatures. Figure 3 confirms that significant residual stresses are only introduced when 7010 is cold water quenched from temperatures >200°C.

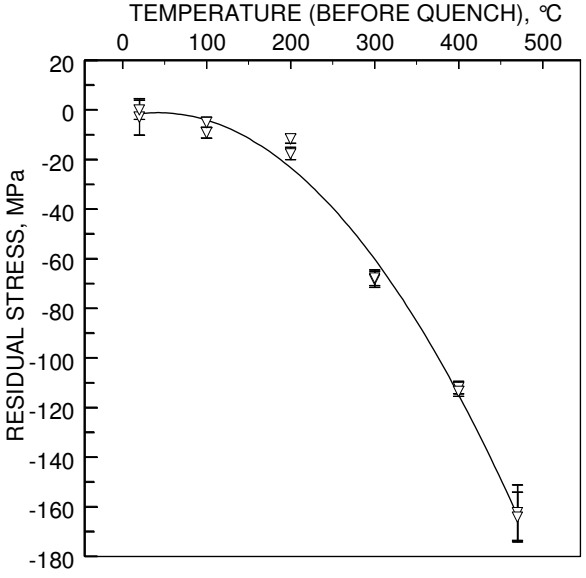

Figure 3 Residual stresses produced by quenching 7010 into cold water (<40°C).

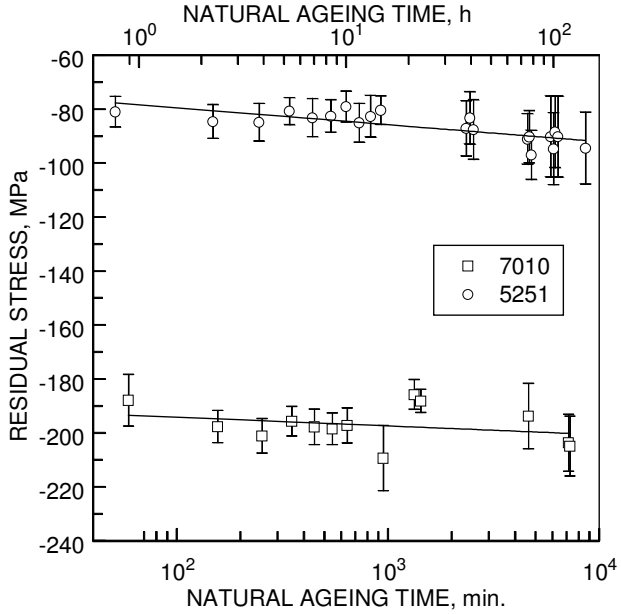

Figure 4 Residual stress versus natural ageing time for 7010 and 5251 samples.

Natural ageing. One sample of 7010 and one sample of 5251 was solution heat treated and quenched at the temperature used to solution heat treat 7010 (475±5°C). Immediately after the samples had been quenched in cold water (<40°C) the residual stress at the centre of one LT-L surface was measured using x-ray diffraction. The results are shown in Figure 4 where the time, after solution heat treatment, is plotted against the measured stress. The plotted error bars are standard deviation values calculated from the straight-line fit of the d versus $Sin^2\psi$ plots. The time plotted is calculated as the time from quenching to the mid-point of time during the residual stress measurement. As determined previously using x-ray diffraction [22], there was no decrease in the

residual stress magnitudes measured in the 7010 aluminium alloy, with the results indicating a slight increase in the magnitude, but this is thought to be due to experimental error or an as yet undetermined microstructural change. Similarly, in the 5251 alloy there was no detected decrease in the residual stress measured but also a slight increase. Further results indicated no decrease in stress magnitude for either alloy after over 720 hours natural ageing.

Ageing at 200°C. For the samples aged at 200°C the only sample to see a measured stress reduction using x-ray diffraction was the 7010 sample aged for 24 hours as can be seen from Table 2. The stress magnitudes obtained using hole drilling cannot be compared directly with the x-ray diffraction measurements as the stress was not uniform with hole depth. However, they can be used to indicate any stress reductions achieved from the ageing treatment. The stress magnitudes determined using the hole drilling technique indicate that the stress reduces in both materials when they are aged for 24hours at 200°C. A larger reduction was measured in the 7010 alloy than in the 5251 alloy.

Table 2 The effect of ageing at 200°C on residual stress magnitudes for 5251 and 7010.

Sample	Heat treatment	X-ray residual stress, MPa	X-ray software fit error, MPa	Hole drilling residual stress, MPa	
				σ_{min}	σ_{max}
5251B	SHT[1]	-66.1	±11.5		
5251B	SHT + 1h @ 200°C	-62.5	±5.9	-67	-52
5251C	SHT	-100.6	±6.0		
5251C	SHT +24h @ 200°C	-100.9	±8.8	-49	-42
7010X	SHT	-161.0	±9.6		
7010X	SHT + 1h @ 200°C	-159.3	±5.2	-152	-133
7010Y	SHT	-199.5	±16.5		
7010Y	SHT +24h @ 200°C	-169.5	±8.8	-90	-79

[1] SHT – 7010 type solution heat treatment at 475±5°C followed by quench into cold water (<40°C)

Discussion.

Subjecting the surface of a thick section aluminium alloy to a large thermal gradient produces inhomogeneous thermal stresses through the section. If the elastic thermal stress in the surface exceeds the local yield stress the material flows in tension. As the temperature difference between the surface and interior diminishes, the surface further contracts and is placed into a state of residual compression with the *yield stress at ambient temperature setting an upper limit*. The majority of investigations into the cold-water quench induced residual stress magnitudes of 7XXX series alloys place a range of values between 150-200MPa for this compressive surface stress. The x-ray diffraction measurements presented here are in agreement with these observations. The lower strength alloy 5251 has a correspondingly lower surface residual stress. However, when the as-quenched uniaxial *tensile* properties in these alloys are evaluated, plastic deformation initiates at stresses of smaller numerical value when compared to the measured residual stresses. The limit of proportionality in as-quenched 7010 as measured in this investigation was around 110MPa. The rate of uniaxial strain hardening in these alloys has been determined and using 7010 as an example, a plastic strain of 0.5% is required to raise the flow stress to 160MPa while 2.25% is required to increase the flow stress to 200MPa. Ignoring the predicted reversal of strain that occurs during quenching, strains of this magnitude should result in some dimensional changes in the surface of the specimen but none were detected.

The cold-water quenched tensile test specimens themselves will contain a residual stress distribution and this will result in plastic deformation occurring at a lower value of applied load than in their absence. To ascertain the magnitude of this effect, tensile samples of 5251 were heated to 475°C, cold water quenched or furnace cooled and then tested. No significant difference in the yield behaviour could be discerned. In addition to strain hardening, strain rate effects during the quench and crystallographic texture could also raise the local surface residual stress but without further investigation it cannot yet be stated with any certainty what causes this discrepancy.

Stress relief in these alloys at 200°C will occur primarily by a dislocation glide type mechanism. Initially, the movement of dislocations will not influence the inter-atomic spacing as exploited by x-ray diffraction and as such it would be expected that this technique could not detect this stress relief. The results presented here confirm that that this is the case and x-ray diffraction is less effective in detecting stress relief when compared to the hole drilling technique.

Conclusions.

1. The as-quenched $R_{p0.2}$ of 7010 and 5251 has been determined to be 150 and 55MPa respectively with corresponding limits of proportionality of 110 and 35MPa.
2. The as-quenched level of surface residual stress in coupons of 7010 and 5251 have been determined to be compressive in the range 160-200MPa and 66-100MPa respectively.
3. The as-quenched strain hardening characteristics of 7010 and 5251 have been determined. While strain hardening may contribute to the increased residual stress it is not thought that it can solely account for the high surface residual stress magnitudes.
4. The discrepancy between as quenched tensile properties and the as quenched residual stresses requires further investigation.
5. When monitored by x-ray diffraction, no reduction in the quench induced surface residual stress is detected in either 7010 or 5251 during natural ageing.
6. Initial stress relief at 200°C occurs by dislocation glide and as such is not detected by x-ray diffraction.

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References.

[1] J. A. Nicol, E. D. Seaton, G. W. Kuhlman, H. Yu and R. Pishko: Adv. Mater. Process., (1996), 149, p. S40.
[2] P. Jeanmart and J. Bouvaist: Mater. Sci. Technol., (1985), 1, p. 765.
[3] S. R. Yazdi, D. Retraint and J. Lu: J. Strain Anal. Eng. Des., (1998), 33, p. 449.
[4] N. Yoshihara, S. Tsuyama, H. Yoshimichi and K. Hirokami: (1992) In NKK Technical Review, p. 21.
[5] J. C. Boyer and M. Boivin: Mater. Sci. Technol., (1985), 1, p. 786.
[6] M. B. Prime and M. R. Hill: Scr. Mater., (2002), 46, p. 77.
[7] J. H. Hull and S. J. Erwin: (1972) In Metals Engineering Quarterly, p. 1.
[8] J. F. Faulkner:ALCOA stress relieved forgings. Aluminium company of America - Sales division, (1963).
[9] Y. Altschuler, T. Kaatz and B. Cina: (1988) In ASTM STP 993), p. 19.
[10] T. Bains: 1st International Non-Ferrous Processing and Technology Conference, St. Louis, Missouri, USA, (1997), (ASM International), p. 221.
[11] T. Ericsson: Adv in Sur Treat, Technol - Appl - Eff., (1987), 4, p. 87.
[12] J. S. Robinson, S. D. Whelan and R. L. Cudd: Mater. Sci. Technol., (1999), 15, p. 717.
[13] B. Cina, I. Kaatz and I. Elror: J. Mater. Sci., (1990), 25, p. 4101.
[14] A. A. Garcia-Granada, V. D. Lacarac, P. Holdway, D. J. Smith and M. J. Pavier: J. Eng. Mater. Technol.-Trans. ASME, (2001), 123, p. 125.
[15] D. J. Smith, G. H. Farrahi, W. X. Zhu and C. A. Mcmahon: Int. J. Fatigue, (2001), 23, p. 293.
[16] Annual Book of ASTM Standards, B557, American Society for Testing and Materials, (1992).
[17] V. Hauk: Adv in Sur Treat, Technol - Appl - Eff., (1987), 4, p. 251.
[18] X-ray diffraction focus group. Unpublished work.: (2002) National Physical Laboratory, UK.
[19] Annual Book of ASTM Standards, E837, American Society for Testing and Materials, (1992).
[20] Measurement of residual stresses by the hole drilling strain gage method, Tech. Note TN-503-4
[21] V. K. Vasudévan and R. D. Doherty: Acta Metallurgica, (1987), 35, p. 1193.
[22] D. A. Tanner, J. S. Robinson and R. L. Cudd: (2000) In Proceedings of the The Sixth International Conference on Residual Stresses ICRS-6, 10-12 July 2000, (Institute of Materials, Oxford, UK), p. 1299.