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Future HFO/GI exhaust valve spindle

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Abstract: State of the art for exhaust valve spindles for large two stroke heavy fuel diesel engines is currently either a fully forged Nimonic 80A version or a cost effective version based on an austenitic valve steel weld coated by a specially hardened Inconel 718 seat hardfacing and Inconel 625 disc coating. These three alloys, originally developed more than 50 years ago for the gas turbine and process industry, show comparable corrosion resistance at usual heat load. The general trend in engine design is steadily pushing combustion chamber component temperatures towards higher levels and the hot corrosion resistance of these alloys is currently being tested to the limit. Furthermore operation on LNG will introduce new challenges. Indeed, it would appear that there is much room for improvement as no focused alloy development has been performed aimed at the special conditions found on the thermally and mechanically stressed parts of the exhaust valve spindle. In the present work new coating alloys, meeting the requirements for the future valve spindle, have been developed by combining litterature study, service experience, experimental data and numerical thermodynamic calculations. This paper describes the considerations and results of this alloy development as well as the details of a required new production technique for manufacturing a compound product by the Hot Isostatic Pressing (HIP) technology which has been developed applying advanced Finite Element Method (FEM) modelling.

INTRODUCTION

In the modern large bore two-stroke diesel engine, the exhaust valve spindle remains a thermally stressed component. The increase in thermal load is not only related to the ever increasing load density but is also seen in connection with engine service at low loads, which offer the benefit of a low fuel consumption, but which may also involve performance conditions with increased valve temperatures. Elevated temperatures adversely affect the service life and reliability of spindles because of "hot corrosion", a term which covers accelerated material loss caused by molten combustion products, such as Na_2SO_4 and V_2O_5, which rapidly dissolve the protective Cr_2O_3 oxide surface coating, Ref. [1]. This primarily takes place in a temperature interval of 650-750^{o}C. In practice this phenomenon is normally seen on parts of the spindle bottom which, depending on the temperature level, usually experience corrosion rates between 0.1 and 0.4 mm/1,000 hrs. and, in extreme cases, up to 1 mm/1,000 hrs. However, it can also be seen attacking the spindle seat, as a result of local temperature increases in connection with a local gas leakage.

Furthermore, additional challenges can be expected for the spindle seat operating in future two-stroke engines running on LNG and other clean fuels. The amount of combustion products will decrease, and so will the amount of seat deposits. The consequence is expected to be a reduced lubricating effect between the valve seat and the bottom piece seat, which is likely to introduce a new wear mechanism on the spindle seat.

This paper presents a novel design of an exhaust valve spindle which is intended to meet the above requirements and perform equally well on heavy fuel oil as on LNG and other clean fuels. The novel design is based on both a novel production technology and on novel materials.

A part of the research leading to these results has received funding from the European Union Seventh Framework Programme (FP7/2077-2011) under grant agreement no. 265861 (Helios), Ref. [2], while another contribution originates from an industrial PhD project at MAN Diesel & Turbo (MDT) and the Technical University of Denmark.

STATE-OF-THE-ART

Present design specifications of MAN B&W two-stroke diesel engines include the DuraSpindle and Nimonic spindle as standard components. The DuraSpindle is a welded design based on an austenitic valve steel with welded Inconel coatings, including a special hardened-seat coating. The Nimonic spindle is fully forged with usually the entire spindle being produced of Nimonic 80A. In the large-bore version, the service performance of the two versions is quite comparable. However, the DuraSpindle offers superior strength and wear resistance, while the Nimonic seat offers superior corrosion resistance.

At usual heat loads, both types of spindles perform reliably with service intervals of usually 30,000 hrs. The deciding factor for the service interval is normally the corrosion rate on the spindle disc bottom, which typically is found in the interval between 0.1 and 0.4 mm/1,000 hrs. This interval relates to service conditions when running at a relatively high load. Low load conditions however may involve drastically increased corrosion rates on the disc bottom, and figures up to 1 mm/1,000 hrs have been reported.

In addition, a recent engine design has implied a general increase of the temperature level of both the disc bottom and the seat area. As illustrated in Fig. 1, local burn marks on the spindle seat have been seen in a few cases.

Figure 1 – Local burn mark on spindle seat created by high temperature corrosion

NEW DESIGN CONCEPT

Materials with low wear rates under severe high temperature corrosive conditions have been known for more than 50 years, the most prominent being high Cr bearing Ni-base alloys and high Ni bearing Cr-base alloys. These alloys are also known for becoming brittle at temperatures in the range of 700^{o}C when fabricated by conventional production processes, Ref. [3]. However, during the mid 1990s, MDT demonstrated that powder metallurgical (P/M) Ni48Cr compacted via hot isostatic pressing (HIP) retained excellent mechanical properties even after more than 4,000 hrs. exposure at 700°C, Ref. [4].

With respect to the seat hardfacing, a review of commercially available hardfacing alloys led to the conclusion that in order to meet the requirements for a high corrosion resistance and wear resistance, it would be necessary to develop a custom-made material.

As to the wear resistance of the seat coating, it is important to acknowledge that, so far, the nature of a possible wear mechanism of the seat during service on e.g. LNG has not yet been identified. Therefore, it is anticipated that a possible subsequent modification of the seat material is necessary to meet the wear requirements. The necessary flexibility for such a modification is evaluated to be best met within the P/M technology, specifically hot isostatic pressing (HIP).

Hence, the overall object of the new design concept is to combine the two above-mentioned coatings in a compound disc in which the coatings are fully dense, have a metallurgic bond to the substrate and have bonding zones with satisfactory mechanical properties.

<u>The HIP process</u>

The HIP production technology can produce fairly large components based on a welded capsule design applying thin sheet material of either low C steel or austenitic stainless steel.

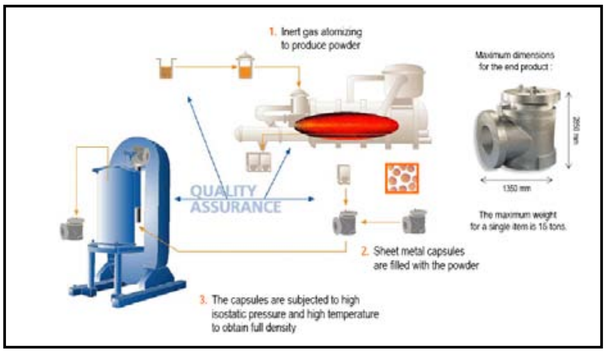

Figure 2 – HIP P/M sheet metal capsule production

The process is shown in Fig. 2, where inert gas atomised metal powder is filled into a capsule via a pipe. The capsule is subsequently evacuated via the same pipe which is then sealed. The capsule is finally processed (compacted) inside the HIP chamber at appropriate process parameters of typically 3 hrs. at 1,150°C and 1,000 bar. After processing the capsule, the volume has decreased by approximately 30% corresponding to the original void space between the powder particles.

DEVELOPMENT OF P/M MATERIALS

In order to rank the corrosion resistance of current exhaust valve materials, a modified DuraSpindle for a marine two-stroke engine with a cylinder diameter of 900 mm was produced according to the layout shown in Fig. 3. At the bottom of the spindle, 54 samples of 9 different Ni-based Ø11 alloys were embedded in a P/M matrix consisting of Ni49Cr1Nb. The composition of the tested alloys is shown in Table 1. Common for the alloys was that they contained 17 wt% Cr or more. Some alloys were present in both a welded ("W") and a forged ("F") state. Samples were positioned at radial positions Ø40 (centre), Ø100, Ø160, Ø220, Ø280, Ø340 and Ø400 mm. Some samples were not placed at position Ø40, and due to practical reasons during spindle production, NiCr22WAl was not placed at positions Ø280, Ø340 and Ø400. After hot isostatic pressing, the spindle was put into active service on the engine of an ocean-going vessel for a total of 2,223 hrs. In order to establish a worst-case, the cylinder was fitted with non-optimal fuel atomisers, which are known to increase the temperature experienced by the spindle bottom.

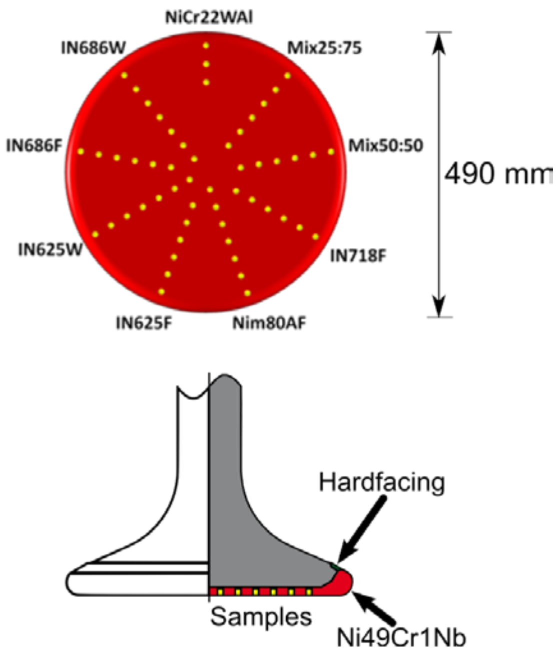

Figure 3 – The design of the modified DuraSpindle with HIP compound embedded samples configured for corrosion testing. Position of corrosion samples was named by their position on the diameter, spanning from Ø40 to Ø400

After the test period, the spindle was extracted and inspected. The material loss was quantified by measuring the hole depth and compensating for the loss of the surrounding Ni49Cr1Nb. This correction was possible, because Ni49Cr1Nb was not corroded to any measurable degree in the center and perimeter of the spindle bottom. On those positions, turning marks were still visible, showing the original level of the machined surface. Using this reference, the material loss of Ni49Cr1Nb could be measured at the intermediate positions.

Table 1 – Composition of material samples in wt% as specified by material certificates. The composition of the mixed metal powder samples Mix25:75 and Mix50:50 are calculated as a weighted average of the two powders. IN686F and NiCr22WAl contains 4 and 6% W, respectively

Material	C	Si	Cr	Fe	Mo	Ti	Nb	Al
Ni49Cr1Nb	0.01	0.05	48	-	-	-	1.04	-
IN625F	0.03	0.21	22	4.08	8.7	0.29	3.61	0.22
IN625W	0.01	-	22	0.48	11.4	0.05	3.22	0.09
IN686F	0.01	0.1	17	0.15	16.7	0.11	0.03	0.29
IN686W	0.01	0.3	19	2.09	18.3	0.05	0.14	0.09
IN718F	0.03	0.14	19	17.3	3	0.94	5.21	0.49
Mix25:75	0.02	0.31	29	3.23	6.8	0	2.83	0
Mix50:50	0.02	0.23	35	2.15	4.5	0	2.23	0
NiCr22WAl	0.42	0.1	20	0.21	0.1	0.06	-	6.07
Nim80AF	0.06	0.08	19	0.22	-	2.41	-	1.7

The material loss shown in Fig. 4 was most severe at the intermediate positions, with a clear maximum loss at Ø220. Samples at Ø400 were generally less corroded than those closer to the centre at Ø100. Samples of NiCr22WAl, which were only included at positions Ø400-Ø280, show the same trend as the other samples with increasing material loss towards Ø220. For each position, the alloys performed similarly, with some fluctuations at lower material loss levels. The most resistant alloy at all positions was Ni49Cr1Nb, which shows no corrosion at positions Ø40, Ø340 and Ø400. IN718F was more severely attacked than all other alloys for all positions except for Ø340, where IN686F suffered a slightly higher material loss. Most of the alloys contained approximately 20 wt% Cr, and they perform comparably with the exception of Nim80AF, the corrosion resistance of which is comparable to that of Mix50:50, which contains 35 wt% Cr on average.

Ni-base disc bottom coating

The above corrosion test clearly illustrates the superior corrosion resistance of the Ni49Cr1Nb material and, as illustrated in Fig. 5, earlier investigations have proven the suitability as an engineering material under prolonged service conditions at 700°C. Contrary to material produced by conventional production methods, the P/M HIP compacted material retains a high ductility level. Therefore, this material has been selected as the coating material for the firing side of the spindle disc.

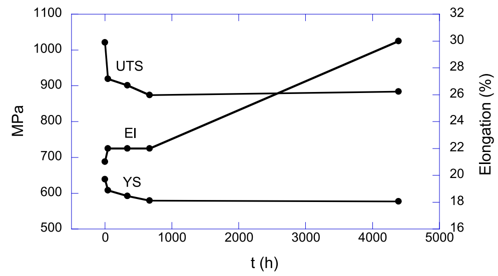

Figure 5 – The influence of heat treatment of Ni49Cr1Nb at 700°C on mechanical properties

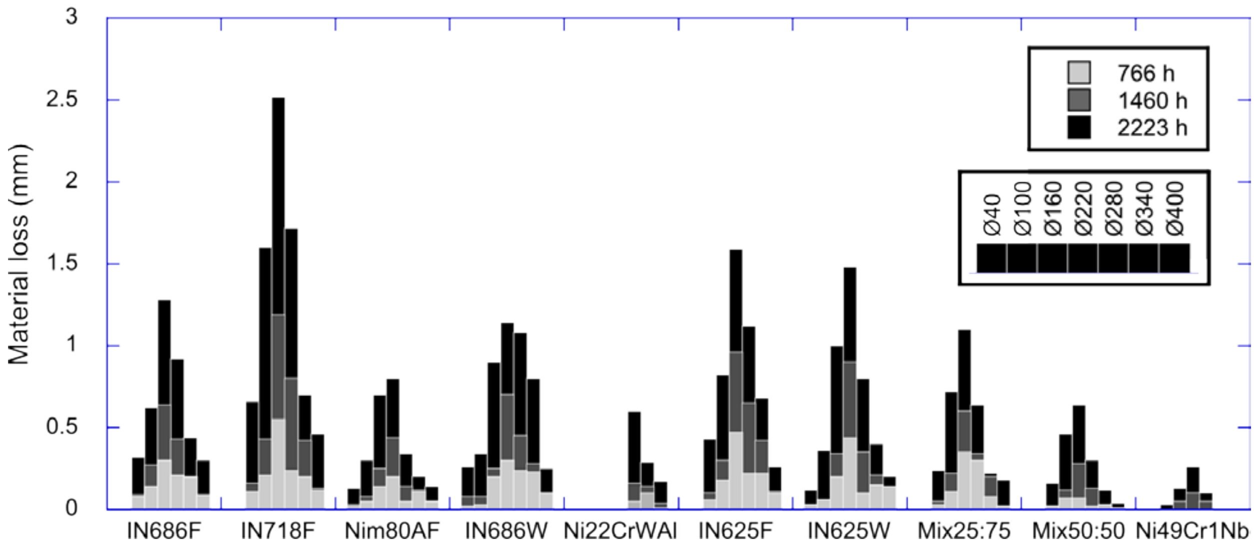

Figure 4 – Material loss of each alloy specimen at the positions on the spindle diameter

For the Ni49Cr1Nb charge applied in the actual project, the mechanical properties and the relevant physical properties have been established. These properties form the basis of the material data applied as boundary conditions in the FEM simulations.

The mechanical data shown in Fig. 6 is regarded as satisfactory, whereas the physical properties, density and thermal expansion coefficient shown in Figs. 7 and 8 revealed a feature that required special consideration during the subsequent FEM simulation.

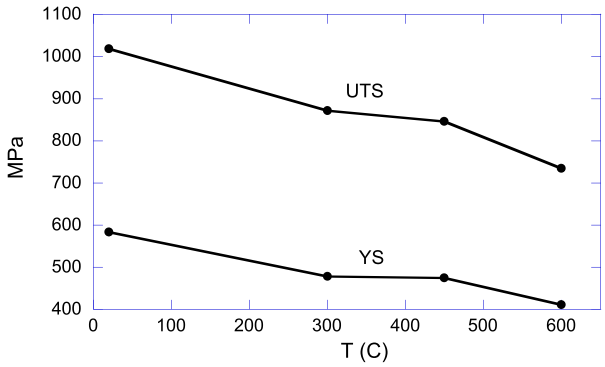

Figure 6 – Yield strength (YS) and tensile strength (UTS) dependency on test temperature for P/M HIP Ni49Cr1Nb. Elongation values remain constant at 23% for the tested temperatures

A dependency of time at service temperature of the density and thermal expansion coefficient was observed.

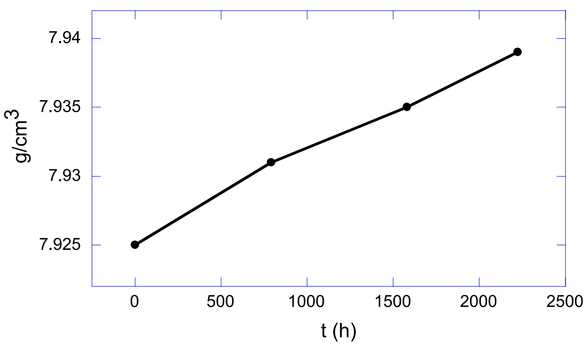

Figure 7 – Density of Ni49Cr1Nb exposed at 700°C

The time-dependent density change is most likely related to Ni_2Cr ordering, which has been shown to take place in high Cr Ni based alloys over time at elevated temperatures, Ref. [5].

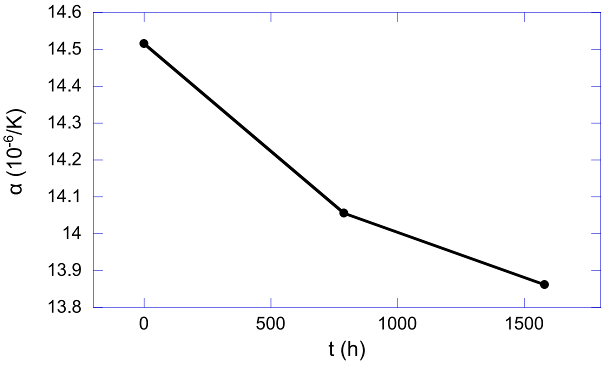

Figure 8 – Thermal expansion coefficient measured from 30-700°C for Ni49Cr after exposure at 700°C

Also, the initial phase distribution of the two phases (α-Cr and γ-Ni) present in the alloys is different from the phase distribution established during the process temperature, shown in Figs. 9 and 10.

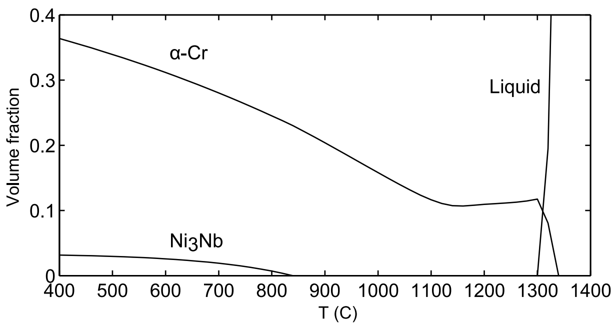

Figure 9 – Phase distribution of Ni49Cr1Nb calculated with Thermo-Calc using the TCNI5 database. γ-Ni phase is balance

The phase composition is calculated with Thermo-Calc using the TCNI5 database, which contains empirical free energy functions, built in the manner described by Saunders and Midownik, Ref. [6].

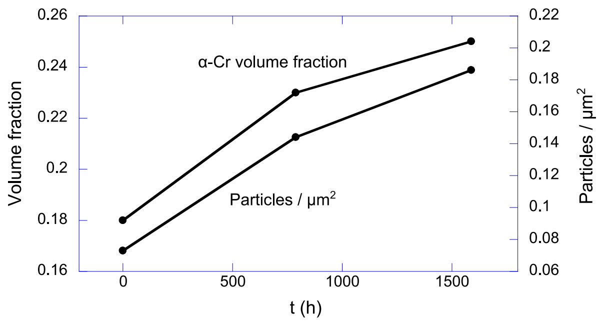

Figure 10 – The influence of heat treatment at 700°C on the α-Cr content of Ni49Cr1Nb

Diffusion causes successive phase redistribution towards the equilibrium condition at service temperature.

New Ni-base seat hardfacing

The very low crack susceptibility and high ductility of welded Inconel 718 (nominal composition 19%Cr, 18%Fe, 5.2%Nb, 0.5%Ti and 0.5%Al, bal Ni) are decisive material properties when performing the combined work hardening and precipitation hardening of the DuraSpindle seat hardfacing. It is possible because the alloy has a sluggish precipitation response, which requires several hours of heat treatment to complete. During such a heat treatment, which typically takes place at 650-750ºC, finely distributed nano-scale particles of Ni_3Nb (γ'') are formed. The structure of these particles is semi-coherent with the atomic matrix of the alloy. This impedes crystal deformations, contributing greatly to the strength and hardness of the alloy. It is possible to calculate the stable phase composition of Inconel 718 with the commercial software Thermo-Calc. The phase composition has been calculated for Inconel 718 and for a number of derived alloys in which the Fe and Mo content is replaced with Cr, see Fig. 11.

The results of the corrosion test and literature (Ref. [7]) confirm that a decrease of Fe and increase of Cr should vastly increase the hot corrosion resistance of Inconel 718. It is clearly illustrated in Fig. 11 that the γ'' content by and large is not affected by a higher Cr content, while the volume fraction of the α-Cr phase is greatly increased. However, the question of whether this is a problem remains. The α-Cr is traditionally regarded as brittle and weak compared to the bulk Ni matrix. However, recent work by Dong et al, Ref. [8], indicates that the morphology of the α-Cr heavily influences the mechanical properties, and indeed experiments shown later in this paper, with P/M specimens of Ni49Cr1Nb with a high amount of globular α-Cr particles, show that an elongation of up to 20 % is not uncommon.

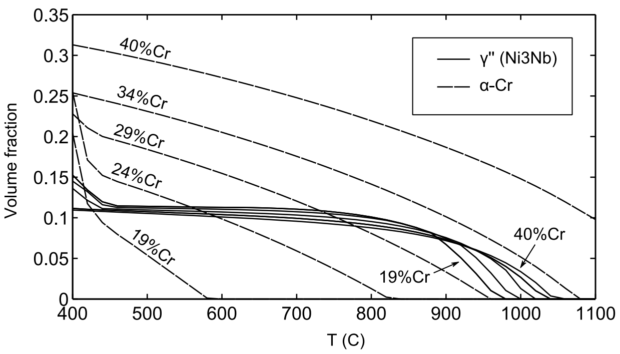

Figure 11 – The phase composition of Inconel 718 when Fe and Mo are substituted with Cr, and Nb content is held constant. Results from Thermo-Calc, using the TCNI5 database

An iterative process was carried out to investigate whether it is indeed possible to create a precipitation hardenable Inconel 718 derivative with twice the Cr content without disastrous loss of elongation and, thereby, processability. The iterative process included a factorial experiment with 5 P/M alloys. The results were used to elucidate a 6th composition named Ni40Cr3.5Nb. For comparative purposes, both iterations are shown together in the following. Table 2 shows the chemical composition of the tested alloys.

Table 2 – Chemical composition in wt% of powder metallurgical alloys as specified by the supplier certificate. Ni bal

Alloy	*Cr*	*Nb*	*Ti*	*C*	*O*	*N*	*Al*
Ni35Cr4Nb	*34.4*	*3.8*	*0.5*	*0.011*	*0.027*	*0.002*	*≤0.02*
Ni35Cr6Nb	*35.2*	*6.0*	*0.5*	*0.005*	*0.019*	*0.001*	*0.01*
Ni40Cr3.5Nb	*41*	*3.4*	*0.5*	*0.007*	*0.009*	*0.013*	*0.12*
Ni40Cr5Nb	*39.3*	*5.0*	*0.5*	*0.014*	*0.032*	*0.001*	*≤0.02*
Ni45Cr4Nb	*44.8*	*4.1*	*0.5*	*0.013*	*0.032*	*0.001*	*0.01*
Ni45Cr6Nb	*44.5*	*6.0*	*0.5*	*0.013*	*0.025*	*0.001*	*≤0.02*

The alloys were HIP'ed to form Ø20 rods. These rods were solution treated by heating to 1,100ºC followed by water quenching. Aging was performed at 700ºC for up to 9 hours. The test methods were conventional tensile testing and 10 kgf Vickers hardness (HV) measurements. The hardness measurements performed during the heat treatment are shown in Fig. 12. All alloys responded rapidly to the aging treatment with a hardness increase of at least 100 HV within the first 2 hours. The high Cr high Nb alloys rapidly reached a high plateau from which only marginal improvement occurred.

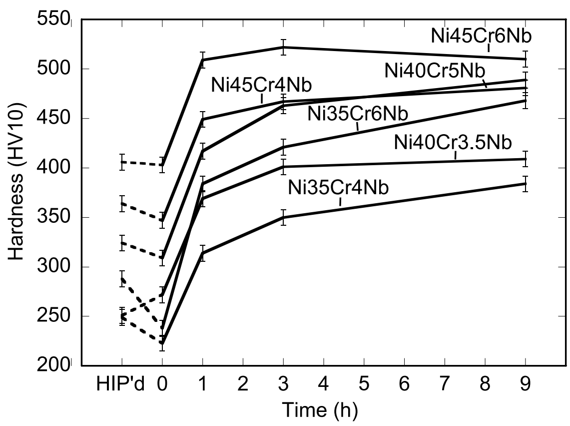

Figure 12 – Hardness of alloys during heat treatment. Time 0 corresponds to the solution treated state (1,100ºC followed by water quenching). Aging performed at 700ºC

The results from mechanical testing are shown in Table 3, with literature values for similarly treated P/M Inconel 718. It appears that there is a cross-influence between the Cr and Nb content. Also, Ni35Cr6Nb appears to have almost the same properties as Inconel 718, Ref. [9].

Table 3 – Mechanical properties of the alloys

Alloy	Solution treated YS (MPa)	Solution treated UTS (MPa)	Solution treated El (%)	Aged YS (MPa)	Aged UTS (MPa)	Aged El (%)
Ni35Cr4Nb	436	840	52.5	913	1175	19.1
Ni35Cr6Nb	443	874	49	1210	1382	6.1
Ni40Cr3.5Nb	533	939	39.3	990	1227	20.5
Ni40Cr5Nb	646	1059	24.2	1305	1444	3.2
Ni45Cr4Nb	-	-	-	1274	1416	4.6
Ni45Cr6Nb	-	-	-	1425	1502	1.3
Inconel 718	480	975	38	1260	1413	8.6

Considering the results from mechanical testing and hardness measurements, it seems that the mechanical properties of Ni35Cr6Nb, by and large, correspond to the performance of Inconel 718, while nearly doubling the Cr content. However, Ni40Cr3.5Nb will, because of the higher Cr content, have a superior hot corrosion resistance and, due to the high ductility in the aged condition, there appears to be a potential for a combined work and precipitation hardening which will further increase the strength level. As these samples are produced by P/M processes, it is possible to mix foreign particles into the powder before consolidation, leading to a bulk alloy with embedded functional particles which could improve the abrasive wear resistance.

<u>Bonding layer</u>

During the compaction in the HIP chamber, diffusion processes are responsible for the metallurgic bonding between substrate and coating. Elements diffuse across the bonding zone driven by activity gradients. Depending on the atomic number, diffusion distances of up to 1 mm have been observed in HIP'ed components. As the C content of the valve steel substrate is far higher than that of the coatings (0.25 vs 0.01 wt%), a direct coupling would lead to C-diffusion from the valve steel into the Ni-base coating and extensive chromium carbide precipitation in the bonding zone. The consequence would be a brittle bonding zone with inferior mechanical properties. Consequently, an intermediate zone with a bonding material must be established between the valve steel substrate and the coatings in order to hinder the detrimental effect of C-diffusion. For this purpose, an austenitic stainless steel (AISI 316L) with a C content of 0.025 wt% has proved to be very effective. As shown in Fig. 13, no appreciable C-diffusion has taken place across the bonding zone with the Ni-base coating during the HIP process. The C-diffusion into the AISI 316L at the bonding zone towards the valve steel causes a moderate carbide precipitation which is evenly distributed in the microstructure. This results in a strengthening of the AISI 316L that is comparable with the strengthening of C in the valve steel.

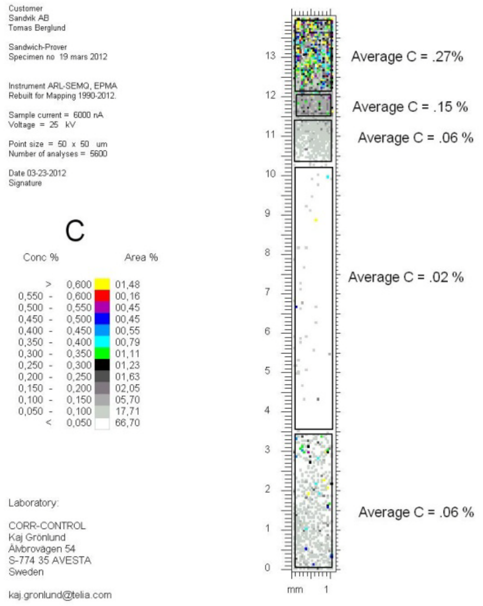

Figure 13 – XRD mapping of the bonding zone between the valve steel substrate (top) and Ni49Cr1NB (bottom) illustrating the magnitude of C-diffusion into the AISI 316L (centre)

Tensile tests of bonding zones have verified a sound and ductile bonding with the majority of the elongation taking place in the bonding material and the fracture position within the bonding material away from the bonding zones.

Direct bonding of the Ni49Cr1Nb and Ni40Cr3.5Nb coatings has been performed without any bonding layer showing satisfactory mechanical properties.

PRODUCTION TECHNOLOGY

The present capsule design involves both forged and powder materials made of four different materials, and when deciding the HIP process parameters, both economical and metallurgical considerations need to be made. Full compaction and bonding can usually be assured during a relatively short process time of 3 hours operating at 1,000 bar and a temperature above 1,000^{o}C.

Furthermore, the process includes ramping up and down for at least another 3 hours. Fixing the process time and pressure at 3 hours and 1,000 bar respectively, the process temperature was set to variable due to its prominent influence on the metallurgic processes. With the criterion of a temperature as low as possible, below 1,150°C on grounds of the microstructure of the forged substrate, and the criterion for optimum mechanical properties of the bonding zone and the coating materials, the possible process temperature range came out as 1,050-1,150°C.

HIP compound capsule design

In case of the compound spindle disc, it was decided to apply a forged disc substrate which accounts for the majority of the disc capsule volume. The advantage of this solution, compared to a solution applying only powder, is that the total capsule shrinkage during HIP is relatively low, allowing for a near-net-shape production.

Furthermore, small tolerances for the coating geometries can easily be obtained. A solution based solely on powder would involve a more complicated capsule design. While preparing the prototype capsule design, the major concern is to assure that no unintended mixing of dissimilar metal powders takes place and that this can be done in the most economical way. Once these principles have been established, the next concern is the stress and strain condition of the finished valve during service. This information is obtained by FEM simulation and may lead to changes in the capsule design. On the basis of input from e.g. production and FEM modelling, the capsule design is going through an iterative process until service tests have confirmed a sound design. The present prototype capsule design is presented in Fig. 14, and this design will be applied on the first prototype spindle intended for service tests.

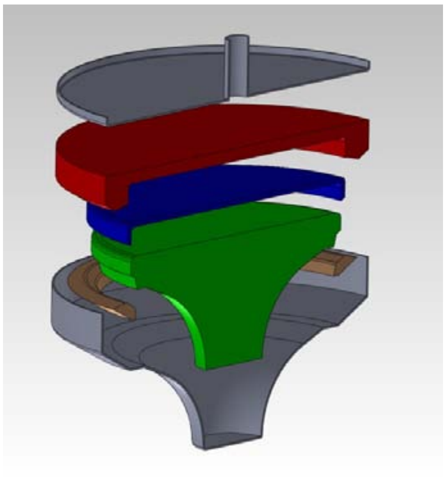

Figure 14 – The capsule design. The grey capsule parts are produced as deep drawn steel sheet being assembled by gas tungsten arc welding (GTAW). The forged disc substrate is green while the coating materials are red and orange, respectively. The bonding zone material is blue

EVALUATION OF COMPOUND DISC DESIGN BY FEM

The prototype valve is designed in four separate parts, see Fig. 15. The valve steel is separated from the disc coating and seat material by a buffer layer. All parts have an outer skin simulating the capsule design. The outermost skin is removed in a final machining operation to give the valve spindle its final geometry.

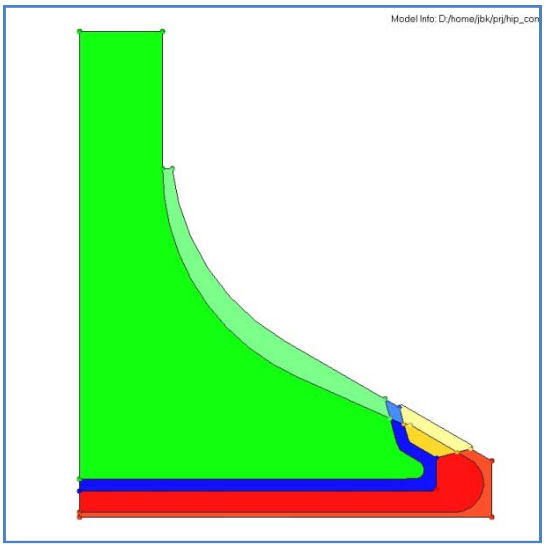

Figure 15 – Prototype geometry: green = valve steel, blue = buffer layer, yellow = seat material and red = corrosion resistant material

The entire valve is modelled using three-noded axi-symmetric triangles CAX3, Ref. [10]. The model characteristics are listed in Table 4. Two different meshes are used to analyse different aspects of the valve service. The stress-strain analysis is made using the fine mesh to get a good resolution of stress and strain gradients. The coarse model is used for a creep analysis, i.e. to analyse the deformation of the spindle during its service life.

Table 4 – FEM model characteristics

	Fine model		Coarse model	
Part	**Nodes**	**Elements**	**Nodes**	**Elements**
Valve	49,455	97,686	2,113	3,959
Valve skin	1,280	2,254		
Buffer	26,242	50,634	1,233	2,094
Buffer skin	144	232		
Seat	5,205	9,998	222	359
Seat skin	948	1,715		
Corrosion layer	24,214	47,055	510	827
Corrosion layer skin	1,816	2,924		

The HIP process is simulated by applying the HIP temperature and pressure. Afterwards the temperature is stepped down to room temperature. During the last stages of the simulation of the manufacturing, the outer skin elements are removed and the seat material is replaced. The latter simulates the phase change resulting from hardening the seat material.

The residual stress field after finish-machining is illustrated in Fig. 16. The results show that the area where three different materials meet is heavily loaded.

Therefore, an updated geometric design of the seat area is under development. However, it should be noted that the exact magnitudes of the peak stresses at the corner, and in the material interfaces, come with an added degree of uncertainty. Some local mixing of the powder materials is to be expected at the interface. The mixing results in a gradient transition, which is not included in this model.

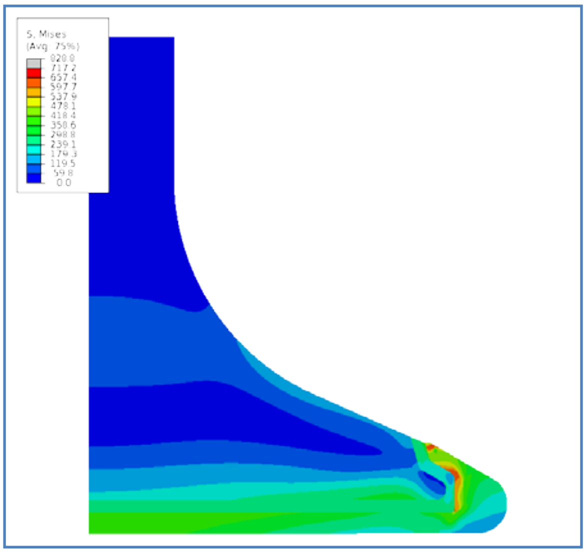

Figure 16 – Residual stress after finish machining

The simulation then continues by applying a number of "operating" cycles. Each cycle consists of:

- heating up the engine to operating temperature (Fig. 17)
- applying combustion pressure
- cooling the engine down to room temperature.

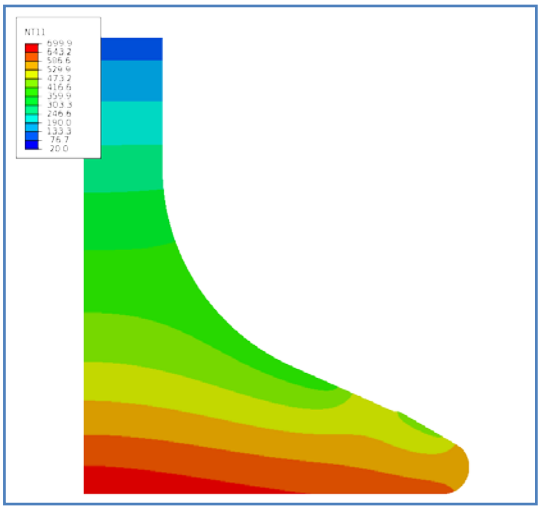

Figure 17 – Operating temperature

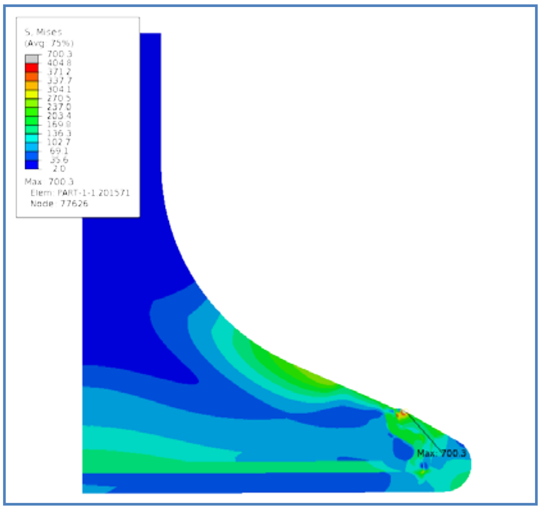

Figure 18 – Von Mises stress for hot engine with combustion pressure applied

Fig. 18 shows von Mises stress for the first operating cycle. The loading consists of operating temperature and combustion pressure. According to the simulation, the highest stresses and strains are located at the seat contact. Although, high stresses are expected in this region, the numerical maximum values shown are unrealistic and a consequence of the numerical model. In reality, the seat will deform plastically and distribute the applied load over a larger area. The figure also displays high levels of stress at the point where the residual stresses after finish-machining are high. Again, it is expected that local mixing of powders will influence the level of the maximum stress and strain towards lower values than calculated by the model. However, the simulation clearly indicates the necessity for a detailed analysis of this position.

To conclude, the initial simulations of the new valve design show that this concept generally is performing satisfactorily. One highly stressed region, where three materials meet, has been identified, and a remedy is about to be implemented. Furthermore, the first creep simulations are under evaluation, and a model taking care of time dependency of material properties is under preparation.

OUTLOOK

The latest design of the novel HIP compound valve spindle represents an important milestone as presently four prototype spindles of this design are under production. One spindle will be applied for verification of the capsule design and material properties. In addition, this spindle will be used for developing the procedure for the combined work- and precipitation hardening of the new seat alloy.

The remaining spindles are intended for service tests to evaluate their performance. Two of these spindles will also include corrosion tests on the disc bottom similar to the aforementioned corrosion test, however with a number of new materials including several versions of the new seat material.

At a later stage, preparations will be made to produce spindles for a first service test on a dual fuel (LNG) engine in order to document the service performance of the seat and to identify the wear mechanism, should it be different than the present one. If required, these data will represent the basis for the continued development of the seat material to a wear-optimised version.

Under the condition that the design is confirmed, the new spindle, when entering regular service, is expected to represent an increase of the design temperature of up to 50^{o}C for both the disc face and seat hardfacing.

In conclusion, it is important to note that the present application of an advanced production technology has the implication that reconditioning of this type of spindle, using presently available technology, cannot regenerate the original properties.

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