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THE PROBLEM OF DIMENSIONAL INSTABILITY IN AIRFOIL MODELS FOR CRYOGENIC WIND TUNNELS

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The Problem of Dimensional Instability in Airfoil Models for Cryogenic Wind Tunnels

1. Introduction

Optimization of the choice of material and fabrication techniques for construction of models for cryogenic wind tunnels such as the National Transonic Facility has presented designers with an almost insuperable problem. Many of the properties required are near the limits attainable with state-of-the-art technology, and in many cases improvements in one direction seem inevitably to be accompanied by losses in others. Thus, for example, the material has to have a yield stress high enough to carry the imposed aerodynamic loadings, yet be tough enough to operate safely at cryogenic temperatures. It has to be capable of being fabricated using available machining and joining techniques to give a model with a precisely known shape and a high surface finish which is able to retain dimensional stability during thermal cycling between ambient and its cryogenic operating temperatures. It has to be either intrinsically resistant to, or capable of being protected against, corrosion and degradation and, if it is to be of maximum use as an aerodynamic test facility, it has to be furnished with a complex array of holes, tubes, sensors, heaters and other experimental facilities.

While many of these requirements have been familiar to generations of experimental aerodynamicists, it is the high Reynolds Number requirement and, in particular, the added cryogenic dimension that has raised the designers' challenge to its present level. Tobler (Ref.1) has given a comprehensive introduction to materials for cryogenic wind tunnel testing, while Hudson (Ref.2) has described in detail the criteria used in the selection of Pathfinder I, a research and development model intended to highlight the problem areas associated with operating at the high loads and low temperatures that will be experienced in the N.T.F..

The need for adequate strength and toughness for safe operation at cryogenic temperatures severely limits the range of materials available for model construction. In order to meet the minimum acceptable toughness requirements, many high-strength materials have to be heat-treated to a lower strength condition and this can lead to dimensional instability, either as a result of the formation of an unstable metallurgical structure or due to the stresses induced on cooling from the heat-treatment temperature. Furthermore, most conventional fabrication techniques create, to a greater or lesser extent, tensile or compressive stresses in the material and in many cases they also result in a localized modification to its microstructure. Considerable warpage can be caused either by the creation or relief of these stresses during model fabrication, while additional dimensional instability can be caused by the strains induced by differential thermal contraction during temperature cycling between room and cryogenic temperatures. Finally, stress relief, conventionally obtained by high-temperature heat-treatments, can also in some cases be obtained by suitable cryogenic temperature cycles.

The ultimate dimensional stability of a cryogenic wind tunnel model will therefore depend on the basic metallurgical stability of the material chosen and the ability of the designer and manufacturer to devise a schedule of machining and heat-treatment cycles that will minimize residual and operationally-induced stresses.

This report seeks to identify, and where possible separate out, the many factors that contribute to dimensional stability in cryogenic models, and to provide a possible framework on which existing information can be understood and future work planned in a co-ordinated and meaningful manner.

2. Background to the Present Investigation and Basic Diagnosis

In June 1981 a post-testing co-ordinate check was carried out on a 12% airfoil manufactured from 15-5PH stainless steel that had been tested in the 0.3m. cryogenic wind tunnel at LaRC. Comparison with similar measurements taken before testing showed a 0.002in. decambering of the aft airfoil section (6in.chord) and a 0.006in. bow over the 8in. span. Following initial discussions between the author and LaRC personnel on possible causes of this warpage, a second similar airfoil was cooled to 78K (-196C,-320F) in a liquid nitrogen storage vessel and it too warped by a similar amount, thereby eliminating aerodynamic loading as a major cause of the problem. As 15-5PH stainless steel had been used successfully for airfoils in ambient temperature wind tunnels, all indications were that it was the cryogenic temperature cycle that was responsible for causing the dimensional instability.

There are two basic mechanisms that can lead to warpage:

- metallurgical structural instability in which one phase transforms partially or fully into a second phase which has a different crystal structure and volume.
- deformation due to the creation, or relief, of unbalanced induced or residual stresses.

In the case of the 15-5PH airfoils, it is highly probable that it was the former of these mechanisms that was responsible for the majority of the observed warpage. In order to achieve the maximum toughness at 78K, the material had been heat-treated to the H1150M condition which gave a Charpy impact energy in excess of the 25ft.lb. minimum required. Data supplied by the manufacturers, ARMCO, (Ref.3) showed the relationship between strength, impact energy and contraction after heat-treatment that is given in Table 1, together with our comments on the probable phase structures produced by the various heat-treatments.

Table 1: 15-5PH Stainless Steel Stability Data

Condition	CV,-320F (ft.lb)	Contraction H.T.-R.T. (in./in.)	Cryocycle Stability	R.T. Tensile (ksi)	Structure
H 900	-	.00045	-	190	Martensitic (m/s)
H 1025	2.0	.00053	good	155	m/s
H 1100	3.5	.0009	-	140	m/s
H 1150	-	.0022	-	135	m/s + austenite
H 1150M	33.0	.00243	poor	115	m/s + austenite

It can be seen that considerable contraction takes place after the H1150 and H1150M heat-treatments due to the reformation of about 12-15% austenite from the previously tempered martensitic structure. This austenite is, however, only metastable and low temperature cycling, machining or other forms of deformation trigger off a partial transformation back to martensite which is accompanied by a volume expansion. In an asymmetric section such as an airfoil, and where the effect of machining would be more pronounced in thinner sections than the thicker parts, these volume changes would show up as warpage.

The energy required to trigger off the austenite to martensite transformation, which takes place by instantaneous shear rather than by a nucleation and diffusion controlled growth mechanism, is probably provided by differential thermal contraction across the section. Large temperature gradients would encourage such transformation, as would rapid temperature changes, and thus it is the RATE of cooling and warming and the NUMBER of cryogenic cycles that determine the degree of transformation, rather than the LENGTH OF TIME held at a particular temperature. Rapid changes of section would also exaggerate the problem as larger temperature gradients, and hence higher thermal stresses, are set up across the thicker sections.

Confirmation for this hypothesis came from subsequent experimental work which showed that the 15-5PH flat specimen in the H1150M conditions continued to distort even after many cryogenic cycles. Furthermore, a similar 15-5PH airfoil that had been heat-treated to the H1025 condition showed it to be more stable when cryocycled, due probably to its structure being 100% tempered martensite with no reformed austenite.

Structural instability is therefore one important mechanism responsible for warpage in cryogenic wind tunnel models. However, tests carried out at LaRC on metallurgically stable materials such as the 300 series and the nitrogen-strengthened, high-manganese Nitronic 40 stainless steels, showed that they too were not completely free from warpage, albeit to a less serious degree, after cryogenic cycling.

These initial investigations also showed up a serious problem with the Nitronic 40 being used for the construction of Pathfinder I and other models for use in cryogenic wind tunnels. It was found to contain small quantities of delta ferrite and to be in a sensitized condition due to the presence of grain boundary carbides, both of which lowered its toughness as measured by Charpy impact tests at 78K. The results of investigations to characterise and restore the properties of this material are given in an LaRC High Number contractor report (Ref.4) written by the author of this report.

3. Microstructural Stability in Steels

3.1 Single phase austenitic stainless steels

The face-centered-cubic austenitic phase of iron normally found at temperatures in excess of 910°C can be stabilized to room temperature and below by the addition of sufficient austenite stabilizers such as Nickel, Manganese, Carbon and Nitrogen. For example, the 20% nickel in type 310 makes it particularly stable, while the high nitrogen contents of types 304N, 316N and 347N and the high manganese Nitronic 40 ensure that these stainless steels are also highly resistant to transformations. Strictly speaking, the austenite phase is only metastable and in a number of the

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leaner grades with nickel contents on the borderline of the 8% required for stability, transformation, accompanied by a volume increase, can take place to the harder ferromagnetic martensite phase. This change can be induced thermally by lowering the temperature below the M_s temperature, values of which can be calculated by formulae such as that given by Eichelman and Hull (Ref.5) which balance out the austenite stabilizing effect of elements such as Ni, Mn, C and N against that of ferrite stabilizers such as Cr, Si, Mo, W, etc. The lower the value of M_s with respect to room temperature, the more resistant is the alloy to martensitic transformation induced by thermal cycling. Thus, for example, some heats of type 304 in which M_s is just below room temperature show large amounts of spontaneous transformation on cycling between room and liquid nitrogen temperatures.

Also significant is the so-called M_D temperature, usually a few hundred degrees higher than the M_s, which indicates the degree of stability against martensite induced by plastic deformation during forming and machining operations. Thus, for example, a material with an M_D temperature above ambient is likely to have considerable amounts of martensite formed during machining. Formulae for M_D based on the chemical composition of the alloy have also been developed for some austenite steels (Ref.6). Furthermore, as martensite is more stable than austenite it is necessary to reheat to temperatures of 1000C (1900F) or higher before the martensite is completely resolutionised. Lower temperature heat-treatments do, however, temper the martensite and make it tougher.

It should be noted in passing that heat-treatment of the austenitic stainless steels needs to be carried out with some care. Excessive time spent in the temperature range between 590C and 920C (1100-1700F) can cause the material to become sensitized due to the precipitation of carbides and sigma phase in its grain boundaries, and at room temperature this causes a loss of corrosion resistance and the consequent liability to intergranular attack. Of much greater importance for cryogenic applications is, however, the serious loss of toughness at liquid nitrogen temperatures due to the ease with which fracture is nucleated and propagated in a low energy mode along the grain boundaries. Unfortunately, airfoil models are frequently cooled through this sensitizing temperature range after post-machining, stress-relieving heat-treatments at 1000C (1900F), and unless cooling is rapid, quite severe loss of toughness can occur. [Sensitization in Nitronic 40 is discussed at length in Ref.4].

One major disadvantage of the 300 series austenitic stainless steels is their relatively low yield strengths, but the addition of up to 0.4% nitrogen, together with the manganese in the 21Cr,6Ni,9Mn stainless, Nitronic 40 gives a material able to attain the 1.03 G.Pa (150ksi) minimum yield stress at cryo temperature required for the NTF Pathfinder I model. As long as the material is not degraded by sensitization or the presence of delta ferrite, it is also capable of exceeding comfortably the minimum impact energy requirement of 34J (25ft.lb) at 78K (-196C,-320F), hence its choice for the construction of Pathfinder I.

Other alloys with different manganese, nickel and nitrogen levels are also available which give mechanical properties similar to those of Nitronic 40. These include ARMCO's Nitronic 33 (18Cr,3Ni,13Mn) and Nitronic 50 (22Cr,13Ni,5Mn), and Carpenter's 18-18 Plus (18Cr,0Ni,18Mn,0.5N). At the 8th International Cryogenic Materials Conference in San Diego, August 1981, many papers were presented on high-manganese, high-nitrogen stainless

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steels by American and Japanese workers and it would seem that this is an area of significant ongoing interest. It is recommended that these developments should be followed to see whether materials are developed that are suitable for cryogenic wind tunnel models.

The remaining fully austenitic stainless steel seriously considered for construction of models for cryogenic wind tunnels is the precipitation-hardenable A286 which has in fact been used successfully for airfoils tested in the 0.3m T.C.T. at LaRC. Its somewhat low yield strength of 827MPa (120ksi) at 78K, together with difficulties in obtaining material in the appropriate product forms, caused it to be eliminated from final consideration for Pathfinder I, but its successful track record in the 0.3m T.C.T. suggests that its future use should not be discounted lightly.

3.2 The martensitic and semi-austenitic precipitation hardenable stainless steels

As noted in the previous section, the minimum yield stress considered acceptable for Pathfinder I is 1.03 GPa (150ksi) at 78K (-196C,-320F). This is not an exceptionally high stress level, and the fracture toughness requirement of at least 93.5 $MPa.m^{1/2}$ ($85ksi.in^{1/2}$)is not excessively cautious, but applied together they serve to narrow drastically the range of materials suitable for model construction. Basically, this is because most metallurgical techniques that increase the yield stress also cause a decrease in fracture toughness. Furthermore, as the critical flaw size in a structure is related to the crack size factor, $(K_{IC}/\sigma_y)^2$, an increase in yield stress without a corresponding increase in fracture toughness will lower the resistance of the material to unstable, low-energy crack propagation.

Yield strengths considerably in excess of 1.03GPa (150ksi) are obtainable from many of the martensitic precipitation-hardenable stainless steels such as 17-4PH, 15-5PH, 13-8Mo and Custom 450, and the semi-austenitic types 17-7PH, PH15-7Mo, PH14-8Mo, AM350 and AM355. However, in the fully aged conditions, their fracture toughnesses are far too low for cryogenic applications, and it is necessary to temper them to the H1150M condition to obtain even the minimum acceptable toughness. Unfortunately, this is achieved metallurgically by a partial reformation of martensite to austenite and, as we saw in Section 2 for the 15-5PH airfoil, this austenite can be unstable and transform back to martensite during subsequent exposure to cryogenic temperatures. As the austenite to martensite transformation is accompanied by a volume increase, warpage can occur in asymmetric sections such as airfoils.

A similar situation exists in 17-4PH in that between 12 and 15% austenite is reformed by the 1150M heat-treatment and it is highly probable that this austenite is in a metastable condition that will revert to martensite during cryogenic cycling, thus creating warpage.

In the case of the 13-8Mo alloy, again much of the picture is the same as that for the 15-5PH and 17-4PH alloys. For exa-ple, Table 2 taken from the ARMCO Product Data Bulletin on 13-8Mo (Ref.7), shows a significantly higher contraction on cooling from the 1150 and 1150M heat-treatments than that found after the lower temperature heat-treatments.

Table 2: Dimensional Changes after Heat-Treatment of 13-8Mo Stainless Steel (Ref.8)

Hardening Temperature, ^{o}F (C)	Contraction* in/in (mm/mm)
950 (510)	.0004 to .0006
1000 (538)	.0004 to .0006
1050 (566)	.0005 to .0008
1100 (593)	.0008 to .0012
1150 (621)	.0030
1400+1150 (760+621)(1150M)	.0035

*Samples in Condition A prior to heat treatment.

Once again this is probably due to the reformation of a significant percentage of austenite during the H1150 and 1150M heat treatments. Perry and Jasper (Ref.8) make the following comment on the microstructure of 13-8Mo in various forms of heat-treatment:

"(After) heat-treatment at the lowest ageing temperature, in this case 900F (482C), the microstructure is essentially completely martensitic. As the ageing temperature increases, so does the amount of reformed austenite. The H1150M condition (the softest for these steels) has a rather complex microstructure. Heating to 1400F(760C) results in much of the martensite going into solution at that temperature. Upon cooling to room temperature, some of the austenite is transformed to untempered martensite. The rest of the austenite remains as austenite and the balance is highly overaged martensite. The 1150F (620C) ageing then ages the martensite that was formed as a result of cooling from 1400F (760C), together with some additional reformed austenite. Therefore, the final microstructure consists of highly overaged martensite, normal overaged martensite and reformed austenite which is completely thermally stable (our underlining). This results in a heat-treated stainless steel with reasonably good impact strength at temperatures as low as -320F (-196C)".

The relevance of considering the metallurgical stability of 13-8Mo in the H1150M condition in such detail lies in its choice for the backup solid wing on Pathfinder I. Of critical importance is the validity or otherwise of the statement quoted above, that the austenite is "completely thermally stable". The possibility has to be faced that, although this might be true for conventional applicatons, the very high dimensional stability demanded of models for cryogenic wind tunnels might not be met by 13-8Mo in the H1150M condition. It is conceivable that, for example, the deformation induced by semi-finish machining might trigger off further austenite transformation, and warpage, during the initial exposure of the wing section to cryogenic temperatures. If so, it would be prudent to check for this possibility at a stage where there is still scope for remedial measures to be taken, for example many thermal cycles carried out between room and cryogenic temperatures might induce all the transformation from austenite to martensite to take place before the model went into service.

The final class of precipitation hardenable stainless steels are the semi-austenitics such as 17-7PH, PH15-7Mo, etc., which have a duplex microstructure of austenite and delta ferrite. They are capable of very high strengths and have been used for room temperature airfoil models, but their toughnesses are too low for them to be considered seriously for cryogenic use, and it is also probable that the two-phase structure would create a problem with their dimensional stability.

3.3 Non-stainless high strength steels

3.3.1 Maraging and quenched-and-tempered steels

Given the need to provide and maintain a high quality surface finish on cryogenic wind tunnel models, the designers' strong preference for stainless steels is readily understandable. However, there are a number of non-stainless steels that appear to offer the attraction of higher yield strengths combined with acceptable minimum toughnesses at cryogenic temperatures and these are being considered seriously for later models that will be required to carry even higher loads than Pathfinder I. It should be noted, however, that surface treatments capable of providing a tough, corrosion-resistant, scratch-resistant, high-quality, finish need to be investigated in parallel with an evaluation of these higher strength steels if an acceptable model is to be created.

Most important of these steels are the 18Ni maraging steels, in particular the 200 grade which has a yield stress of 1.86GPa (270ksi), together with a Charpy impact energy of 39J (28ft.lb) at 78K (-196C,-320F). Furthermore, the material is readily machined in the annealed condition and there is very little dimensional change during the single step ageing cycle at 480C (900F) which precipitation hardens the soft, low-carbon martensite. The resultant microstucture is stable and not affected by cryogenic temperature cycling. Despite problems in obtaining the 18Ni maraging steels in all required product forms, the 200 grade offers probably the most realistic opportunities for the construction of highly loaded models. It is, however, generally thought that the toughness of the higher-strength 250 grade is unacceptably low at 78K (-196C,-320F) for use in critically loaded components.

The quenched and tempered 9Ni-4Co steels, such as the HP9-4-20 which contains 0.2% carbon, have also been considered for model manufacture. HP9-4-20 has a yield strength of 1.3GPa (183ksi) at 78K (-196C,-320F) but its Charpy impact energy is only barely acceptable and it does not appear to match the advantages offered by the 200 grade 18Ni maraging steels. However, preliminary trials have shown that it might be possible to improve its toughness by the grain refining technique to be described in the context of the 12Ni steels.

3.3.2 Super grain-refined steels

The classic, fundamental metallurgical technique of grain refinement is the only mechanism that increases both yield stress and improves toughness. Virtually any alloy is capable of having its mechanical properties improved by grain refinement, but it is in steels that the technique has been exploited to its fullest. This is due primarily to the ease with which grain refinement can be achieved by suitable thermal cycling through the austenite to ferrite+carbide or austenite to martensite phase transformations.

In the iron-12%nickel-reactive metal alloys, Stephens and

Witzke of NASA Lewis (Ref. 9) Bhat at Carnegie-Mellon (Ref. 10) and Morris at Berkeley (Ref.11) have shown that it is possible to combine yield strengths of 1.64GPa (240ksi) with fracture toughnesses of 220MPa.m$^{1/2}$ (200ksi.in$^{1/2}$) at 78K (-196C,-320F). These properties are obviously very attractive for cryogenic model use and an evaluation exercise is being initiated by LaRC to give data on their machinability and dimensional stability.

4. Induced Stress Systems and their Relationship to Dimensional Stability

Warpage can be brought about either by the creation or relief of stresses which can be increased, or decreased, by one or more of the following mechanisms:

- a) unbalanced residual compressive and tensile stresses resulting from the original fabrication of the plate or bar stock;

- b) quench-induced stresses generated on cooling from high temperature heat-treatments;

- c) compressive or tensile stresses introduced by machining. These can be elastic and/or plastic depending on the degree of deformation created during mechanical working of the material, and they can cause phase transformations in the surface layers;

- d) localized stresses created by joining or clamping techniques;

- e) stresses created during temperature cycling by differential temperature gradients, particularly across uneven sections.

These stress systems can be of considerable magnitude, often locally exceeding the yield stress, and the resultant strains can produce deformation and warpage which gives rise to deflections of many thousandths of an inch on typical airfoil sections. Identifying and separating these mechanisms in individual cases is difficult, or often impossible, but some general guidelines can be laid down.

4.1 Residual fabrication stresses and possible relief by heat-treatment

Most product forms, plate, tube, bar, etc., that have undergone a significant amount of plastic deformation during fabrication arrive from the suppliers containing some form of residual stress system. It is common practice to carry out a stress-relieving heat-treatment prior to fabrication, but it is essential to have a good appreciation of the possible metallurgical effects of different heat-treatments. For example, the higher the temperature, the more rapid and extensive the stress relief, but the greater the risk of grain growth or of over-ageing a precipitate structure. Furthermore, in many alloys, particularly the austenitic stainless steels, there are intermediate temperature ranges in which undesirable structures, such as grain boundary carbides and sigma phase, can be formed if the material is held for too long. In such cases, it is normally recommended that oil or water quenching, or forced air cooling, is used to minimize the time spent cooling through the undesirable temperature range. Uneven heat removal during very rapid cooling can, however, create quenching stresses and, in extreme cases, quench cracks, that are often as serious as those to be removed by the heat-treatment.

One possible variation on the quenching theme that was used successfully during the fabrication of an experimental Nitronic 40 airfoil involved cryoquenching into liquid nitrogen. Despite its apparently drastic nature, this technique in fact gives an even and controlled heat flow out of the material through the gas blanket that is formed at the metal/liquid nitrogen interface. This technique could have increasing use for the construction of models for cryogenic applicatons because it has the added benefit of exposing the material to low temperatures early on in the manufacturing sequence, so giving an early indication of its cryogenic behaviour. Fuller details of this technique are given in Ref.4.

4.2 Stresses created during machining and other fabrication processes

All conventional machining techniques have in common the basic microscopic mechanism of removing material by shearing or fracturing a layer from the remaining mass, and for this to happen stresses must locally exceed the shear or fracture stress. Very high instantaneous temperatures can also be generated and the net effect of these high stress and temperature levels can be the creation of residual stresses in the remaining material, termed macrostresses if they are distributed uniformly over appreciable areas or microstresses if they vary from grain to grain or within the grain itself. Of particular relevance are the stresses developed in the thin layer beneath the machined surface which can significantly exceed the yield stress, and also change from tensile to compressive or vice versa within a few thousandths of an inch.

In metallurgically stable materials, plastic deformation in the surface layer creates excessive distortion and work hardening of the grains, indeed the induced stresses are closely dependent on the work hardening properties of the material. Lower metal removal rates and increased tool wear during subsequent machining passes follow from the creation of these hard surface layers, and in many cases intermediate annealing heat-treatments have to be employed to enable machining to continue. Temperatures ranging from 30 to 50% of the melting point are generally necessary to achieve recovery of the high dislocation densities created during cold working, and some recrystallization of the distorted grains can take place at the upper limit of this temperature range. In metallurgically unstable materials, such as the austenitic stainless steels discussed earlier, higher temperature heat-treatments are necessary to restore the deformation induced martensite phase to austenite.

Thus one technique for coping with machining-induced stresses is to reduce them by heat-treatments which may, or may not, create dimensional changes in the workpiece. Another technique frequently employed is, in effect, to balance the induced stresses by machining alternate faces, so setting up opposing forces which keep the material in the required shape. For simple shapes and sections, considerable success can be achieved in this way, but in more complex situations such as an instrumented airfoil, it would be very difficult to specify, let alone control, the appropriate degree of residual stress needed at different parts of its varying section and profile.

For cryogenic use the concept of balancing residual stresses is even more unlikely to be successful. The thermal stresses set up during cycling between room and cryogenic temperatures can be of a similar magnitude to the residual stresses induced by machining and, depending on their tensile or compressive nature, they can either increase, or decrease the magnitude of a particular local stress system.

> Thus, even in metallurgically stable materials, cryogenic cycling can alter significantly the residual stress system and thus create dimensional changes. It is often possible, however, to achieve dimensional stability by carrying out a number of thermal cycles between room and cryogenic temperatures, such that the stress system is stabilised to the extent that no warpage occurs during subsequent temperature cycles.

This conditioning or training effect is in some respects similar to the vibratory stress relief technique sometimes used to reduce residual stresses in pressure vessels and other welded structures. In the context of a cryogenic wind tunnel model in which extreme dimensional accuracy is required, the main difficulty comes in finding a technique by which the dimensional changes resulting from cryogenic or elevated temperature (or possibly vibratory) stress-relief can be corrected to give the required final profile. Let us assume for the sake of discussion that a model has been put into a balanced or stress-free state while still adequately oversize to allow removal of the small amounts of material required to establish the final profile. If the technique used in removing this material induces further machining stresses, then the cycle of warpage - thermal treatment for stress relief - dimensional-out-of-tolerance, will be restarted.

Ideally, machining techniques which do not create residual stresses are needed, and recent progress in Electro-Discharge Machining (EDM) using numerically controlled wires is to be welcomed. Indeed, it is possible to envisage a possible future, but as yet undeveloped, technique for the fabrication of 3-dimensional models in which the shape is created from a stabilized block of material by a combination of chemical milling, EDM and electropolishing that would be completely stress free.

In the meantime, however, it is unfortunate that the technique most frequently used for finish machining, namely surface grinding, can be responsible for creating high, localized surface stresses. In a rather old review on residual stress, Horger (Ref.12) quotes a number of tests in which surface stresses of tens or even hundreds of ksi were created during grinding. In one case, for example, (Ref.13) a tensile stress of 50ksi was set up at the surface by wet grinding SAE 1020 annealed steel even though the depth of cut was limited to 0.0003in.

Lapping is generally considered to induce fewer stresses than grinding but it is unlikely to be a practical possibility for creating complex profiles such as airfoils, except in the context of hand lapping or polishing to remove localized high spots. Electrochemical machining or polishing might be a more acceptable

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technique for removing metal from selected areas but further development will undoubtedly be required to build up experience of the factors necessary to achieve the required degree of control and reproducibility.

Nevertheless, for the immediate future and in particular for the current generation of models such as Pathfinder I, it will be necessary to stay with tried and established techniques of metal removal.

> We would, however, strongly recommend the adoption of the practice of exposing a model or component to one or more cycles of cryogenic cooling and co-ordinate verification <u>before</u> the final stages of finish machining. In this way the material will have a chance to undergo some form of stress adjustment and possible structural modification before final machining. By conditioning a model in this manner, it might be possible to avoid the embarrassment and expense of having a model warp excessively during its initial exposure to a cryogenic environment in the wind tunnel.

5. A Recommended Specimen Configuration for the Systematic Evaluation of Factors Influencing Warpage

From the arguments presented so far in this report it will be appreciated that there are many different factors which affect the dimensional stability of models designed for use in cryogenic wind tunnels. Many different tests and investigations have already been carried out, and which will be reported by LaRC.

Nevertheless, it was felt that a need clearly existed for the establishment of a standard specimen configuration that would be representative of the nature and scale of a typical airfoil model. Different machining and heat-treatment cycles could then be carried out on a range of materials in such a way that meaningful comparisons could be made between their results. Eventually, a comprehensive cross-correlation of these findings should, hopefully, help clarify the present somewha confused situation.

Accordingly, the configuration illustrated in Figure 1 was proposed by LaRC personnel and refined by the author. Its size was chosen to be economical in its demands on material, yet still have a section representative of a reasonable thickness of material. (A standard size Charpy V notch impact test specimen can in fact be cut from the thickest part of the sample). A stepped profile was chosen in preference to a wedge, partly because steps should be easier to machine and the reference flat on the underside be simpler to create, but also because the results should be easier to obtain and interpret. Furthermore, the steps are also typical of the sharp changes of section found in some airfoil models such as those with welded-on coverplates. Finally, the thinnest sections are typical of those found at the trailing edge of an airfoil.

A preliminary evaluation of this concept has been carried out at Southampton University and detailed results will be presented in a separate report. Nevertheless, some of the initial results on Vascomax 200 are worth including to give an idea of possible scope for the technique. The schedule of machining, heat-treatment and measurement cycles is given in Table 3, and Figures 2 and 3 demonstrate the results of traces across the length and width of the sample respectively.

Table 3: Machining and Heat-Treatment Schedule for Vascomax 200, Initially in Annealed Condition

a) Rough cut to shape 60mmx60mmx12mm.
b) Machine out 6mm and 3mm steps to leave 3mm step, 36mm long.
c) Heat-treat for 4.5hrs at 480C (900F): Check macro and microhardness.
d) Prepare 60mmx60mm unstepped face to reference standard.
e) Carry out appropriate measurement and mapping.
f) Use 0.5in. diameter ball end mill to reduce thickness of end 24mm to 1.5mm, 15thou.max. milling steps, liberal fluid to cool.
g) Remeasure and map reference surface after each milling step.
h) Re-check microhardness of milled surface (and microstructure?)
i) Grind end 18mm to final 1.5mm thickness, using maximum of 0.5thou. per cut, liberal water based emulsion to cool, 32 μinch finish.
j) Remeasure and map reference surface.
k) Grind end 12mm step to 0.75mm thickness using maximum of 0.5thou. per cut, same lubrication system and surface finish as for step i).
l) Remeasure and map reference surface.
m) Re-check microhardness of ground surface and microstructure.

Direct comparisons of the deflections recorded along the length of the specimens are possible because the capacitance probe is motor driven in this direction. The readings across the width have, however, had to be rescaled laterally because the probe had to be traversed by hand in this direction.

One immediately striking observation to be made about the longitudinal traverses is the magnitude and direction of the deflections produced by the ball end milling stages, namely a total deflection of almost 0.006in (0.15mm) at the extreme end, and in a sense that infers that compressive stresses have been created during milling. Even more fascinating is the result of the first grinding operation carried out on the end 18mm which reduced the total cumulative deflection at the end to about 0.002in (0.05mm), thus indicating that grinding had induced tensile stresses into the surface layer which partially cancelled out the compressive stresses created by milling.

If this result can be confirmed, quantified and controlled, it opens up the tantalizing prospect of being able to manipulate the stresses in the surface layers by appropriate combinations of milling and grinding steps. Indeed, it may even be possible to fine tune the dimensional changes required to establish, or correct, the required surface profile of an airfoil by this technique.

The traverses across the width of the sample near its thinnest edge show that milling causes it to bow with a greater deflection at the edges than at the centre. Furthermore, the magnitude of this effect builds up with increase in the number of milling stages. The dramatic effect of the subsequent grinding stages is also illustrated vividly as the bowing effect is not only reduced in magnitude but also changed in sign with maximum deflection now at the centre of the span.

Thus, even though this preliminary series of measurements is not yet complete, the initial results demonstrate that it is possible to generate meaningful data using the proposed standard specimen configuration.

6. Specification and Documentation of Materials, Machining Stages, Heat-Treatments and Co-ordinate Verification for Model Components and Experimental Samples

If maximum return is to be achieved on the time and money spent investigating the problems of dimensional instability in cryogenic wind tunnel models, it will be necessary to adopt some form of recording system for the various samples and models studied. In particular, establishment of an accurate record of the thermal, mechanical, chemical, processing and fabrication histories of each sample or model will make it very much easier to correlate and interpret the data and results generated during the warpage programme. Table 4 illustrates one possible format for such a record card. Further development would obviously be required before such a system were to be put into operation and the format given in Table 4 is to indicate the idea, rather than define the details.

7. Classification of Materials in Terms of their Metallurgical Structure and Possible Liability to Dimensional Instability

As has already been noted, there are not very many alloys suitable for the construction of models such as Pathfinder I for use in cryogenic wind tunnels. The reviews by Tobler (Ref.1) and Hudson (Ref.2) have discussed their choice in detail but it has been suggested that some form of condensed summary, together with comment on possible problems of dimensional instability would be of use. Accordingly, a first attempt at such a summary is presented in Appendix I. Further refinements could then be made as and when more definitive information becomes available as the warpage programme proceeds.

8. Conclusions

1. The problem of dimensional instability in airfoil models for cryogenic wind tunnels has been discussed in terms of the various mechanisms that can be responsible for the observed effects.

2. It is concluded that the initial problem encountered with the 12% airfoil fabricated from 15-5PH stainless steel in the H1150M condition, which warped in the 0.3m Transonic Cryogenic Tunnel, was due primarily to intrinsic metallurgical instability of 15-5PH in the H1150M condition.

3. The interrelationship between metallurgical structure and possible dimensional instability in cryogenic usage has been discussed for those steels of most interest for wind tunnel model construction. It is concluded that the fully austenitic or fully martensitic steels are likely to be of most practical use.

4. Other basic mechanisms responsible for setting up residual stress systems are discussed, together with ways in which

their magnitude may be reduced by various elevated or low temperature thermal cycles.

5. Although it is accepted that the immediate priority is to overcome the problems encountered by the first generation of cryogenic wind tunnel models such as Pathfinder I, it is strongly recommended that this opportunity be taken to set up a programme to co-ordinate and correlate the data already available and that yet to be produced.

6. A standard specimen configuration has been proposed for use in experimental investigations into the effects of machining, heat-treatment and other variables that influence the dimensional stability of materials of interest for cryogenic wind tunnel models.

7. Preliminary results of the use of this configuration for Vascomax 200 are presented briefly. They show that ball end milling induces residual compressive stresses and that subsequent grinding reduces the compressive stresses, probably by inducing residual tensile stresses.

8. It is recommended that a system of specification and documentation of materials, machining stages, heat-treatments and co-ordinate verification be set up for use with model components and experimental samples concerned with the dimensional stability programme. One possible format for such a record system is proposed.

9. A brief classification of materials in terms of their metallurgical structure and possible liability to dimensional instability has been attempted. It is intended that this summary would be extended and updated as more definitive information becomes available.

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Table 4: Possible Format for Record Card detailing History of Material Used in Wind Tunnel Model Manufacture
(Hypothetical possible entries as indicated)

Side 1
Identification header to give:
Order Number, Cast Number, Mill Cert., Delivery Details,
Physical State, e.g. rolled, cast, cold drawn.
Chemical Composition
Ex-Mill Cert. or independent analysis.
Processing Record

Date	State	Procedure	Size (in.)	Heat-Treatment	Comment	Initials
-/-/-	As delivered	-	30x12x2	As rolled	-	AJP
-/-/-	Pre-machining	Sawcut	18x6x2	-	-	CG
-/-/-	Stabilise HT	-	-	30min @ 1950F then LN quench	-	AB
-/-/-	Rough machine	Mill	taper 5" to 2" thickness 1.5"to 0.5"	-	High cutter wear rate	CG
-/-/-	HT to anneal cold work	-	-	3hr @ 1100F then forced air cool.	-	AB
-/-/-	Further machine	Drill	.25" locating holes	-	-	MM
-/-/-	Braze HT	-	-	5 min @ 1800F then forced air cool	-	AB
-/-/-	Finish machine	Ball end mill	-	-	-	MM
-/-/-	Surface finish	Hand polish	-	-	16 in finish	FP
-/-/-	Protection	Lacquer	-	-	.0005"coating	DW

etcetera...

Side 2

Measurement Record

Date	State	Technique	Result/Comments	Initial
-/-/-	Finish machined	Validator	-	SPH
	"	Magne gauge survey	Sheet OOX	PQL
	Cryocycled	Validator	Room Temp -320 in 10mins	AR
		Magne gauge survey	Sheet OOX	PQR
	7-day exposure	Visual/photograph	Report......	ZR

etcetera....

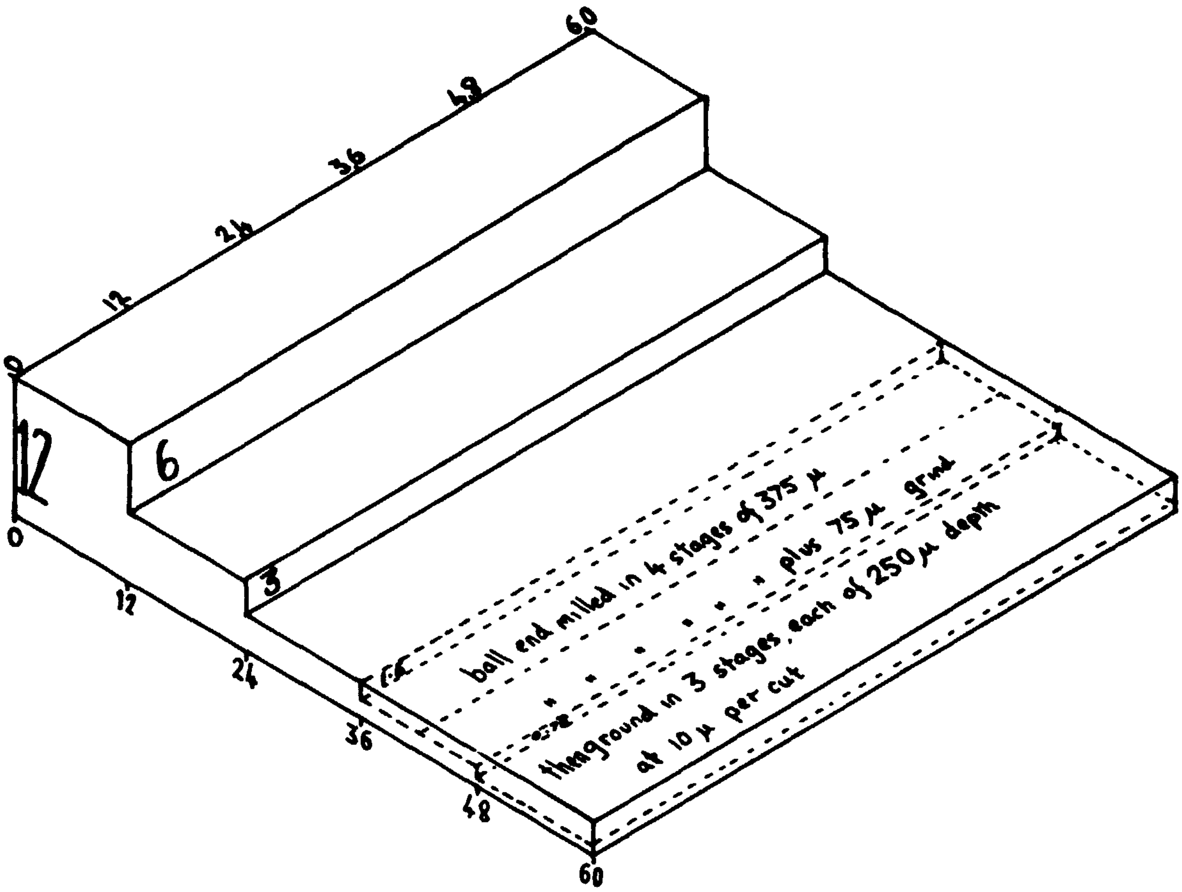

Figure 1: Recommended Configuration of Proposed Standard Specimen for Warpage Experiments. (Dimensions in mm.)

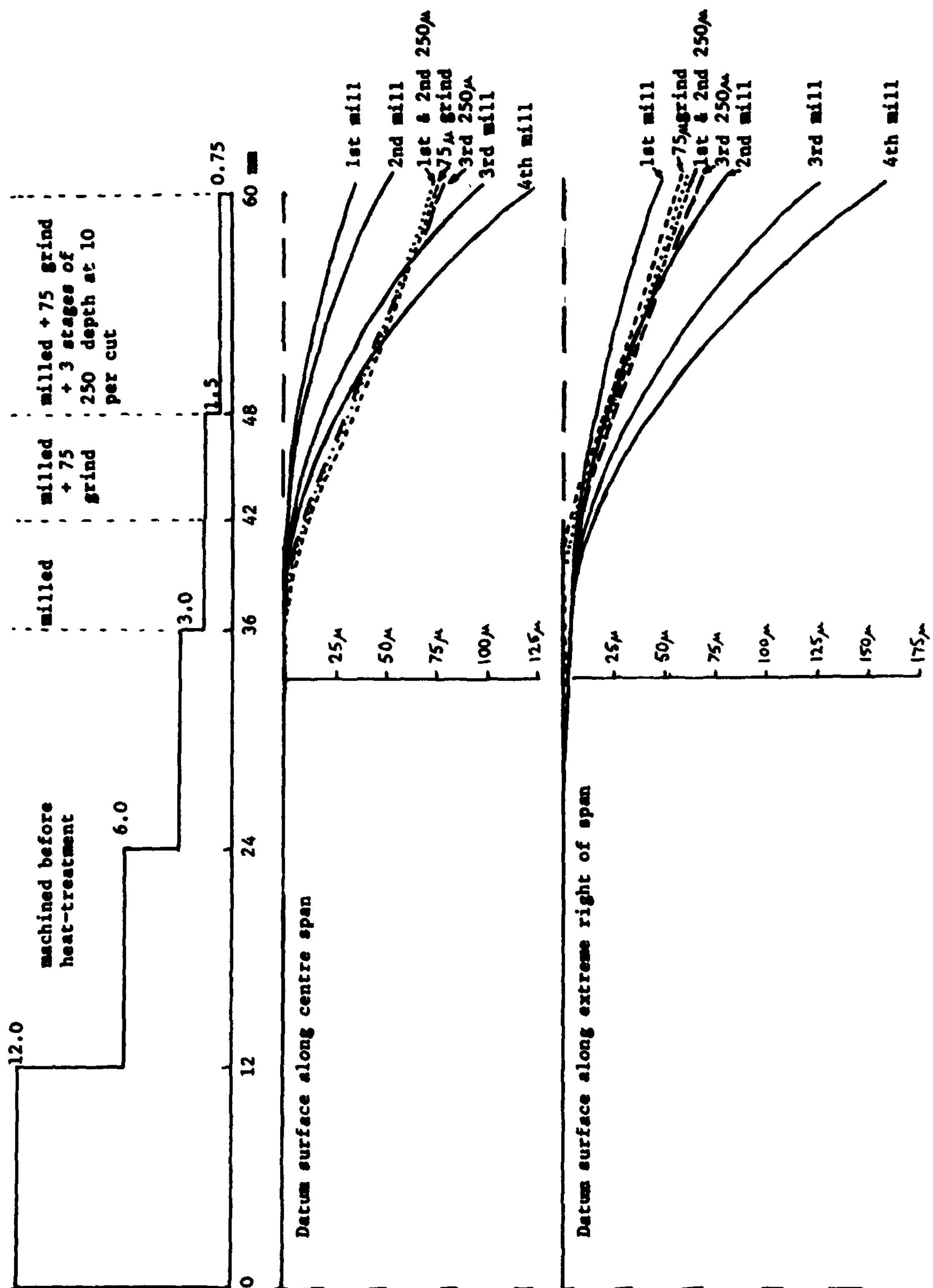

Figure 2 Longitudinate profiles of machined Vascomax 200

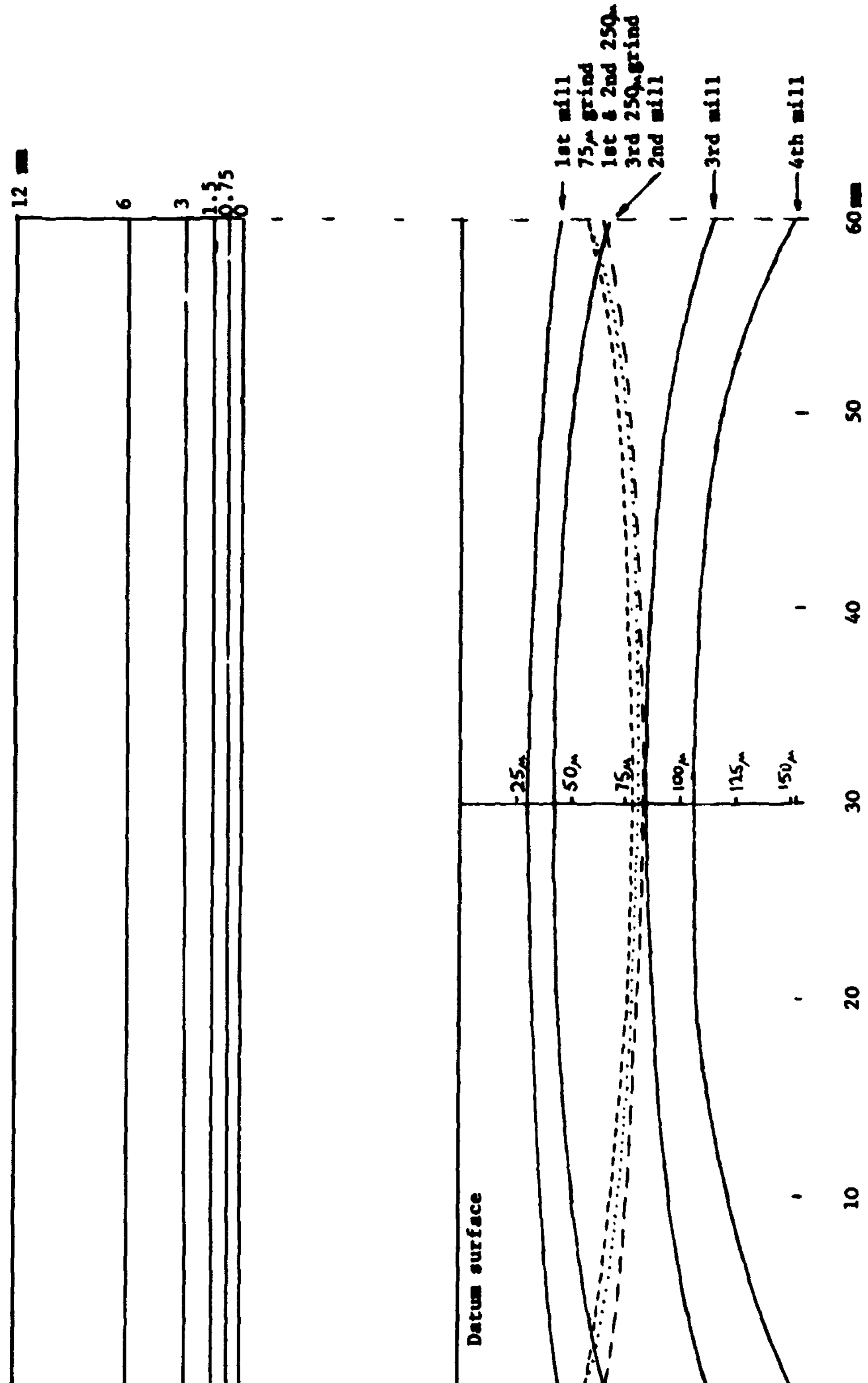

Figure 3 Transverse profiles of machined Vascomax 200

Material	Structure	Cryocycle stability	Mechanical stability	Strengthening mechanism	Heat-treat temperatures	Temperatures to avoid	Properties at -320F Cv ft lb	Yield Ksi	Comments
310SS 25Cr, 20Ni	austenitic stable	v. good M_s v. low	v. good M_D v. low	cold work	S.R > 1900F	1700-1100 carbide & sigma formation	100	76	Strength too low
304SS 18Cr, 8Ni	austenitic metastable	poor $M_s \simeq -100F$	bad M_D > RT	"	"	"	163	62	Low strength & liable to transform
321, 347SS weld.stab.18/8	austenitic metastable	suspect $M_s \simeq -200F$	poor $M_D \approx$ RT	"	"	"	130	42	Ti or Cb added against weld decay
316SS 18/10/2Mo	austenitic metastable	moderate M_s < -300F	suspect $M_D \approx$ RT	"	"	"	166	75	Better corrosion resistance
304N, 316N 347N	austenitic stable??	v. good M_s low	good M_D < RT	" $+N_2$	"	"	30	100	Nitrogen improves strength & stability
21-6-9Mn + N Nitronic 40	austenitic stable?	v. good M_s v. low	good M_D	" $+N_2$	"	"	65	150	Higher Nitrogen for higher strength. Good toughness
A286	austenitic	v. good	v. good	ppt.^{n} of Ti, Al	anneal 1800F & WQ, age at 1300-1400F.A.C	overage at temps > 1300F	50	120	Stable, good record, moderate streng'h
15-5PH H1025	martensitic	moderate	poor	martensite + ppt.^{n}	age at 1025F	Excessive time of temperatures above 1000F	2	220	Too brittle
15-5PH	martensite + unstable aust.	bad	bad	tempered martensite+ppt.^{n}	double age at 1400F then 1150		33	146	Reformed austenite transforms to martensite during cryogenic cycles & gives volume change
17-4PH H1150M	martensite + unstable aust.	bad	bad	tempered martensite+ppt.^{n}	double age at 1400 then 1150	overages temperatures over 1150F	(28)	(150)	

Appendix 1.1 Summary of information on structure properties and stability of materials for wind tunnel models

Material	Structure	Cryocycle stability	Mechanical stability	Strengthening mechanism	Heat-treat temperatures	Temperatures to avoid	Properties at -320F Cv ft lb	Yield Ksi	Comments
13-8Mo H1150M	martensite stable aust.?	suspect	suspect	tempered martensite+ppt.^{n}	double age at 1400F then 1150F	temperatures over 1150F	30	145	Possible more stable austenite
18Ni,200grade	maraging	good	good	ppt.^{n} in low carbon martensite	900F	900F & above over age	28	270	Non-stainless, marginal toughness, but good strength and stability
18Ni,250grade	maraging	good	good	"	"	"	10	320	Too brittle
9Ni, 4Co HP9-4-20	Q & T martensite	?	?	tempered martensite	1500F & WQ temper at 1025F	overages at temperatures > 1025F	25	183	Marginal toughness
Special 12Ni	ferrite/ martensitic	?	?	super fine grain, quenched and tempered		-	(50-60)	240	In development stage
Special 9Ni	ferrite/ martensitic	good	good	double normalized, quenched and tempered	tempered at 1050F	over tempers at > 1050F	80	145	Special development
Invar	austenitic	good	good	cold work			50	90	Low strength but very small thermal contraction
Inconel 718	austenitic	good?	good?	ppt.^{n} hardened	-	-	20	175	Need further consideration machining?
Inconel X750	austenitic	?	?	ppt.^{n} hardened	-	-	35	125	

Appendix 1.2 Summary of information on structure properties and stability of materials for wind tunnel models

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Material	Structure	Cryocycle stability	Mechanical stability	Strengthening mechanism	Heat-treat temperatures	Temperatures to avoid	Properties at -320F Cv ft lb	Yield Ksi	Comments
Ti-6Al-4V	α (hex.c.p) + β (b.c.c)	?	?	duplex structure + cold work	-	-	15	228	fabrication?
Ti-5Al-2.5Sn ELI	α stabilised hex. c.p	?	?	cold work	-	-	29	200	
Beryllium-Copper 2% Be	f.c.c	good?		ppt^n hardened	-	-	7	175	Poor toughness but excellent thermal conductivity
Aluminium 2024	f.c.c	good	good	ppt^n hardened	-	-	7	50	

Appendix 1.3 Summary of information on structure properties and stability of materials for wind tunnel models