# Strengthening of a cold worked 17% Cr ferritic stainless steel (Type 430) by heat treatment

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### Synopsis

Slat-band chains, currently manufactured from Type 430 ferritic stainless steel, are used as conveyors in various industries. A process to strengthen these chains—with the aim of prolonging the lifetime of the conveyors—was developed recently. The process entails heat treating the cold-worked chain at 475°C, strengthening the material by precipitation of the Cr-rich phase $\alpha''$ at relatively long periods of aging. In the present work, it was found that increasing the interstitial content by solution heat treatments (at 880°C or 930°C) prior to cold rolling and aging, causes an increase in strength at shorter aging times as a result of the precipitation of chromium-rich carbides and nitrides. The solution treatments can introduce martensite into the microstructure if carried out at a sufficiently high temperature. This, along with the deformation caused by cold work, also increases the strength. Corrosion resistance is not affected by the treatments, but ductility is lowered. However, calculated critical crack lengths indicate that solution treatment at a temperature below the $Ac_1$ allows for a larger critical flaw size, and that the lifetime of the chains will probably be determined by fatigue. Although the duration of the service of the chains are reportedly controlled by wear, with these treatments the tensile strength of the material is increased to within acceptable limits for conveyor chains. The tensile strength is an indication of the wear or abrasion resistance of a material. It is expected that after the treatments, the wear resistance will be improved, but since embrittlement occurs, the fatigue resistance is lowered. It is concluded that the preferred processing route for the chains would be solution treatment at 880°C, fast quenching to trap the interstitial atoms, cold rolling and then aging at 475°C for 8 minutes.

### Introduction

Slat-band chains are used as conveyors by the food industry, breweries and bottling plants. These chains consist of flat cold-rolled and cold-formed links which are connected by pins passing through interlinking holes. The operating conditions require abrasion resistance and strength. One of the materials currently used for these chains is cold worked AISI Type 430 stainless steel, and it is used to the limit of its capabilities. By strengthening the chain it would

*a′ is the generally accepted symbol for martensite. Thus, to avoid confusion, the a″ notation is used to indicate chromium-rich precipitate, rather than a′ as is sometimes used.

be possible to prolong the life of the conveyors. Several strengthening possibilities exist, such as alloying of the current material, or replacement with a material having the desired properties. An unconventional process for strengthening the existing chains was developed by Mintek. The process entails heat treating the cold-worked chains between 450°C and 500°C, within the '475°C embrittlement' temperature range[1].

Embrittlement at, or near, 475°C occurs because of the formation of a Cr-rich phase, $\alpha''$, (*see footnote), with resulting increased hardness and lowered ductility. The $\alpha''$ phase forms because of the presence of a miscibility gap in the Fe-Cr system[2]. For the steel composition used in the current work (16 to 17 weight per cent Cr), $\alpha''$ forms by nucleation and growth, since the composition lies outside the spinodal[3-6].

In addition to the (slow) formation of $\alpha''$, the precipitation of carbides and nitrides may occur at shorter times during heat treatment in this temperature range.

The present work aims to propose, based on the Mintek heat treatment route, an optimum processing route to obtain a stronger chain and hence prolonged conveyor life. To allow an informed choice of processing route, attempts were made to elucidate the strengthening mechanisms.

### Material and experimental procedure

The material used in this investigation, in its initial form, was an 8 mm thick, hot rolled and batch annealed sheet of Type 430 ferritic stainless steel. The chemical composition is given in Table I. To minimise bias from solidification inhomogeneities in the sheet, it was cut into 140 smaller pieces (170x70x8 mm) and allocated randomly to three groups. Specimens were then procured as needed, sampling from

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one group until it was depleted. Thus it was assured that samples would be extracted randomly and from different positions in the sheet. At least two specimens, from different positions in the sheet, were used for each experiment.

Initial results revealed that with cold rolling and aging of the as-received sheet, increases in hardness could only be obtained after long aging periods (> 64 hours). Hence, it was decided to increase the interstitial content by additional solution heat treatments prior to cold rolling. Since the interstitial solubility increases with increasing temperature, the solution treatment is carried out at elevated temperatures. At sufficiently high temperatures, austenite appears in the microstructure. Because of the anticipated strong effect of austenite on hardness (the austenite transforming to martensite upon cooling), the $Ac_1$ temperature was determined experimentally.

This entailed solution heat treatment of 8 mm thick samples, at temperatures between 800°C and 1200°C, followed by water quenching, and then determining the amount of martensite in the structure, as well as the hardness (Figure 1). The volume percentage of martensite was determined by image analysis. Based on the results in Figure 1, the $Ac_1$ temperature was estimated to be 920°C. Hence two solution temperatures were selected, namely 880°C and 930°C, to test the aging behaviour of the material with and without martensite in the microstructure. While the solution treatment was applied here as a separate processing step, in-plant continuous annealing (followed by rapid cooling) would achieve the same end.
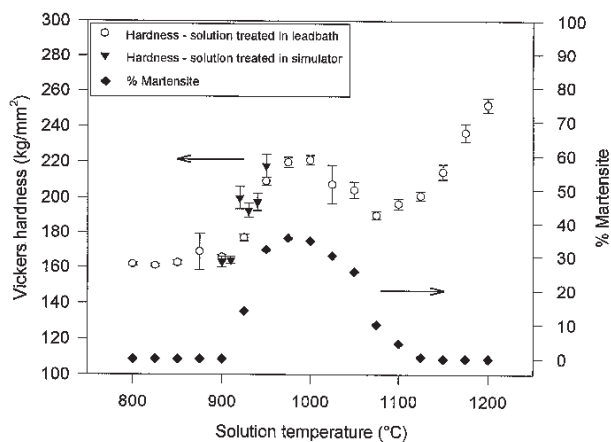


**Figure 1—Hardness and volume percentage martensite, after solution treatment at temperatures between 800°C and 1200°C, followed by quenching in water. As indicated, some solution treatments were performed by resistance heating in a weld simulator, which allowed closer temperature control. The increase in hardness corresponds with the formation of martensite, indicating that the Ac₁ temperature of the material is approximately 920°C**

The following sequence was used in preparing the test specimens:

➤ solution heat treatment (in air) at 880°C or 930°C (30 minutes at temperature) referred to as "solution treated"
➤ quenching in brine
➤ cold rolling at ambient temperature in the original rolling direction, giving a total of 38% reduction in thickness
➤ aging at 475°C, for periods ranging from 2 minutes to 2072 hours, in a lead bath. For aging times shorter than 0.5 hours, electrical resistance heating in a weld simulator was used, since this allowed more rapid attainment of the required temperature.

The effects of these treatments on the mechanical properties and corrosion resistance were investigated.

Hardness values were determined by using a Vickers diamond pyramid indentor, each value being the mean of twenty measurements. The specimens were ground to a 1200 grit finish before hardness testing, which used a load of 20 kg.

The corrosion resistance was established by potentio-dynamic testing in a 0.5M $H_2SO_4$ solution at room temperature. A scanning rate of 2 mV/s was used and an 123 mm² area was exposed to the solution. The solution was not deaerated. The specimens were wet ground on 1200 grit silicon carbide paper before exposure to the test solution.

Tensile tests were performed on specimens solution treated at 880°C and 930°C, cold rolled after quenching in brine, and aged at 475°C for either 8 minutes, 32 hours or 277 hours. Specimens which had not been aged or deformed were also tested. Elongation was measured using a 15 mm gauge length (the length between the shoulders of the specimen was approximately 28 mm, and the cross sectional area in this region 5x5 mm²). The gauge is not a standard length for the specific specimen cross section. However, the tensile behaviour was measured to compare the plastic flow characteristics. These are properly studied by comparing true stress–true strain results, which were only determined for stresses up to the tensile strength limit (i.e. the onset of necking). For these uniform elongations, the choice of gauge length does not affect the results. Force and elongation were converted to engineering stress and strain curves; which, in turn, were converted to true stress and true strain curves for strains up to the onset of necking. For the final graphs the elastic strain component ($\sigma/E$, with E = 210 GPa) was subtracted from the total true strain, to give plots of true stress vs. true plastic strain.

Impact tests were performed on sub-sized Charpy specimens of full width (10 mm wide, 5 mm thick, 55 mm long). A standard notch (45°, with a 0.25 mm radius) was cut into the 10 mm side. The material was solution treated at 880°C and 930°C respectively, cold rolled and aged for 8 minutes and 32 hours at 475°C, before the Charpy specimens were cut.

Transmission electron microscopy studies were performed in a 100 keV microscope. The foils were prepared to lie parallel to the original rolling plane. The cold-rolled material, before and after aging for 260 hours at 475°C, was investigated.

Mössbauer studies of selected specimens (with no

solution treatment, or solution treated at 880°C or 930°C, all cold rolled and aged at 475°C) were performed using a cobalt source. Tests ran for two days. Small shifts in the energies of transmitted gamma-rays are obtained in the Mössbauer apparatus as a result of the Doppler effect. This is achieved by moving the source and the sample relative to one another. The Mössbauer spectra are normally given as plots of absorption against relative velocity (in mm/s). The absorption characteristics are affected by the electronic environment of the nucleus, producing the hyperfine interactions. The principal hyperfine parameters derived from the Mössbauer spectrum are the hyperfine field, isomer shift and quadrupole splitting. In this instance the hyperfine field is of especial interest. This field arises from the splitting of the nuclear energy levels in a magnetic field producing transitions which are each seen as a line in the Mössbauer spectrum. The energy difference between the first and last lines is directly related to the effective magnetic field experienced by the nucleus[7]. This magnetic field is sensitive to the chromium content, and was hence used here to detect compositional changes in the alloy matrix.

## Results and discussion

### Hardness tests

Hardness tests (Figure 2) revealed two distinct stages in the aging behaviour of the solution treated specimens. The first is an increase at short periods of aging (around 8 minutes) and the second, an increase at long aging periods. The first increase in hardness is presumably associated with the precipitation of carbides and nitrides. The high interstitial content, due to the solution treatments, renders conditions for precipitation more favourable than in the material without prior solution treatment, where the hardness increased only after long periods of aging (more than 64 hours). In a separate test, where specimens were aged at 100°C for up to 14 hours, it was established that the initial increase in hardness is not the result of strain aging, since the aging was found to give no increase in hardness.

The precipitates that cause the initial strengthening could not be detected by transmission electron microscopy, probably due to the high amount of cold-work which would obscure any sign of the strain fields around the precipitates. The precipitates are thought to be little more than clusters of carbon or nitrogen[8]. The decrease in hardness with aging beyond 8 minutes appears to be an effect of excessive aging.

Mössbauer spectroscopy indicated that these initial precipitates contain more chromium than the average alloy composition. The increase in hyperfine field with aging (Figure 3) indicates that parts of the matrix became enriched in iron (with a resulting increase in magnetic field). The implication is that other regions of the alloy are enriched in chromium. If this is taken into account, as well as the fact that the solution treatments prior to cold rolling increase the interstitial carbon and nitrogen content of the material; and since the initial increase in magnetic field (for aging for up to 8 minutes) is much larger for solution treated specimens than for the untreated material (echoing the absence of an initial hardness increase in the untreated material) it can be concluded that the initial hardness increase is probably caused by the precipitation of chromium-rich carbides and nitrides.

The formation of chromium-rich precipitates may be expected to reduce the corrosion resistance, regardless of whether these precipitates are carbides or nitrides, or α″. However, no significant effect of aging on corrosion behaviour was detected, as demonstrated by Figure 4. The absence of an effect on the corrosion resistance is presumably caused by the chromium-enriched phases being distributed throughout the cold-worked microstructure, rather than localised precipitation at grain boundaries.

The second increase in hardness, which is observed in all three hardness curves (Figure 2), is attributed to the formation of α″. No indication of overaging was noted, as both the 880°C and 930°C curves displayed a sustained increase with aging time. The Mössbauer studies revealed that α″ does form at this composition and temperature. In
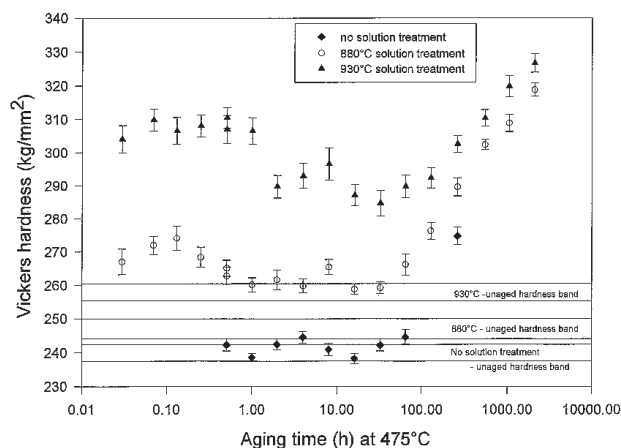


Figure 2—Hardness after solution treatment at 880°C and 930°C respectively, followed by cold rolling (38% reduction in area) and aging at 475°C. 95% confidence intervals are shown. The hardness of the material after the different solution treatments and cold rolling, but before aging, is given as hardness bands
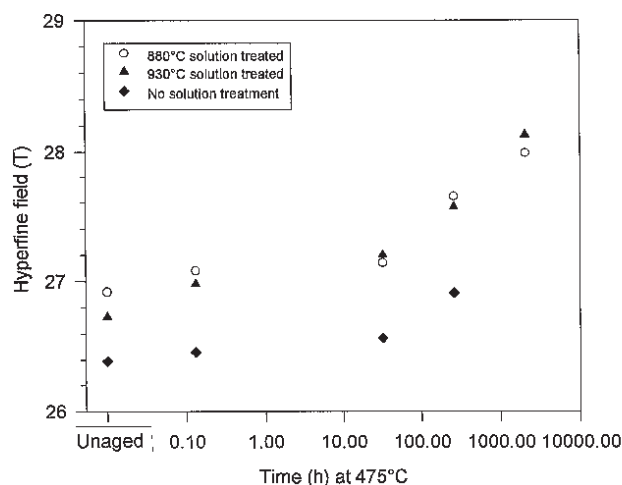


Figure 3—Increase in hyperfine field with aging at 475°C, for specimens solution treated at 880°C and 930°C, as well as samples not solution treated. The initial increase in magnetic field of the solution treated specimens is apparently caused by the precipitation of Cr-rich carbides and nitrides
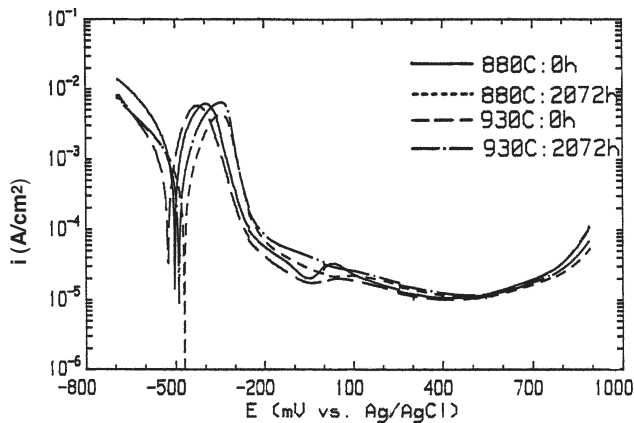
Figure 4a—Polarisation diagrams of (i) solution treatment at 880°C, no aging; (ii) solution treatment at 880°C, aging at 475°C for 2072 hours; (iii) solution treatment at 930°C, no aging; (iv) solution treatment at 930°C, aging at 475°C for 2072 hours
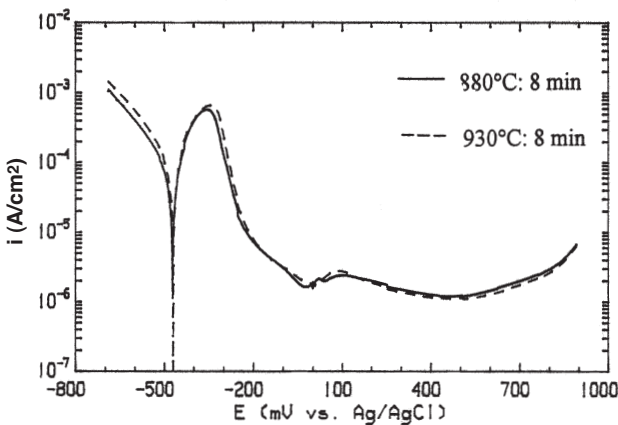


Figure 4b—Polarisation diagrams of (i) solution treatment at 880°C, aging at 475°C for 8 minutes and (ii) solution treatment at 930°C, aging at 475°C for 8 minutes. Results indicate that aging has little or no effect on the corrosion resistance

Figure 5 (the Mössbauer spectrum of a sample solution treated at 930°C and aged for 2072 hours at 475°C) the six peaks are characteristic of $\alpha$-Fe[9,10], and the presence of the $\alpha''$ is indicated by the small peak at zero velocity (indicated by the arrowhead), arising from the paramagnetic contribution of the chromium-rich phase.

The aging behaviour depends not only on the initial dissolved interstitial content, but also on the initial microstructure. This is demonstrated by the material containing martensite.

Solution treatment at the higher temperature (930°C) increased the initial hardness by 10 kg/mm² from 247 kg/mm². This increase is most probably due to the hardening effect of martensite in the structure, since all the austenite which had formed during the solution treatment transformed to martensite upon quenching, yielding approximately 15% martensite in the structure (see also Figure 1). The initial hardness increase during the first few minutes of aging is also much higher for the 930°C solution treatment, but at very long periods of aging the two curves converge. This convergence is probably due to tempering of the martensite since examination under the microscope did indicate
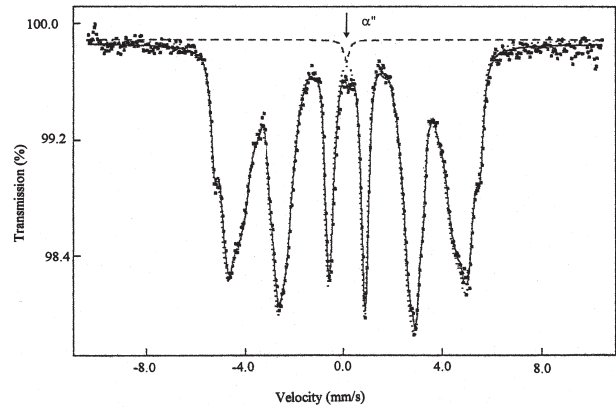


Figure 5—Mössbauer spectrum of a 930°C solution treated specimen, cold worked and aged at 475°C for 7072 hours. The small peak at zero velocity indicates the presence of $\alpha''$

tempering at the longer aging times.

From the hardness results it can be concluded that initial solution heat treatments have a positive effect on the short-time aging behaviour, yielding the same increase in strength in 8 minutes as can otherwise be obtained in 22 days of aging. It utilises the precipitation of Cr-rich carbides and nitrides, rather than that of $\alpha''$. It is worth noting that no effect of dissolved interstitials on the kinetics of $\alpha''$ formation could be detected. A plausible explanation is that the matrix is strained as a result of cold work, minimising the effects of additional strain by dissolved interstitials.

**Tensile tests**

The results of the tensile tests are summarised in Figure 6. Two aspects are noted.

(a)  In the unaged material, plastic flow occurs immediately upon application of an external stress—no elastic region is observed. This is attributed to the residual stresses which arise as a result of inhomogeneous deformation during cold rolling. Because of the residual stresses, parts of the material are at, or close to, the tensile yield stress, even with no externally applied stress. This causes the material to exhibit macroscopic plastic flow even at low applied stress, as is evident in the tensile behaviour of the as-rolled
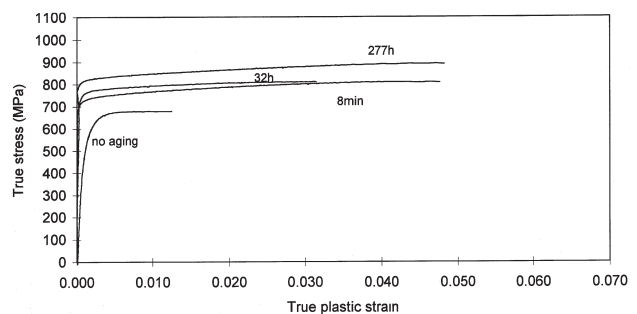


Figure 6a— True stress/true plastic strain during tensile testing of 880°C solution treated specimens, aged at 475°C. The curves terminate at the start of neck formation

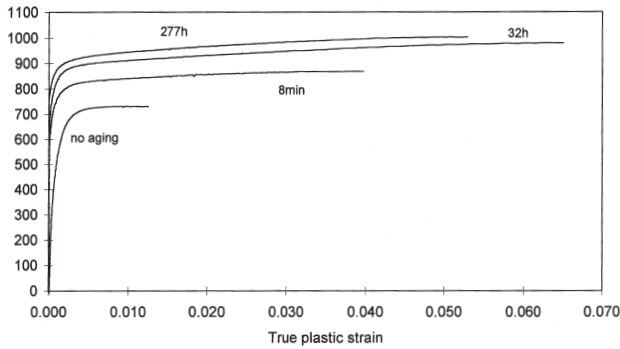# Strengthening of a cold worked 17% Cr ferritic stainless steel



**Figure 6b— True stress/true plastic strain during tensile testing of 930°C solution treated specimens, aged at 475°C. The curves terminate at the start of neck formation**

material[11,12]. Such residual stresses can be relieved, by heating the metal to a temperature where the yield stress of the material is the same or less than the value of the residual stresses, thus allowing the material to deform and relieve the stress[11]. With aging, even after only 8 minutes at 475°C, these residual stresses are relieved, as is evident from the tensile results. This behaviour was noticed in both the 880°C and 930°C solution-treated specimens and is in accordance with the results seen in transmission electron microscopy analyses, showing recovery with aging (Figure 7). Cold deformation increases the hardness by increasing the dislocation density; Figure 7a illustrates the high dislocation density after cold rolling. With recovery the dislocations group together (Figure 7b) in low energy arrangements, resulting in the formation of a sub-grain structure in which the centre of each grain is relatively perfect[11]. No recrystallisation was observed, which explains in part why no reduction in strength occurred during initial aging.

(b)   In addition to the relief of residual stresses, the work hardening behaviour is changed by the first aging step (0 to 8 minutes). This is evident from the increase in uniform elongation and in plots of the work hardening rate ($\theta$). The work hardening rate is defined as the slope of the true stress–true strain curve ($\theta = d\sigma/d\varepsilon$); results are shown in Figure 8.
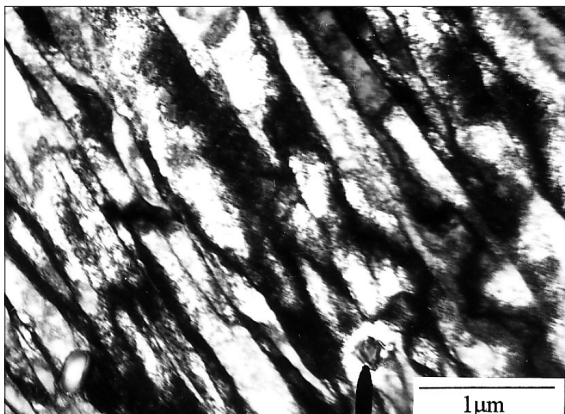


**Figure 8a—The effect of plastic strain on the work hardening rate ($\theta$) of samples solution treated at 880°C; cold rolled and aged at 475°C**



**Figure 8b—The effect of plastic strain on the work hardening rate ($\theta$) of samples solution treated at 930°C; cold rolled and aged at 475°C. The results indicate a change in work hardening behaviour after the first aging step, with higher work hardening rates at higher strains for the aged specimens**



**Figure 7a—TEM micrographs (bright field) of the matrix after cold working with no aging**

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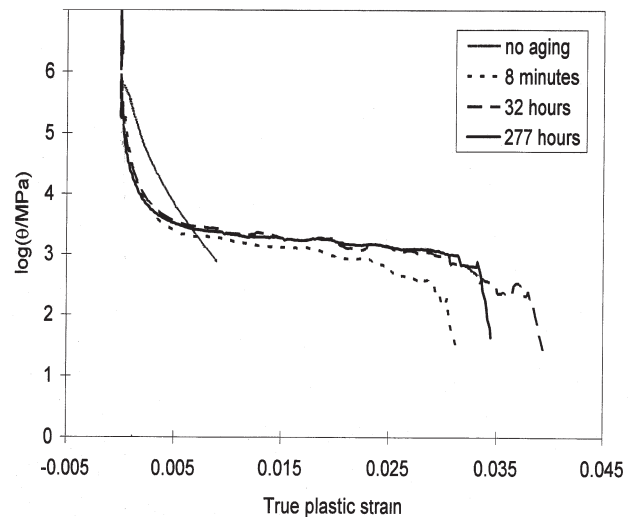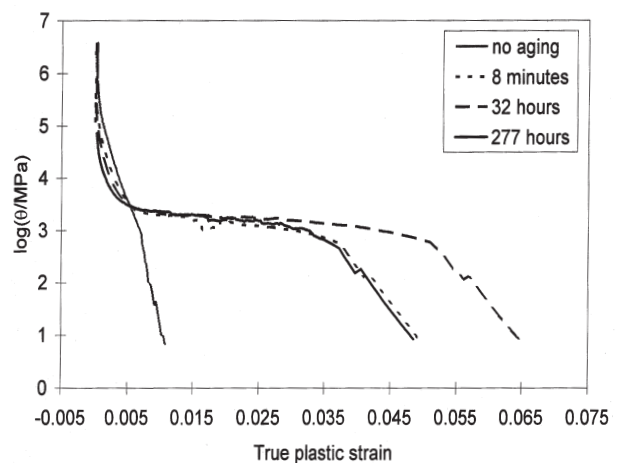
The apparently higher initial work hardening rate of the unaged specimens results from the occurrence of flow at low applied stress (i.e. the effect of residual stresses). However, the work hardening rate of the unaged specimens drops below that of the aged samples at a true strain of around 0.006, resulting in the lower uniform elongation of the unaged specimens. The higher work hardening rate of the aged specimens indicates a more rapid increase in dislocation density during deformation; this is presumably the result of the precipitation of non-deforming particles during aging (or the creation of similar barriers to dislocation movement)—the well-known Orowan mechanism for dispersion hardening. According to this mechanism, new dislocation loops are created around the non-deforming particles as external deformation causes dislocation movement. These particles acting as barriers in this case are probably the proposed Cr-rich carbides and nitrides and—after longer aging times—$\alpha''$. Similar work hardening rates for all the aged specimens probably indicate the same mechanism of strengthening.

A plot of the yield and tensile strengths (Figure 9) indicates that these properties exhibit similar aging behaviour to the hardness results, namely, an increase at short aging periods and again at longer ones. Both the 880°C and 930°C solution treatments result in yield strengths within the required limits for conveyor chains (750-850 MPa).
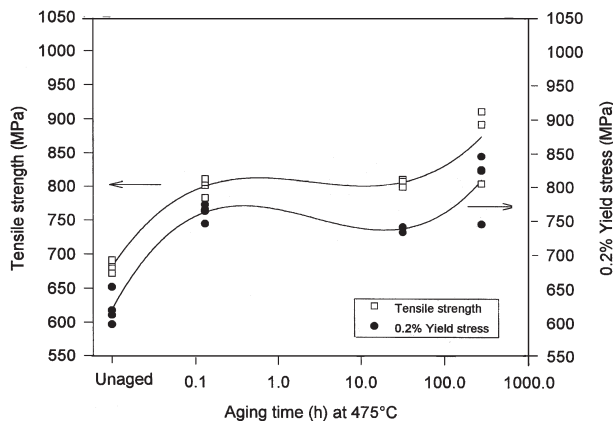


Figure 9a—Yield and tensile strengths of 880°C solution treated specimens, cold rolled and aged at 475°C
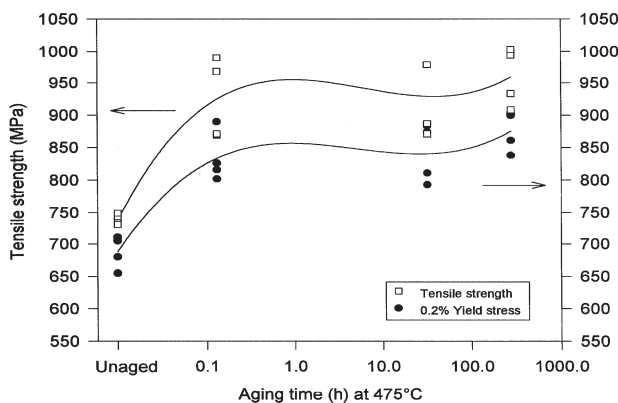


Figure 9b—Yield and tensile strengths of 930°C solution treated specimens, cold rolled and aged at 475°C. Results echo the hardness data with an increase at short aging periods and again at longer times
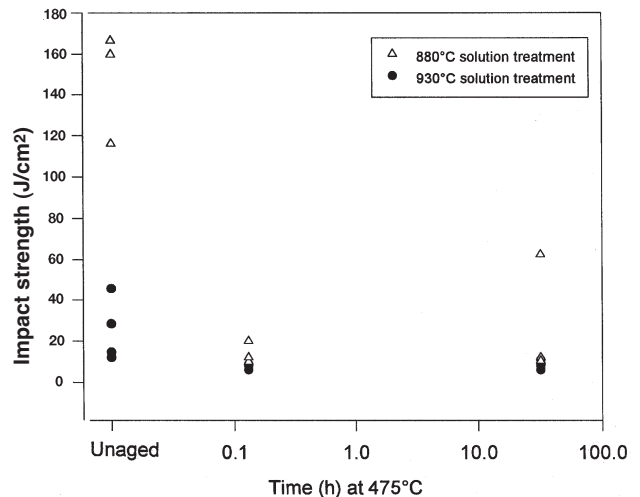


Figure 10—Impact strength of specimens aged at 475°C, after solution treatment (at 880°C and 930°C respectively) and cold rolling. A marked decrease in toughness occurs with aging

## Impact tests

A marked drop in impact strength after 8 minutes of aging is evident from the impact results (Figure 10), which indicate that embrittlement accompanies the increased hardness obtained by aging. For some indication of the significance of this decrease in impact strength, $K_{Ic}$ toughness values were estimated from the impact strength (CVN), using the following relationship:

$$K_{Ic} = 6.79 \, (CVN)^{0.75} \qquad [1]$$
$$\text{with} \qquad K_{Ic} \text{ in MPa (m)}^{0.5}$$
$$CVN \text{ in J/cm}^2.$$

Equation 1 can be used for ferritic steels[13] having a yield stress between 270 and 1700 MPa, and an impact range of 14–82 J/cm². The properties of the material investigated here mostly fall into these ranges.

Using the estimated $K_{Ic}$ values, critical crack lengths were calculated using the relationship for a single edge crack:

$$K_{Ic} = \sigma \, (\pi a)^{0.5} \qquad [2]$$
$$\text{with} \qquad \sigma \text{ the flow stress in MPa (determined through tensile tests)}$$
$$a \text{ the critical crack length in m.}$$

The critical crack lengths were determined by using the calculated $K_{Ic}$ values and actual, measured yield stress values. This represents a fairly conservative approach, as the yield stress is expected to be rather higher than the work stresses to which the chains are exposed. The same yield strength value was used for all the specimens of a particular heat treatment.

It is indicated by the calculated critical crack lengths that in the worst case, the maximum flaw length that can be tolerated is 0.3 mm at a stress equal to the yield strength of 860 MPa (specimen solution treated at 930°C, cold rolled and aged at 475°C for 8 minutes). For the specimens solution treated at 880°C, aging results in a maximum allowable flaw length of 0.8 mm after 8 minutes' aging at 475°C (for a stress of 760 MPa).

It is important to emphasise that these results are for tests at room temperature; the fracture toughness is expected

to drop further at lower temperatures—which would be of importance should the chains be used in a refrigeration plant.

The critical crack length (for the 880°C solution treatments) is comparable to the thickness of the chain links (3 mm). It is not expected that there would be flaws of this (calculated) size in the as-manufactured links. Hence, for catastrophic failure to occur, a fatigue crack must first be initiated. Therefore, fatigue is expected to determine the practical lifetime of the chains, although the lower $K_{Ic}$ values indicate that less crack propagation will be tolerated before brittle fracture. The 880°C solution treatment is the more prudent option, as it allows for a larger critical flaw size. Practical applicability of chains in this heat-treated condition is expected to hinge on whether the toughness after heat treatment is adequate under the operating conditions.

### Field trials

Trial runs of this material are under way in breweries and bottling plants both in South Africa and neighbouring states. Its performance appears to be satisfactory, but final results are not yet available.

### Conclusions

Strengthening is caused by a combination of different mechanisms which include cold deformation, the presence of martensite, precipitation of carbides and nitrides and formation of $\alpha''$.

From the hardness results it can be concluded that additional solution heat treatments have a positive effect as these affect the short-time aging behaviour, and make it possible to obtain the same increase in strength in 8 minutes as can be obtained in 22 days of aging. It utilises the precipitation of Cr-rich carbides and nitrides, rather than that of $\alpha''$. Both the 880°C and 930°C solution treatments result (after aging) in yield strengths within the required limits for conveyor chains (i.e. 750–850 MPa). Aging does not seem to have a significant effect on the corrosion behaviour.

Only 8 minutes of aging is necessary to ensure recovery from the residual stresses induced by cold rolling. The recovery does not cause a decrease in strength.

Similar work hardening rates for all the aged specimens probably indicate the same mechanism of strengthening, namely the precipitation of undeforming particles.

Impact strengths indicate that the 880°C solution treatment is the more prudent option, as it allows for a larger critical flaw size while providing adequate strength.

In the light of the results it can be concluded that the preferred processing route for the chains would be solution treatment at 880°C, fast quenching to trap the interstitial atoms, cold rolling and then aging at 475°C for 8 minutes.

### Acknowledgements

### References

1. CORTIE, M.B. Stainless steel components made from sheet material. *S. A. Patent no.* 94/0880. 1994.
2. KRAUSS, G. Stainless steels. *Steels: Heat Treatment and Processing Principles*. ASM International, USA. 1989.
3. CHANDRA, D., and SCHWARTZ, L.H. Mössbauer effect study of the 475°C decomposition of Fe-Cr. *Metallurgical Transactions,* vol. 2. pp. 511–519. 1971.
4. DE NYS, T., and GIELEN, P.M. Spinodal decomposition in the Fe-Cr system. *Metallurgical Transactions,* vol. 2. pp. 1423–1428. 1971.
5. GROBNER, P.J. The 885°F (475°C) embrittlement of ferritic stainless steels. *Metallurgical Transactions,* vol. 4. pp. 251–260. 1973.
6. NICHOL, T.J., DATTA, A., and AGGEN, G. Embrittlement of ferritic stainless steels. *Metallurgical Transactions series A*, vol. 11. pp. 573-585. 1980.
7. CZAKÓ-NAGY, I., and VERTÉS, A. Mössbauer spectroscopy as an analytical tool. *Trends in Analytical Chemistry,* vol. 7, no. 8. pp. 305–310. 1988.
8. CORTIE, M.B. Embrittlement and aging at 475°C in an experimental ferritic stainless steel containing 38% Cr. *Materials Science and Engineering A,* vol. 199. pp. 153–163. 1995.
9. GONSER, U. Mössbauer spectrum in physical metallurgy. *Hyperfine Interactions,* vol. 13. pp. 5–23. 1983.
10. MONTANO, P.A. Technological applications of Mössbauer spectra. *Hyperfine Interactions,* vol. 16. pp. 147–159. 1986.
11. DIETER, G.E. *Mechanical Metallurgy*. McGraw-Hill, London. 1988.
12. HOSFORD, W.F., and CADDELL, R.M. *Metal forming, Mechanics and Metallurgy*. Prentice-Hall, Englewood Cliffs, New Jersey. 1983.
13. HERTZBERG, R.W. *Deformation and Fracture Mechanics of Engineering Materials*. John Wiley and Sons, U.S.A. 1983. ◆

# Jubilee conference selects Wits Professor*



Professor H. Eric of the School of Process and Materials Engineering at the University of the Witwatersrand is one of a select group of experts around the world invited to participate in the 1996/97 golden jubilee celebrations of the Department of Metallurgy of the Indian Institute of Science in July.

He has been invited to address the international conference on Recent Advances in Metallurgical Processes in Bangalore, India.

The conference will highlight diverse areas in the Department of Metallurgy, with in-depth discussions on current trends in the processing of minerals, metals, polymers, ceramics and a wide range of composites. It will provide a forum for researchers and practising engineers to interact, review and assess the latest developments in the field.

Professor Eric's contributions in this field are renowned and the conference organisers have asked him to deliver a lecture on his involvement in the carbothermic reduction behaviour of chromite from the Bushveld Complex.

This subject has been researched extensively during the past decade by the School of Process and Materials Engineering at Wits University. An intensive study has been made of solid-state reductions from 1300 to 1500 degrees centigrade in the absence and presence of a silica flux. The processes studied focus on the reduction, smelting and refining stages of chromium.

The Department of Metallurgy at the Indian Institute of Science, established in 1945, is recognised as one of the foremost centres for advanced metallurgical education and research in India.   ◆

# Gold plant commissioned successfully*



The Mintek-built solvent-extraction (Minataur) refining plant at Randgold's Harmony Gold Mine in Virginia was successfully commissioned during June.

The Mintek-developed technology, which had been proven at a precious-metals refinery in Mexico, and at Harmony during 1996, will process Harmony's total output, and indications are that material savings, both at the operational level and in financial turn-around time, could be in excess of R5 million per annum.

The Harmony Pure Gold Project, which marks a number of milestones in the South African gold-processing industry, will refine gold slimes to produce gold of 'four-nines' purity on site, reducing the amount of time required to produce this gold to a mere 18 hours.

This is the first time a South African gold mine will refine its own gold on site, and Harmony has furthermore been given permission by the SA Reserve Bank and the Department of Trade and Industry to establish a jewellery-manufacturing facility at the plant, which will enable the mine to both add value to, and market its own gold.

Mintek President, Dr Aidan Edwards, long the champion of adding value in the South African gold industry, has heralded this event as a significant milestone which will have profound implications for the future of South Africa's gold-related industries.

He sees high-purity gold refining becoming commonplace on South African mines as economic sense will prevail when it becomes evident how much time and money is being saved at the Harmony plant.

'In addition, when we consider that South Africa produces only 5 tons of gold jewellery (some 0.2% of the world total), and that Harmony alone could potentially supply 8 tons through its on-site manufacturing facility, with other producers ultimately following suit, the writing is on the wall that at last South Africa will become a major participant on the international market.

'Such activities could very well provide the critical impetus for the extension of the lives of our mature mines - because of the significant reduction in the costs of refining, and because a percentage of profits from such downstream operations could be ploughed back into the operations'.   ◆